

**Advanced Laser Weld Brazing of Zinc-Coated Automotive Steels:
Process Optimization for AI-Enhanced Inspection, Tailored
Intermetallic Formation, and the Evolution of Mechanical Behavior**

by

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Author's Declaration

This thesis consists of material all of which I authored or co-authored: see Statement of Contributions included in the thesis. This is a true copy of the thesis, including any required final revisions, as accepted by my examiners.

I understand that my thesis may be made electronically available to the public.

Statement of Contributions

This thesis consists of 8 chapters, including the introduction (Chapter 1) and literature review (Chapter 2). The data chapters, Chapters 3 to 6, consists of published or submitted manuscripts that the candidate wrote as the primary author. The manuscripts have been modified to ensure consistency in figure and reference formatting and to harmonize abbreviations throughout this thesis. The majority of the lab work (except where noted), including, but not limited to, laser operation, process optimization, sample preparation, characterization, and mechanical testing was done by the candidate. The writing and initial editing of the manuscripts (except where noted) was also done by the candidate. The following co-authors have made contributions to the present work, as outlined below:

Professor Elliot Biro supervised the research progress, advised the research direction, reviewed and edited all manuscripts.

Dr. Sheida Sarafan from National Research Centre of Canada, Montreal, Canada, supervised the work, reviewed and edited all manuscripts.

Chapter 3

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The candidate's contribution was as described above. Dr. M. Shehryar Khan collaborated with the candidate and was involved in technical discussions and reviewed the manuscript. O. Sherepenko worked with the candidate and provided advice on fixture design and during preliminary trial-and-error experiments with laser welder robot. M. Lee provided the base material. P. Wanjara reviewed the manuscript. E. Biro's role was as described above.

Chapter 4

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Abstract

Laser weld brazing (LWB) is a key joining process in automotive manufacturing, offering minimal substrate melting, reduced Zinc (Zn) coating burn-off, and superior joint appearance. However, challenges such as sub-surface defect formation and intermetallic compound (IMC) embrittlement hinder its optimization and broader adoption. Due to LWB's relative novelty for Zn-coated steels, limited research has explored brazing defects, segregation behavior, and IMC formation, as well as their direct impact on mechanical performance. The potential for maximizing braze joint mechanical performance is limited. Even with the elimination of interfacial premature failures caused by the brittle IMC layer, the extent of mechanical performance improvement in defect-free joints remains confined to the strength of the brazing filler material. Moreover, the role of IMCs and any second phases forming within the braze matrix in either degrading or improving mechanical performance within the literature is inconclusive. This work systematically investigates LWB across Zn-coated steel grades using Si-bronze filler wire, advancing the understanding of laser-wire-substrate interactions that dictate microstructure and mechanical behavior, to maximize mechanical performance of LWB joints. Additionally, it establishes a novel LWB optimization approach that enables modification of microstructure of LWB joints as well as improvement of an in-line non-destructive testing (NDT) method assisted by artificial intelligence (AI) for real-time defect inspection and precise braze geometry measurement.

Comparative analysis with gas metal arc brazing (GMAB) confirms that LWB significantly enhances wettability of molten filler by refining the surface morphology of substrates and alters elemental segregation. This study introduces a novel wire-adjusted heat input (HI) approach, a systematic optimization strategy that directly correlates developed HI-related parameters with defect formation, overcoming inconsistencies observed in conventional nominal heat input methods. This approach enables accurate distinction between defect-free and defective seams, including lack of adhesion (LoA), pores, base metal melting (BMM), and non-conforming geometry. Process window validation through three-dimensional (3D) X-ray micro-computed tomography (μ CT) using semi-automated pore segmentation and statistical

analysis of pore evolution, supports the robustness of the optimization strategy. The braze joints made within the process window successfully enabled acquisition of reliable UT signal feedback for defect detection. The two-dimensional (2D) slices of μ CT data, enriched AI training data base, effectively contributed into AI predictions of the seam geometry from the UT data.

Another major breakthrough of this study is developing a novel IMC category, termed surrounded interface-IMCs (SI-IMCs), formed under controlled elevated HI, distinct from conventional interface-IMCs (I-IMCs). A critical $HI_{Relative}$ threshold of 32 J/mm is identified for defect-free brazing in roof joints of GI-IF steels, with an additional 12.44 J/mm required to promote SI-IMC formation, occupying up to $38.2 \pm 16.9\%$ of the braze cross-sectional area. These SI-IMCs, consisting of a shell-like Fe-Si layer and a (Fe-rich)-Cu eutectic phase, enhance mechanical properties. Increasing SI-IMC area fraction from $1.2 \pm 2.4\%$ to $38.2 \pm 16.9\%$ results in a 14% increase in tensile peak load, a 350% increase in displacement, and improvement of joint toughness by 525%.

The complex geometry of roof joint coupons, combined with the low strength and high ductility of galvanized coated interstitial free (GI-IF) steel substrates, makes it difficult to fully attribute the improved mechanical performance to the presence of SI-IMCs. To fully elucidate the contribution of SI-IMCs to joint mechanical performance, this study conducts the first comprehensive investigation into a wide range of SI-IMC structures formed at different $HI_{Relative}$ levels during LWB in a lap joint configuration for GI-coated Gen 3 Q&P980 steel. It provides entirely novel insights into characterization of SI-IMC structures, the Cu-rich braze matrix, and how they impact mechanical performance. SI-IMCs formed at lower $\Delta HI_{Relative}$ (2.3–7.5 J/mm) improve mechanical performance, increasing peak load by 91.7%, displacement by 319.8%, and toughness by 1200%, whereas those formed at higher $\Delta HI_{Relative}$ (13–17.5 J/mm) degrade properties. Systematic elemental and texture analyses identify five SI-IMC types (α , β , γ , δ), with fine α -, β -, and γ -SI-IMCs delaying failure, while coarse α - and δ -SI-IMCs accelerate it. A strong {100} texture alignment between SI-IMCs and the Cu-rich

brazing matrix enhances strain accommodation and crack resistance. The newly introduced *GND.Ratio (Cu/Fe)* quantifies dislocation density, revealing that when *GND.Ratio (Cu/Fe)* > 0.67, SI-IMC-induced twinning activates, locally rotating the Cu-rich matrix orientation to <112> and forming twins near crack paths. This transition shifts fracture behavior from brittle to ductile, enhancing mechanical performance. These findings redefine the role of IMCs and second phases that are formed in the brazing interior, demonstrating their potential to optimize strain distribution, strain-hardening behavior, and crack propagation. Tailoring SI-IMC structures strategically can further enhance the mechanical performance and reliability of multi-material automotive structures.

Keywords: Laser weld brazing (LWB); Zn-coated steels; AI-friendly process optimization; Heat input (HI); Microstructure; Texture; Fracture analysis; Mechanical properties; Intermetallic compound (IMC); Pore; Dislocation; Twin.

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Dedication

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List of Abbreviations¹

AHSS	Advanced High Strength Steels
AI	Artificial Intelligence
BCC	Body Centered Cubic
BIW	Body-in-White
BMM	Base Metal Melting
BSI	Back Scatter Imaging
BoP	Bead-on-Plate
CI	Confidence Interval
CR	Cooling Rate
DIC	Digital Image Correlation
DL	Deep Learning
DP	Dual Phase
EBSD	Electron Backscattered Diffraction
EDS	Energy-Dispersive
EPMA	Electron Probe Micro-Analyzer
EVs	Electric Vehicles
FCC	Face Centered Cubic
GA	Galvannealed
GI	Galvanized
GI-IF	Galvanized Coated Interstitial Free
GND	Geometrically Necessary Dislocation
HAZ	Heat-Affected Zone
HI	Heat Input
HI_{Nominal}	Nominal Heat Input
HI_{Relative}	Relative Heat Input
HI_{Wire}	Wire Heat Input

¹ Only abbreviations mentioned more than once is listed.

I-IMC	Interface Intermetallic Compound
I.U.B.	Imperfection Upper Bound
IF	Interstitial Free
IMC	Intermetallic Compound
IPF	Inverse Pole Figure
KAM	Kernel Average Misorientation
LW	Laser Welding
LWB	Laser Weld Brazing
LoA	Lack of Adhesion
ML	Machine Learning
MT	Metallography Technique
NDT	Non-destructive Testing
OM	Optical Microscope
P	Laser Power
PF	Pole Figure
PMZ	Partially Melted Zone
ROI	Region Of Interest
SEM	Scanning Electron Microscope
SFE	Stacking Fault Energy
SI-IMC	Surrounded Interface Intermetallic Compound
T	Temperature
TEM	Transmission Electron Microscopic
TG	Thermal Gradient
TS	Travel Speed
T_b	Boiling Temperature
T_m	Melting Temperature
UT	Ultrasonic Testing
UTS	Ultimate Tensile Strength
WB	Weld Brazing

WFS	Wire Feeding Speed
YS	Yield Strength
μCT	Micro-Computed Tomography

Chapter 1. Introduction

1.1 Background of Study

Decreasing environmental impact while ensuring safety of passenger vehicles is critical for automotive companies in the last two decades [1,2]. The urgent need to mitigate global warming has driven a major shift toward sustainable transportation. Electric vehicles (EVs) have emerged as a key solution for reducing CO₂ emissions in the transportation sector by replacing conventional gasoline/diesel vehicles [3]. One of the key challenges in EV design is the need for lightweight vehicle body to improve energy efficiency [4]. Various generations of steels, from conventional to advanced high strength steels (AHSSs), have been developed with increased strength-to-formability ratio enabling reduction in sheet thickness in construction of body-in-white (BIW) structures while keeping strength intact to reduce the weight and maintain passenger safety simultaneously [5]. Moreover, the majority of automobile skin made from high formability-low strength interstitial free (IF) steels [6].

Selecting the right material for particular vehicle parts is important, but ensuring the crashworthiness of BIW structures and automobile skin is equally critical. The second step is the robustness of joint design in auto assemblies [2,7,8]. Due to the poor corrosion resistance of IF steels and AHSS in-service, they are typically coated with zinc (Zn), which offers a desirable degree of galvanic protection [2]. While Zn-coated steels offer enhanced corrosion resistance, the presence of Zn, and its associated intermetallic compound (IMC) (i.e. Fe-Zn IMCs that form during galvannealing process), on the substrate surface decreases the mechanical strength of the final joints [9]. The Zn coating in fusion welding embrittles the joint, and its burn-off further damages joint appearance and reduces mechanical performance in service. To mitigate Zn-related damage, it is crucial to minimize Zn vaporization, which can be achieved by reducing the processing temperature for joining. Weld brazing (WB) is a joining method with lower temperatures than fusion welding processes, which can mitigate Zn vaporization damage, as well as limit substrate melting [10,11]. In fact, WB is a gap-bridging welding method that does not require melting the base metal. WB is also often referred to as

brazing; however, this term can be misleading, as it does not reflect that a filler metal undergoes fusion while the base metal remains solid.

Among various joining processes, WB using laser heat source, laser weld brazing (LWB) has received significant attention in joint design in the realm of modern manufacturing [12]. According to Koltsov and Cretteur this method has been mostly employed for the joint assembly in luxury line of the well-known car companies in the world [13]. LWB has streamlined the assembly process in skin panel joining in automotive manufacturing, reducing the need for stitch welding, additional sealing, and molding that were traditionally required, while providing authentic joint appearance with minimal reworking before painting. Moreover, due to the precise energy input and the localized heat generation achieved by the laser heat source, LWB can elevate the production rate. This is achieved by allowing for increased travel speed, making LWB a more energy-efficient choice when compared to traditional brazing methods [14]. Producing high-quality LWB seams relies on optimizing the wetting process and interfacial reaction between the brazing filler and the substrates as well as minimizing the formation of brazing defects during the assembly process.

While surface defects are typically related to issues at the joint's outermost layer, sub-surface defects in LWB are often tied to the precise control of heat distribution and penetration within the joint, as well as material interactions and joint design [15]. Hence, the challenge lies not only in detecting sub-surface defects but also in understanding how their formation is influenced by process parameters, heat input distribution, and the underlying metallurgical mechanisms. While various in-line NDT methods exist to be used for detecting sub-surface defects in LWB seams, their sensitivity and range are limited, making it challenging to fully capture defects of different sizes and types. To overcome these limitations, enhancing ultrasonic testing (UT) for LWB is crucial because, unlike surface-focused methods, it can detect deeper sub-surface defects in real time, providing a cost-effective, radiation-free alternative to off-line methods such as X-ray micro-computed tomography (μ CT). Establishing a systematic optimization strategy to produce braze joints with customized geometry and sub-surface defects, along with their full-scale characterization, is crucial for adequately sourcing

artificial intelligence (AI) training data for UT enhancement. However, the current literature lacks an interactive approach to facilitate the production of a broad range of brazing conditions. Beyond its critical application in car body structures, this study investigates LWB as a method for integrating multi-material designs in automotive body construction to enhance structural and mechanical performance [16]. However, any multi-material joining process involving dissimilar metals, such as the use of a brazing filler in LWB, inherently introduces metallurgical complexities. These include the formation of IMC at the interfaces, which can significantly influence mechanical performance [17,18]. IMC formation is an unavoidable by-product of the brazing process. In WB of steels, Cu-based alloys containing Si, commonly known as Si-bronze, are widely used as filler material [19]. Fe-Si IMCs with various stoichiometries can form either within the braze/substrate interfacial layer or as dispersed particles within the joint [20–24]. While IMCs are often associated with embrittlement [25], their exact role in deformation and fracture behavior remains uncertain for those located within the interior region of the braze bead. As IMCs are stoichiometric phases have limited ability to accommodate lattice strains induced by deformation [26,27]. This rigidity makes them susceptible to crack initiation at IMC/braze-matrix interfaces, increasing the risk of premature failure compared to joints without IMCs. To date, studies on LWB have primarily focused on IMC formation at the braze/substrate interface and their presence as precipitates. Understanding the effect of IMC formation within the braze interior region, on mechanical performance, whether beneficial or detrimental, requires further investigation. The broader impact of these IMCs on fracture behavior, particularly their morphology-dependent influence on the surrounding matrix, mechanical response, strain redistribution, and crack propagation, remains a critical research gap. Given the increasing use of LWB in automotive structural applications [28,29], developing pathways to further explore more variation of IMCs and understanding their microstructural evolution and impact on mechanical response is essential for optimizing joint reliability and ensuring passenger safety.

1.2 Objectives

Improving the mechanical performance of joints made using LWB will improve overall vehicle crashworthiness and expand the application of LWB as a high-speed joining method with elevated joint appearance. The sub-surface defects forming within braze joints, which degrade joint integrity, have yet to be fully inspected in real-time during automotive assembly. Moreover, the role of IMC formation in joint mechanical performance remains unclear, whether it degrades or improves mechanical properties. This study aims to advance the LWB process by developing a novel optimization strategy for producing AI-friendly braze joints that enhance real-time defect inspection and by modulating IMC formation to maximize the mechanical performance of LWB joints which are critical automotive safety. The underlaying metallurgical mechanisms during LWB of Zn-coated steels using Si-bronze brazing material were systematically investigated to correlate the laser-wire-substrate interactions, that are controlled by process conditions, with sub-surface defect evolution, segregation behavior, and IMC formation. Additionally, this study deeply examines and clearly identifies the exact role of IMC morphologies and distributions in joint integrity and mechanical performance. This goal will be fulfilled by achieving the following objectives:

1. Characterize the evolution of substrate surface morphology and braze segregation under laser beam interaction and assess their influence on the wetting behavior of the brazing filler material.
2. Establish LWB optimization strategy accounting for filler wire deposition to accurately determine the variations in type and size of sub-surface defects and to modify IMC formation.
3. Develop a robust process window to facilitate automated detection and measurement of a wide range of sub-surface defects and braze geometry in an AI-enhanced UT inspection.
4. Modify braze microstructure using the developed optimization scheme to maximize mechanical performance and develop a fundamental understanding of the chemical

segregations and underlying IMC formation mechanisms govern the modified braze microstructure.

5. Characterize the effect of a wide range of IMCs on crack propagation, local fracture modes (i.e., ductile or brittle), deformation mechanism responsible for fracture behavior, and texture evolution dictating the joint fracture behavior and mechanical properties, to establish a scheme of tailoring IMC/braze-matrix interactions to maximize joint mechanical performance.

1.3 Organization of the Thesis

This thesis has been organized as follows:

- Chapter 1 presents the study background and the objectives.
- Chapter 2 presents a systematic review of Zn-coated automotive steels, their joining challenges, LWB process parameters, joint integrity, IMC formation mechanisms, and their impact on mechanical performance. It concludes by identifying key gaps in the literature and outlining pathways as accomplished through this thesis.
- Chapter 3 characterizes the effect of surface morphology and surface chemistry of Zn-coated steels at the bead edge properties during LWB of galvanized (GI)- and galvanized (GA)-coated DP600 steels in bead-on-plate (BoP) configuration. Additionally, it elucidates the role of a laser heat source in LWB, compared to an electric arc in gas metal arc brazing (GMAB), in influencing the wetting process and chemical segregation.
- Chapter 4 focuses on establishing a custom-designed process window for LWB to control sub-surface defects in braze seams manufactured in a representative roof joint configuration using galvanized coated interstitial free (GI-IF) steel for automotive applications. It presents a detailed LWB optimization strategy, which incorporates wire deposition rate into heat input calculation, offering a reliable metric for the thermal conditions in the joint. This refined metric enables clearer indication of heat input thresholds associated with the formation of various sub-surface defects across different

LWB parameter sets. This chapter evaluates the robustness of established process window by full scale characterization of the braze seams and assessing the performance of AI-assisted UT system trained using these joints.

- Chapter 5 focuses on modifying the braze microstructure in LWB of GI-IF steel in roof joint configuration using the novel heat input approach established in Chapter 4. This chapter provides new insights that broaden the understanding of IMC formation occurring within the interior region of the braze, and its manipulation to modify the braze microstructure. It comprehensively delineates the IMC formation and elemental segregation within the modified microstructure and evaluates the resulting joint mechanical properties.
- Chapter 6 provides an in-depth characterization and discussion to underscore the exact role of IMCs that are formed within the braze interior region on joint mechanical performance. It comprehensively characterizes the fracture microstructure and texture in braze joints containing a variety of distinct types of IMCs within the braze interior, formed using the novel heat input approach for Zn-coated Gen 3 Q&P980 steel substrates in lap joint configuration. This chapter elucidates the morphology-dependent interactions between IMCs and the braze main matrix affecting crack propagation and strain accommodation capacity of braze that ultimately dictate joint mechanical performance. It identifies the tailored IMC structures that maximize joint tensile peak load and displacement, enhancing joint's overall mechanical performance.
- Chapter 7 outlines the major conclusions and provides recommendations for future research.
- Chapter 8 provides the list of research contributions from the candidate.

The papers presented in Chapters 3 and 5 are printed in their final (accepted) form. The manuscripts presented in Chapter 4 and 6 are presented in its ready for submission versions and are subject to revision depending on the reviews received on the manuscript. However, the

figures, tables, equations, references numbers and abbreviations for all manuscripts have been re-formatted to provide a consistent style throughout this thesis.

Chapter 2. Literature Review

2.1 Zinc-coated Automotive Steels

2.1.1 Conventional to 3rd Generation Steels

Although nonferrous alloys, such as aluminum, are a popular choice when it comes to choosing lightweight materials for EVs, ferrous alloys still play a crucial role by offering excellent crashworthiness and strong protective structure for occupants during accidents [30,31]. In the pursuit of reconciling simultaneous weight reduction and vehicle safety, a wide variety of steel grades has been developed to provide the best combination of formability, strength, and weldability while reducing the strength-to-formability ratio over the past two decades [5]. Since 2007, as seen in Fig. 2.1a, the majority (55%) of automobile body structures have been constructed of mild or interstitial free (IF) steels, which exhibit the highest level of formability due to their inherently low strength. AHSS was used predominately for the remaining portion of car body while use of non-ferrous lightweight metallic alloys such as aluminum (Al) and magnesium (Mg) had marginal use (1%). Just a decade later, IF steel usage significantly dropped to 29%, being replaced by AHSS, while Al and Mg only increased to 2%.

The elongation vs. strength status of various steels' generations is classified in Fig. 2.1b, showing the improvement in the strength-to-formability (strength-to-elongation) ratio. This ratio increases from below 20% GPa in conventional low carbon mild steels or IF steels to above 20% GPa in 1st, 3rd, and 2nd generations of AHSS [32]. IF steels and mild steels are both ferritic with low alloying content. IF steels consist purely of ferrite with exceptionally low carbon (C) content (typically below 0.02 wt%) and only trace amounts of elements such as manganese (Mn), silicon (Si), and Al, primarily to aid in processing while preserving high formability. Mild steels also have a ferritic structure but contain slightly higher carbon levels (0.05–0.15 wt%), leading to a minor increase in tensile strength while exhibiting slightly lower elongation compared to IF steels (as seen from Fig. 2.1b). Both remain in the category of high-ductility, low-strength steels, making them suitable for applications requiring excellent

formability. In contrast, the microstructure of AHSS is more complex than IF and conventional steels due to the addition of alloying elements and microstructure modification through use of precise thermomechanical processing technologies [5]. AHSS is alloyed with elements such as C, Si, Al, and elevated amount of Mn content act as austenite stabilizers, inhibiting the presence of ferrite, [33]. 1st generation AHSS are with high strength with reduced ductility due to the formation of the martensitic phase while improves work hardening. 2nd generation steels were developed to enhance ductility greater than 1st generation, while maintaining strength by the adding large Mn additions, as high as 30 %wt. [34]. Nevertheless, the widespread implementation of 2nd generation AHSS was hindered by their comparatively high cost, mainly due to the expensive alloying elements involved and low production yield. The 3rd generation or Gen 3 AHSSs were developed to reduce production costs by reducing alloying contents compared to 2nd generation and to elevate strength-to-ductility ratio to levels greater than the 1st generation AHSS [35].

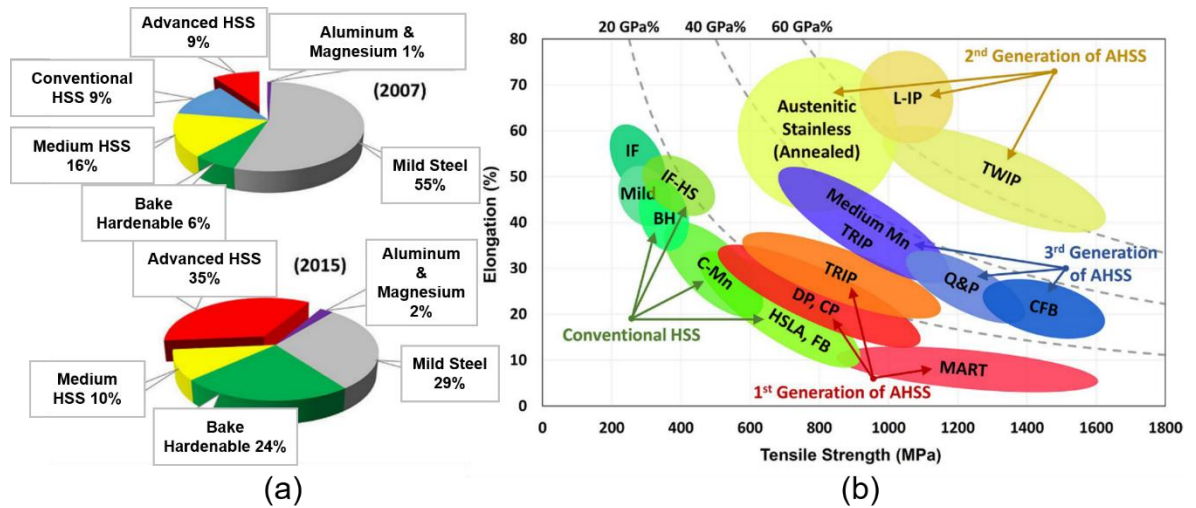


Fig. 2.1: (a) The adoption rate of different generations of steel besides other metallic alloys in automobile bodies, and (b) The most up-to-date Banana diagram shows elongation vs. tensile strength graph [32].

In this thesis, the joint mechanical performance of IF steel, and among AHSS, Gen 3 Q&P980 steel is investigated. The IF steel, widely used for automobile skin applications, is the conventional choice of material where high formability is required. In contrast, the Gen 3

Q&P980 steel features a complex microstructure provide superior mechanical properties at a reduced cost as intended for Gen 3 AHSS; however, it has yet to be widely adopted in the automotive industry.

The quenching and partitioning (Q&P) heat treatment was initially proposed by Speer et al. [36] in 2003. The process was designed to optimize the balance between strength and ductility in AHSS. Q&P steels undergo a four-step transformation from an austenitic phase to a multi-phase microstructure through precise thermal treatments, as shown in Fig. 2.2. The process begins with heating the austenitic steel above the A_{c3} temperature, followed by rapid cooling to the quenching temperature (QT) to form a controlled fraction of martensite. During the partitioning stage, the steel is held isothermally at the partitioning temperature (PT), allowing carbon to diffuse from martensite into austenite, stabilizing the remaining austenite. In the final stage, the stabilized and carbon-enriched austenite is retained at ambient temperature after the steel is fully quenched. This multi-step process is necessary because conventional high-strength steels typically suffer from a trade-off between strength and ductility—stronger steels tend to be brittle, while more ductile steels lack sufficient strength. The Q&P process is designed to overcome this limitation by producing a steel that combines the high strength of martensite with the ductility of retained austenite. The controlled diffusion of carbon ensures that enough austenite remains stable under deformation, preventing premature fracture and allowing the steel to absorb more energy before failure. This enhanced strength-ductility balance makes Q&P steels particularly attractive for automotive applications, where high crash resistance and formability are both critical requirements. Compared to the single-phase ferrite structure in IF steel, Q&P steel's multi-phase structure offers superior mechanical performance, justifying the complexity of its production process.

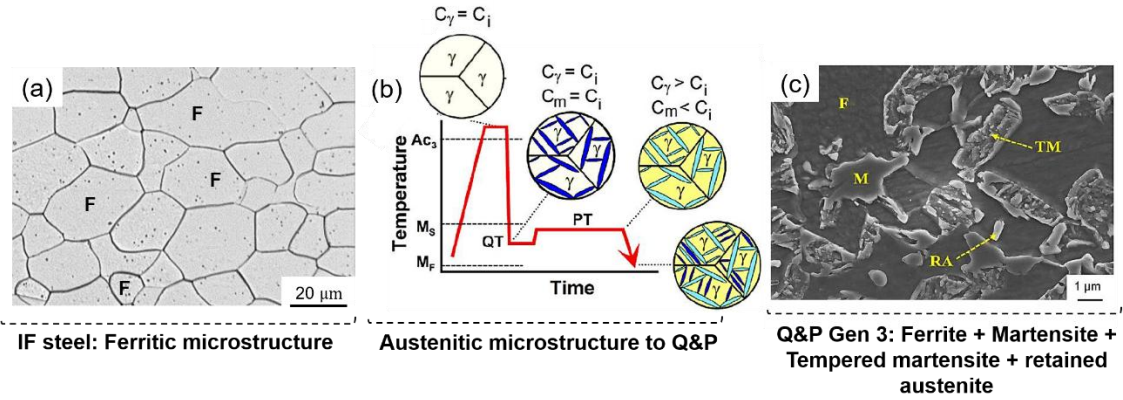


Fig. 2.2: (a) Optical micrograph (OM) of Ferritic microstructure of IF steels, (b) Q&P thermal treatment on pure austenitic steel, (c) resulting Q&P Gen 3 with complex microstructure revealed in scanning electron micrograph [36–38].

2.1.2 Zinc Coating

Like all steels, IF and AHSSs thin-gauge sheets exhibit poor corrosion resistance in-service. Therefore, they are typically protected using a Zn coating [5], which offers galvanic protection to the steel's surface. There are many types of Zn coating used, hot-dip GI and GA are the two most adopted coating types in the automotive industry [39]. GI consists of pure Zn with a very thin layer of Fe-Al intermetallic compounds located on the substrate surface. GA is produced by annealing the GI coating, allowing Fe from the substrate diffusing into the Zn coating to form a variety of Fe-Zn compounds at the surface which further enhances corrosion resistance [40]. GA exhibits higher corrosion resistance but greater surface roughness, whereas GI provides good paint-ability with a smoother surface finish due to lower surface roughness.

2.1.3 Challenges in Joining of Zn-coated Steels

While Zn-coated steels offer enhanced corrosion resistance, the presence of Zn and its associated IMC (i.e., Fe-Zn IMCs that form during galvannealing process), within the substrate surface degrades post-welding properties of joints [9]. This degradation in joint integrity occurs due to substantial Zn vaporization when subjected to conventional fusion welding processes. Zn vaporization results in welding defects; for instance, damage to Zn protection at the Zn depleted area of the joints degrading improved corrosion properties of Zn-coating as well as

lack of adhesion and porosity within the joint seam; these are the main concerns that can act as crack initiation sites, degrading joint mechanical performance [41]. In fusion welding processes, the temperature needs to be high enough to melt the substrate by exceeding the base material's melting temperature (T_m). The T_m of steel substrate is far above the boiling temperature (T_b) of Zn ($T_b(\text{Zn}) = 907^\circ\text{C}$ and $T_m(\text{Steel}) = 1425\text{-}1540^\circ\text{C}$), which leads to high Zn vaporization [42,43]. To mitigate Zn-related damage it is crucial to minimize Zn vaporization, which can be achieved by reducing the processing temperature.

When assembling the body panels with thin-gauge sheets, achieving an attractive joint is crucial, especially the joints connecting the roof panels to body sides and assembling upper to lower decklid panels, as they may be visible to end users [13,44]. These types of joints require a “seamless” appearance with minimal reworking required before painting [45]. Conventional joining methods, such as roof molding on the spot-welded joints, are becoming obsolete in the joint design of the rooftop and decklid in modern vehicles. When joining Zn-coated steels via fusion welding, achieving a unified joint design is challenging due to Zn vaporization, leading to poor joint appearance.

In addition to the challenges posed by Zn coating, requirement of base material melting in fusion welding leads to a significant alteration in the substrate's microstructure. This alteration can have a detrimental impact not only on the joint's appearance but also on its performance. To understand the source of the issue, fusion welding processes can be listed below, classified into two groups depending on the heat source used:

- Arc welding such as: Gas Metal Arc Welding (GMAW/MIG), Gas Tungsten Arc Welding (GTAW/TIG), and Shielded Metal Arc Welding (SMAW)
- Laser welding (LW)

In arc welding processes as depicted in Fig. 2.3a, heat is generated by an electric arc to melt the base metal and filler material, or electrode, to form a weld joint through steps including 1) establishment of an electrical circuit by connecting the workpiece and an electrode to a power source; Electrode can be chosen as consumable (melts into the weld) or non-consumable

(remains intact). 2) Electric arc formation: a high electrical potential difference (arc voltage) is applied when the electrode encounters base metal or with a small gap. The environment is ionized by the arc voltage to create an electric arc. 3) Heat transmission: the electric arc results in intense heat with temperatures of several thousand Celsius degrees, melting the base metal and filler/electrode material to form a molten pool. 4) Solidification: as heat is transferred from the molten pool releases (it cools down), solidification starts, creating bonding between the workpiece and the filler material [46,47].

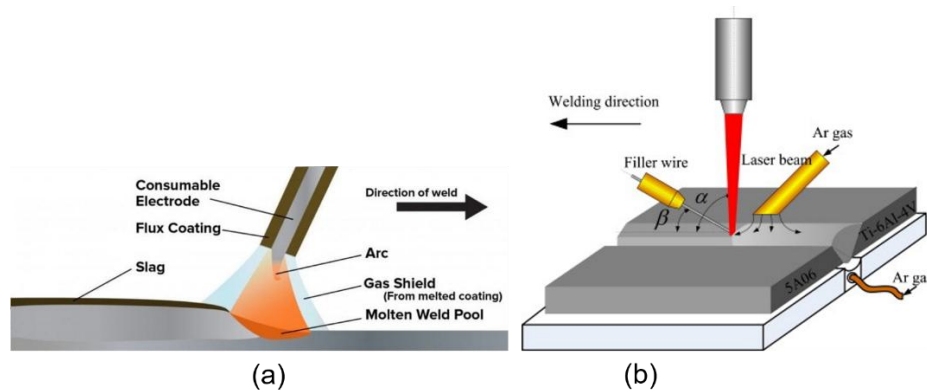


Fig. 2.3: Schematic diagram of comparing (a) arc vs. (b) laser welding [47,48].

In LW (Fig. 2.3b), a high-intensity laser beam is used to generate heat and melt the workpiece and/or the filler metal through the following steps: 1) Laser beam generation: a medium such as a crystal or gas is excited with energy. The emission of photons through this excitation results in a coherent and concentrated laser beam. 2) Focusing to define laser spot: optical elements, such as lenses or mirrors, are used to focus the laser beam to a small spot size. The focused beam allows precise control over the amount of heat applied to the workpiece. 3) Heat input transmission: by irradiation of the laser beam onto the workpiece, the material is rapidly heated. The high-energy photons transfer their energy to the atoms of the workpiece, causing these atoms to vibrate and collide, heating and melting are more localized than in arc welding. 4) Solidification: as the laser beam is removed, the solidification of the molten material occurs rapidly starts by cooling down to form a weld joint [48].

Comparing laser and arc heat sources in fusion welding methods, LW provides a more localized weld seam and narrower heat-affected zone (HAZ), compared to arc welding processes. However, due to the general nature of fusion welding, which requires melting of the substrate, the issues resulting from Zn burn-off causing degradation of corrosion resistance and joint mechanical properties still exist in LW. Hence, a joining process has been proposed without a need for fusion in the base material, significantly lower heat input to effectively minimize issues from substrate microstructural damages and Zn burn-off.

2.2 LWB Technology

A joining method to solve the problem of substrate melting by lowering heat input is WB, combining both welding and brazing elements. WB occurs by melting brazing filler wires with minimal melting or penetration of the molten pool into the substrate, creating joints between two or more similar/dissimilar substrates. The required heat input is limited to the input energy needed to melt the filler wire. The molten filler can wet the contact surface to create bonds between two or more solid metals; thereby substrate melting is minimized. This means that the processing temperature only needs to exceed the T_m of the filler metal, which is typically selected to be lower than the T_m of the substrates [10,11]. In fact, WB is a gap-bridging welding method that does not require melting the base metal. WB is also often referred to as brazing; however, this term can be misleading, as it does not reflect that a filler metal undergoes fusion while the base metal remains solid. WB in automotive applications is commonly done using either arc or laser heat sources. It is called LWB when using a laser heat source; LWB is an advanced joining process offering a higher joining speed by providing high energy density using a laser heat source, which is more consistent and controllable compared to an electric arc (i.e. GMAB) [49]. LWB offers high-speed joining that has played a transformative role in the automotive industry. It has significantly improved BIW assembly by enhancing processing efficiency and enabling aesthetically superior joints with minimal post-processing. For instance, in roof joint assembly, LWB has streamlined the process, reducing the need for multiple joining steps that were traditionally required, as shown in Fig. 2.4.

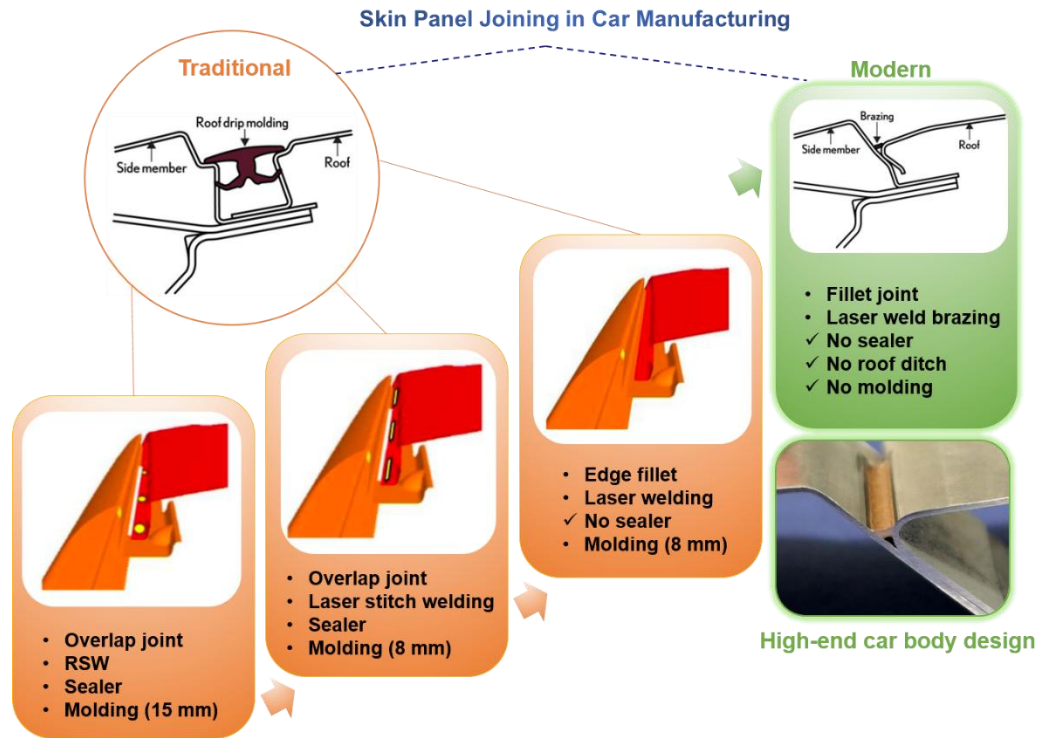


Fig. 2.4: Evolution of skin panel joining techniques for roof joints for automotive applications (reconstructed from Ref. [50]).

Heat generation in LWB is similar to the conduction mode in LW, as illustrated in Fig. 2.5. LW can be executed through two distinct heating modes, namely the keyhole and conduction modes [51,52]. In keyhole mode, the heat penetrates deeper into the material compared to conduction mode, resulting in the formation of a vapor-filled cavity within the material. In conduction the laser energy is used to heat the surface of material, causing localized melting and fusion without the formation of a vapor-filled keyhole, resulting in a shallower weld compared to the keyhole mode. In LWB, however, the heat generation is carefully controlled to ensure that the majority of the heat is absorbed by the filler, with only limited heat input transferred to the substrate. This is achieved by defocusing the laser beam to increase the spot size at the contact point, thereby reducing heat intensity and preventing excessive heat into the substrate. In LWB, when a defocused laser is used, the heat distribution is similar to LW in a conduction mode, assuming the effect of the wire on heat delivery to the substrate is ignored. With wire deposition into this laser/substrate interaction volume, instead of forming a FZ, the

filler wire melts and wets the substrate surface through capillary action, forming a metallurgical bond upon solidification,

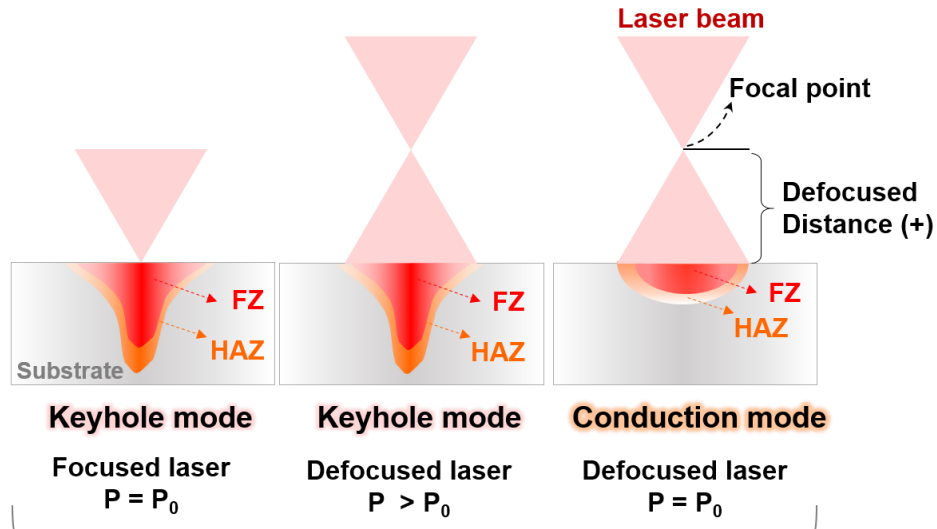


Fig. 2.5: Comparisons of different LW modes. LWB is at conduction mode.

LWB offers many options in positioning laser-beam/substrate/filler-wire at the interaction area, as long as the wire melts and the wetting process occurs without melting of the substrate. Conventionally, in wire-leading configuration the laser beam is positioned ahead of the wire (as seen in Fig. 2.6a) [53,54]. Conversely, if the laser beam interacts with the filler wire and the substrate from underneath the wire, it is called laser-leading configuration as can be seen in Fig. 2.6b [55]. The wire-leading configuration is generally preferred as it ensures better control over wire melting and wetting dynamics, reducing the risk of substrate overheating and excessive dilution. In Fig. 2.6c-d the laser/materials interaction is captured for a laser-leading LWB on BoP configuration. Fig. 2.6c shows laser impingement on the solid filler wire tip, with the braze metal not yet in contact with the substrate. Then, Fig. 2.6d shows the actual beginning of the bead formation process, where the filler wire is partially molten and not in full contact with the substrate. As shown in Fig. 2.6e, the solid filler wire fully molten making full contact with the substrate; this results in the melt pulling slightly to the front of the molten pool along the travel direction. In Fig. 2.6f, the wetting process, and consequently the LWB process will be stabilized [55].

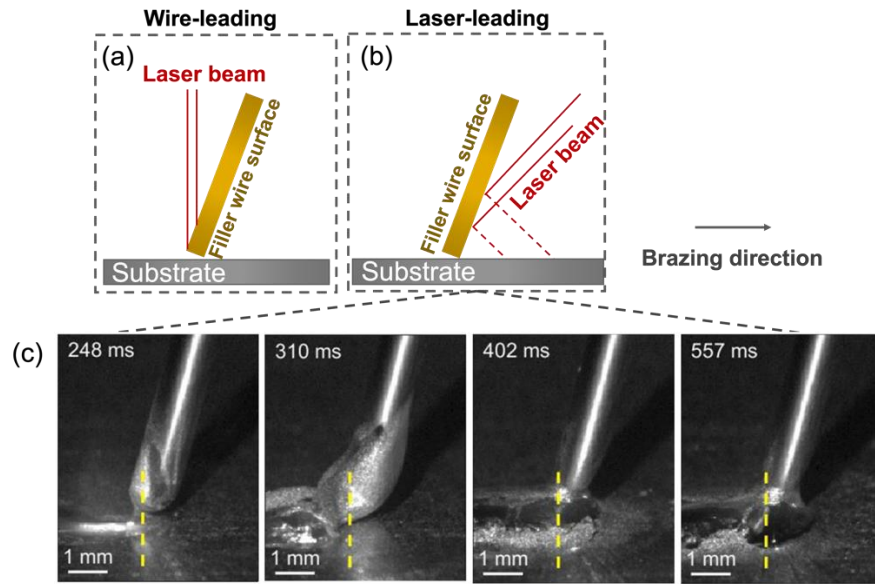


Fig. 2.6: Schematics of two LWB configurations with respect to laser/wire positioning: a) wire-leading, b) laser-leading, and (c) laser/filler-wire/substrate interaction illustrated by high speed imaging. The yellow dashed-line indicates the wetting front [55].

To provide a broad overview on the LWB process, several parameters are involved to offer a great degree of versatility in process adjustments to design intricate details in BIW structures with high precision and less thermal damage. Fig. 2.7 presents a comprehensive overview of LWB process parameters. Laser parameters are as follows:

- a) **Power:** The laser power refers to the total energy output delivered by the laser source per unit time. It determines the energy input to the material, influencing penetration depth, wetting behavior, and defect formation in LWB. Higher power levels enhance melting but may also increase the risk of excessive dilution and distortion. The laser power in operational condition is limited to the maximum power that can be provided by the laser heat source [56].
- b) **Beam profile:** The beam profile describes the spatial energy distribution of the laser beam, which affects the heat input and interaction with the filler wire and substrate. Common profiles include gaussian, top-hat, doughnut, Bessel, Multi-beam, etc. [51,57–62]. Among them, the gaussian beam profile is more frequently used in welding

and joining due to its diffraction-limited performance, efficient energy distribution, and compatibility with optical components [63].

- c) **Defocused distance:** The defocused distance refers to the distance between the focal point of the laser beam and the workpiece surface (working distance). A positive defocus (above the surface) results in a wider beam with lower intensity, while a negative defocus (below the surface) concentrates energy, increasing penetration. Proper defocus adjustment is essential for optimizing wetting and minimizing excessive dilution in LWB [56].
- d) **Travel speed and travel path (for wobble module):** Travel speed dictates the rate at which the laser moves along the joint, directly affecting heat input. Higher travel speeds reduce the heat input, limiting the molten pool size, while lower speeds increase energy absorption, promoting deeper penetration. In LWB, controlled travel paths, including wobbling techniques, help improve filler wire distribution, optimize wetting, and reduce porosity formation [64].

Regarding the brazing filler feeding system parameters, in LWB, the filler wire is fed into the joining area through a wire feeder, at a desired wire feeding speed (*WFS*), in close contact with the base metals to create the braze seam joint. As shown in Fig. 2.7b, the position of the wire and the laser beam spot is variable, with divergence/convergence to one another and the seam centreline. The filler wire parameters, including wire diameter, wire feed speed, Wire feed direction, and wire incidence angle and position with respect to the laser beam and the substrate, besides the convection coefficient of the metal surface, both wire and substrates, in the atmosphere have a significant effect on the degree of heat absorption by the metals at the laser/material interaction volume.

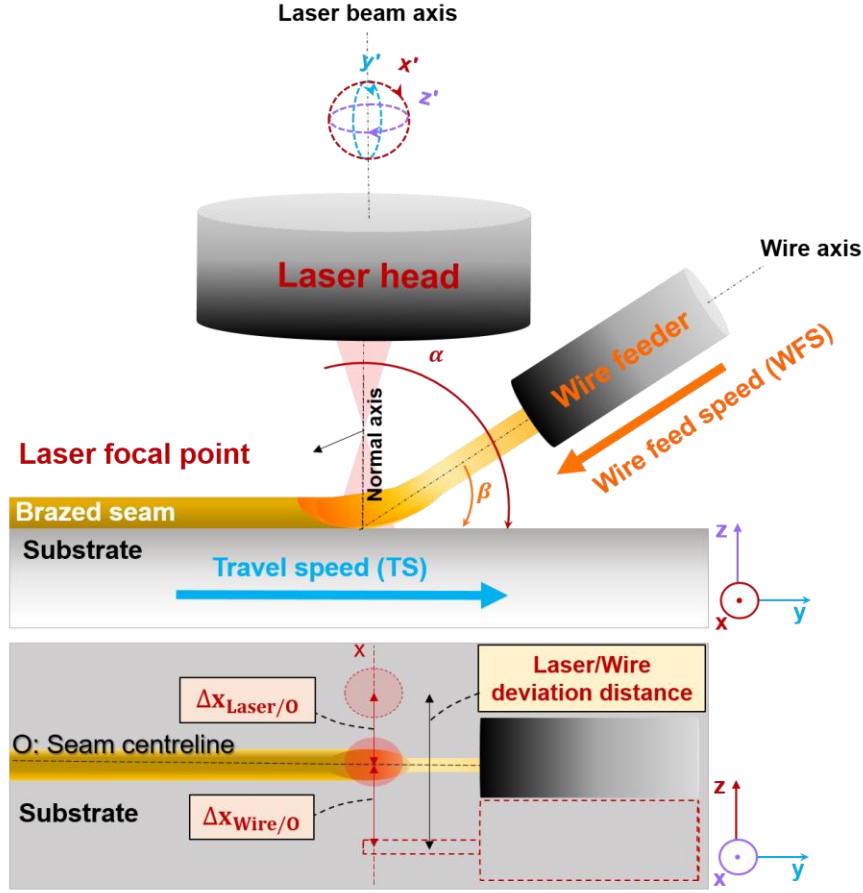


Fig. 2.7: LWB process parameter at a glance in (a) side-view and (b) top-view. The “xyz” coordinates are the principal axes with respect to the workpiece. The laser beam and wire alignments are defined with respect to the principal axes.

Parameter selection in LWB is strongly influenced by the heat input of the process. In many studies, nominal heat input ($HI_{Nominal}$) is commonly used as a simplified measure of energy input, as expressed in Equation 2.1 [65]:

$$HI_{Nominal} = \frac{P}{TS} \quad (2.1)$$

Where P represents the laser power, and TS is the travel speed of the laser beam along the seam or brazing speed. The unit of $HI_{Nominal}$ is energy per unit length of the joint. This equation offers a straightforward approach to process optimization, as it directly links power and travel speed, two easily adjustable parameters. However, Equation 2.1 does not account for key laser

parameters such as beam spot size, energy distribution, and absorption efficiency, leading to limitations in accurately predicting heat input. Accurately determining the actual heat input of the process requires more advanced modeling approaches [65–68].

Existing heat input models for laser heat sources typically assume full laser energy transfer, neglecting the thermal effects of filler wire melting, which is critical in LWB. For instance, Gaussian heat source models are commonly used to represent the laser intensity profile, yet most assume ideal energy transfer without considering absorptivity variations, beam defocus effects, or filler wire interactions. While 2D heat source models may be suitable for pure brazing, where heat remains mostly on the surface and penetration is minimal, they are often insufficient for LWB, where partial melting of the base material can occur. In such cases, a 3D volumetric heat source model is more appropriate, as it provides a more accurate representation of thermal gradients and depth effects [69].

The choice between two-dimensional (2D) and three-dimensional (3D) heat source models is particularly important for optimizing filler wetting behavior, as heat distribution directly influences capillary flow and interfacial bonding. However, implementing 3D Gaussian heat source models requires finite element (FE) simulations, making them computationally expensive. As a result, no direct heat input-based optimization strategies exist for LWB. Instead, process optimization is often based on iterative trial-and-error, relying on $HI_{Nominal}$ and empirical adjustments rather than a structured optimization framework that can predict brazing defects, intermetallic formation, or microstructure evolution in advance.

2.3 LWB Joint Integrity

LWB is an excellent joining method for the assembly of exterior BIW components and automobile skin, as it offers desired joint appearance with minimal reworking. However, producing high-quality LWB seams requires precise strategies to minimize formation of brazing defects during the assembly process. Surface defects result in poor joint appearance, these include: cavities, surface pores, blow holes, spatter, skipping, undercut, and wavy appearance seam [52]. In addition to surface defects degrading joint appearance, sub-surface

defects which cannot be seen from the surface, significantly limit joint integrity. Porosity, lack of adhesion, incomplete fusion, insufficient penetration, base metal melting, and cracking are common types of sub-surface defects, as schematically depicted in Fig. 2.8.

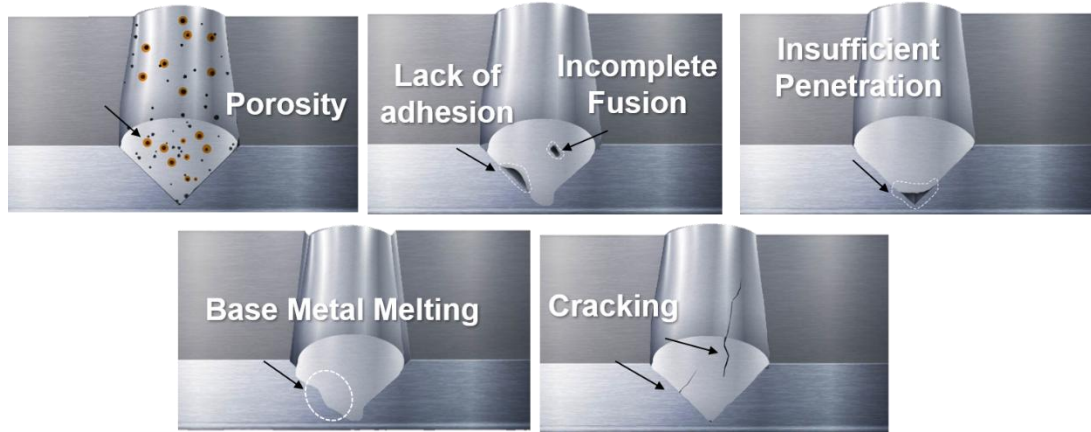


Fig. 2.8: Sub-surface defects classifications [70].

While surface defects are typically related to issues at the joint's outermost layer, sub-surface defects in laser brazing are often tied to the precise control of heat distribution and penetration within the joint, as well as material interactions and joint design [71]. Hence, the challenge lies not only in detecting sub-surface defects but also in comprehending their complex origins. To inspect sub-surface defects in LWB seams at a manufacturing scale, various in-line NDT methods are employed, such as eddy current testing (ECT) and UT. Additionally, X-ray and micro-computed tomography (μ CT) scanning are used for off-line evaluation and detailed defect characterization [49,72–74]. CT scanning is a widely used NDT method that enables the detection and 3D visualization of subsurface defects down to micro and nano scales depending on the voxel size. Fig 2.9 presents examples of CT scanning results for detecting voids, cracks, and porosity, as well as identifying part geometry.

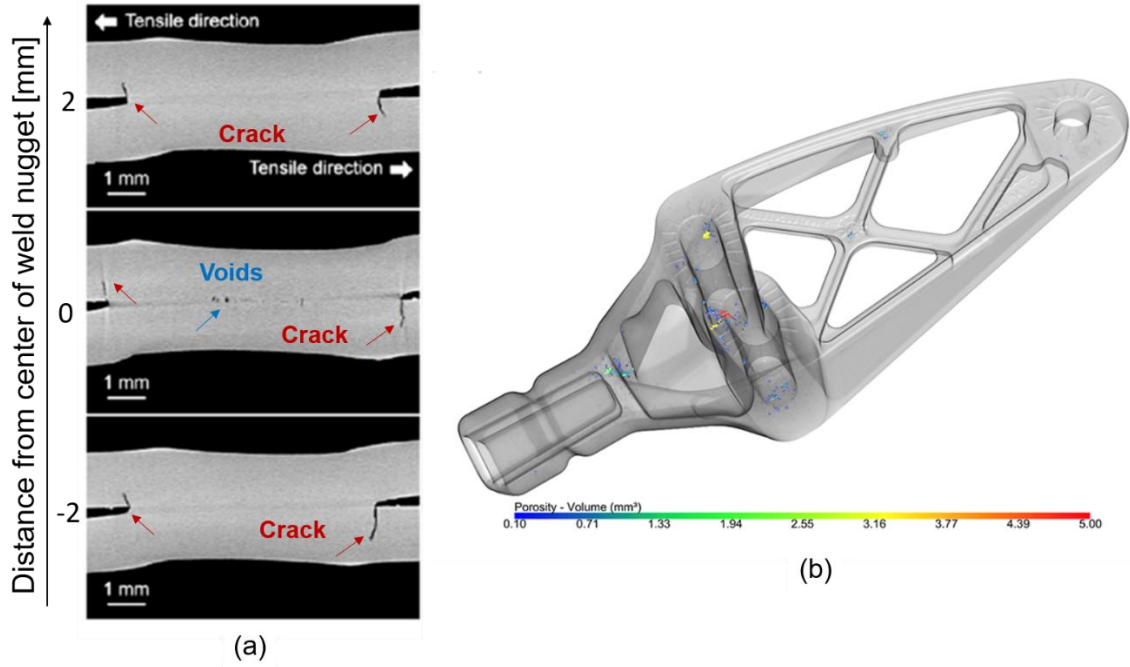


Fig. 2.9: Sub-surface defect and inner geometry detection using CT scanning. (a) 2D slices from RSW sample after tensile testing. (b) 3D visualization of porosity formed within a casting part with color coding by volume [75,76].

While CT scanning provides precise detection of sub-surface defects, it is expensive and limited to offline analysis, making it unsuitable for real-time quality control. Additionally, its long processing time and the need for specialized expertise further hinder its integration into high-speed manufacturing environments. Over the last decade, NDT technologies have undergone remarkable advancements through the adoption of AI to improve real-time interpretation of in-line detection methods such as UT [77–81]. AI employs algorithms designed to replicate or augment human expertise in interpreting NDT data. The convergence of AI and NDT has ushered in a new era by edging toward the goal of zero-defect manufacturing by enabling real-time monitoring and predictive maintenance strategies to diagnose the root causes of a detected defect and adjust the variables to resolve the issue instantly without shutting down the process line, as seen in a flowchart of Fig. 2.10 [82]. This can reduce human error, improve inspection reliability, achieve cost savings, and enhance efficiency. Defect detection using machine learning (ML) and deep learning (DL) often

involves framing a signal/image processing stage as anomaly detection [83], classification (identification of a category for a given input) [84], object detection (classifying regions of interest and tightly enclosing them in bounding boxes) [85], or semantic segmentation (classification of pixels/voxels in inputs) [86]. Though these strategies have proven powerful in NDT applications, including in ultrasonic, the development and deployment of modern AI approaches has many significant challenges, especially in NDT of parts being made where safety is critical such as automotive assembly.

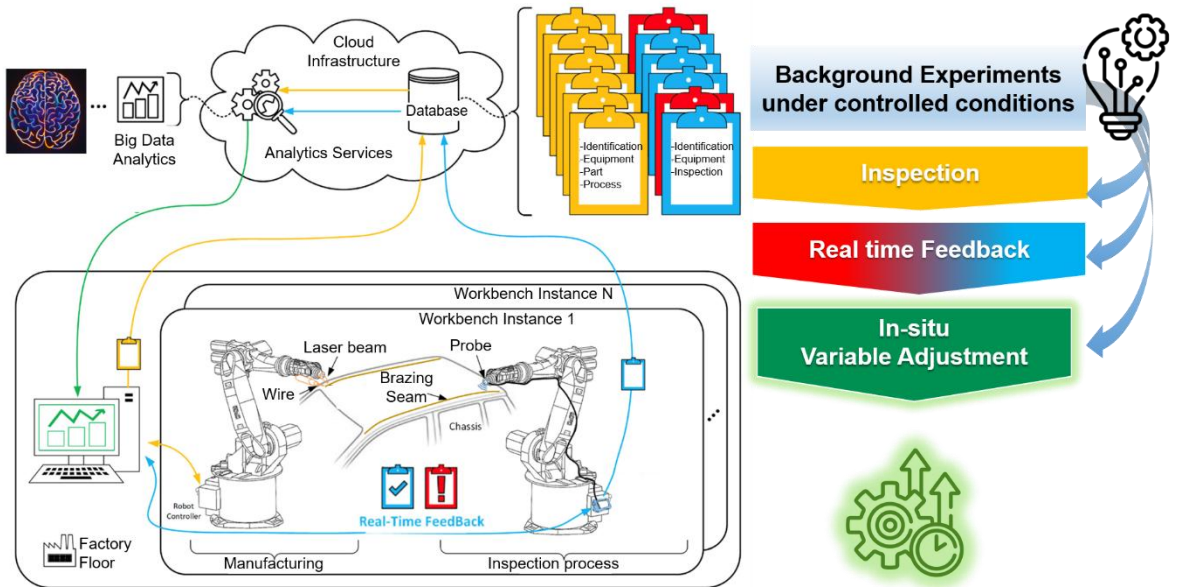


Fig. 2.10: Real-time inspection and process correction using AI augmentation in the factory line (inspired from [74]).

For LWB joints, however, UT is not the primary NDT method due to several challenges, including wave attenuation in Zn-coated steels, the complex microstructure caused by IMCs, and seam geometry. Additionally, the industry primarily relies on optical and X-ray inspection. Ongoing research aims to address these limitations and enhance the feasibility of UT for laser brazed joints. Enhancing UT applications for LWB is crucial because, unlike surface-focused methods, it can detect deeper sub-surface defects in real time, providing a cost-effective, radiation-free alternative to X-ray and CT. Advances in phased array UT, AI-driven signal

processing, and automated inspection systems are increasingly making UT a viable solution for real-time defect monitoring in LWB production, aligning with Industry 4.0 objectives. To achieve this for LWB demands high-quality datasets containing well-characterized sub-surface defects, which remain a major challenge. These defects must be systematically identified across various brazing conditions to enhance model accuracy and deepen understanding of their formation mechanisms. However, no established guidelines exist for optimizing defect formation during LWB to facilitate AI training. The effectiveness of DL algorithms in detecting and visualizing defects and joint geometry depends on diverse and extensive datasets, yet significant gaps remain in AI databases for assessing LWB joint integrity [87,88].

2.4 Intermetallic Compound (IMC) Evolution in LWB

IMC phases are the products of interfacial reactions at the interface between the molten filler and the substrate(s) during wetting process in LWB. The bonding of two dissimilar metals at their interface is closely associated with the formation of an IMC layer [89]. The thermodynamic stability of IMC phases determines which compound would form within the interface based on the phase diagram corresponding to the existing elements [90]. However, these IMC phases are brittle and known to act as crack initiation sites, posing a considerable challenge to brazing applications [91]. Hence, understanding IMC formation and its characteristics is important, as it influences joint reliability and life span. In this section, IMC formation and factors affecting IMCs evolution, will be discussed with a focus on the joining of Zn-coated steel substrates using Si-bronze filler wire, which is the aim of the present thesis. The Si-bronze is a brazing alloy composed of Cu with an addition of Si (~ 3 wt.%) and other elements such as Mn (< 3 wt.%) commonly used for WB of steels. The presence of Si reduces the filler's T_m , allowing a reduction in required heat input to melt the filler wire. Furthermore, Si enhances the flowability of the brazing alloy, improving the wetting behavior of Cu. Mn is often included to further enhance deoxidation, strength, and fluidity, contributing to improved joint quality [92]. The literature on LWB for Zn-coated-steel/Si-bronze joints concerning IMC is only limited to a few studies, thereby, IMC formation during GMAB has also been discussed throughout the following sections.

2.4.1 IMC Formation Phenomenon

IMC formation follows certain thermodynamic principles. Only compounds with a decreased solute diffusion coefficient (D) and lower Gibbs free energy (ΔG) will be stable as an IMCs in the joint [93–95]. The ΔG involves the enthalpy change (ΔH) and entropy change (ΔS) of the reaction between braze constituents in a construction of the Equation 2.2 as shown below:

$$\Delta G = \Delta H - T\Delta S \quad (2.2)$$

Where T is the temperature of the reaction, which is the molten material temperature during LWB. Stable IMCs have set stoichiometric ratios and a well-defined crystal structure, exhibiting explicit physical and mechanical properties distinct from the parent metals.

To understand the basics of IMC formation in WB of a steel substrate using Si-bronze brazing alloy, it is essential to analyze a basic system of a liquid Cu-Si alloy coming to contact with solid iron (Fe), the substantial element of the steel [96]. The role of Zn-coating from GI steel was not considered in the following example but will be discussed later in section 2.4.2. Fig. 2.11 illustrates the sequence of the IMC formation during GMAB process. Fig. 2.12 presents the Fe-Si binary phase diagram alongside ΔG plots for Fe-Si IMCs. The phase diagram shows phase stability across temperature and composition ranges (Fig. 2.12a). The ΔG formation plot indicates the thermodynamic stability of different Fe-Si IMCs (Fig. 2.12b). In Fig. 2.12c, the feasibility of phase transformations among these IMCs is given. The mechanisms underlying IMC formation were first detailed by Yu et al. [96] as follows:

- 1) Initially, diffusion of Fe atoms into the Cu side occurs under the “anode spots” action of the arc as soon as the wire filler melts (Fig. 2.11a). At the same time, Cu and Si atoms diffuse into the interfacial zone, the interface between the steel substrate (Fe) and the filler metal (Cu-Si).

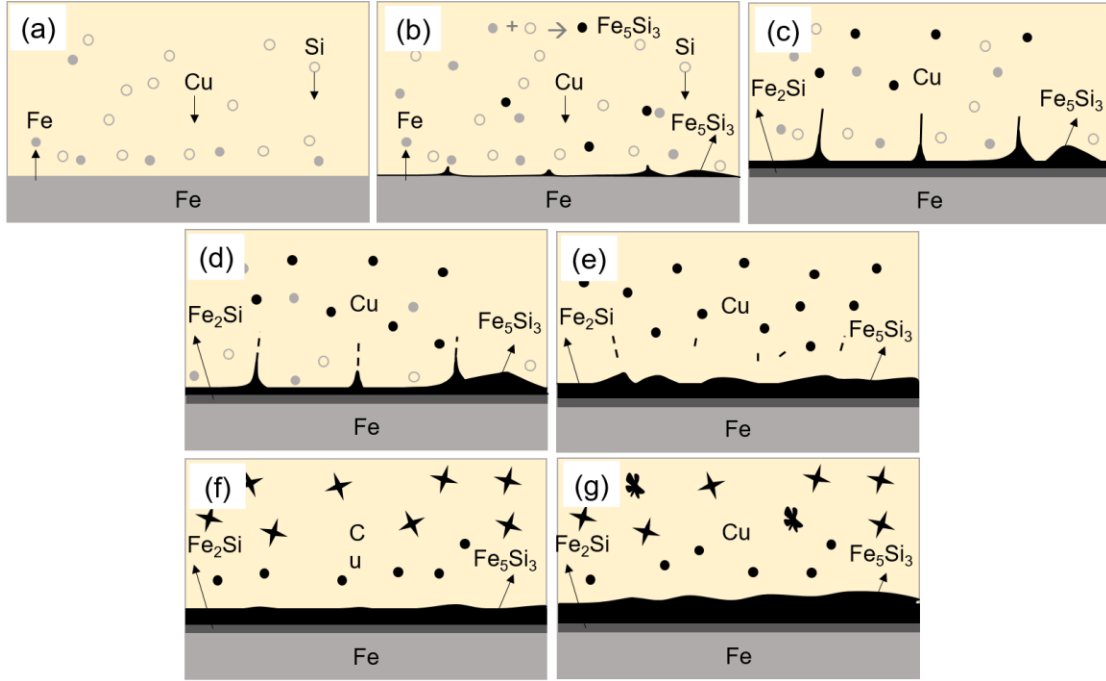


Fig. 2.11: Schematic diagram of IMC formation during solidification of liquid Cu-Si when contacting the solid Fe (Reconstructed from Ref. [96]).

- 2) As the arcing time increases, Fe atoms react with Si atoms at the interface to form a $\text{Fe}_5\text{Si}_3(\text{Cu})$ layer on the Cu side of the interface with two different morphologies, wave-like, and sprouts (Fig. 2.11b). $\text{Fe}_5\text{Si}_3(\text{Cu})$ is the first IMC formed in this system due to its high affinity to survive among several Fe-Si compounds produced at high temperatures including FeSi , FeSi_2 , Fe_2Si_3 , and Fe_5Si_3 . According to Fig. 2.12 the formation ΔG of Fe_5Si_3 is lower than other compounds, so at the primary stage, Fe and Si have more tendency to form Fe_5Si_3 .
- 3) The wavy-form Fe_5Si_3 grows deeper with a whisker-like shape into the braze. At the same time, Fe and Si atoms at the Fe side react with the developed Fe_5Si_3 layer reactions of i), ii), iii), and iv) (Fig. 2.11c and 2.12c). It was found that the D of Si in FeSi and FeSi_2 are relatively low and ΔG of the corresponding reactions of i) and ii) is high, so it is difficult for them to remain and not being dissolved. Thus, Fe_2Si_3 nucleates at this stage while Fe_5Si_3 contributes to the growth mechanism (Fig. 2.11c).

- 4) As the arcing time increases, both the Fe_2Si_3 and Fe_5Si_3 IMC layers fragment into short-stick-like, which are pieces pushed into the braze. This is called the IMC fragmentation stage. This is because of the arc stirring force and the sweeping force of the liquid flow at the interfacial zone.
- 5) Solidification of fragmented IMCs in two ways in the cooling stage: either supercooling or sweeping into the cooler zones by imposed stirring.
- 6) Solid-state growth of fragmented IMCs to shape in star-like and flower-like morphologies, which are indicators of high and low cooling rates, respectively.

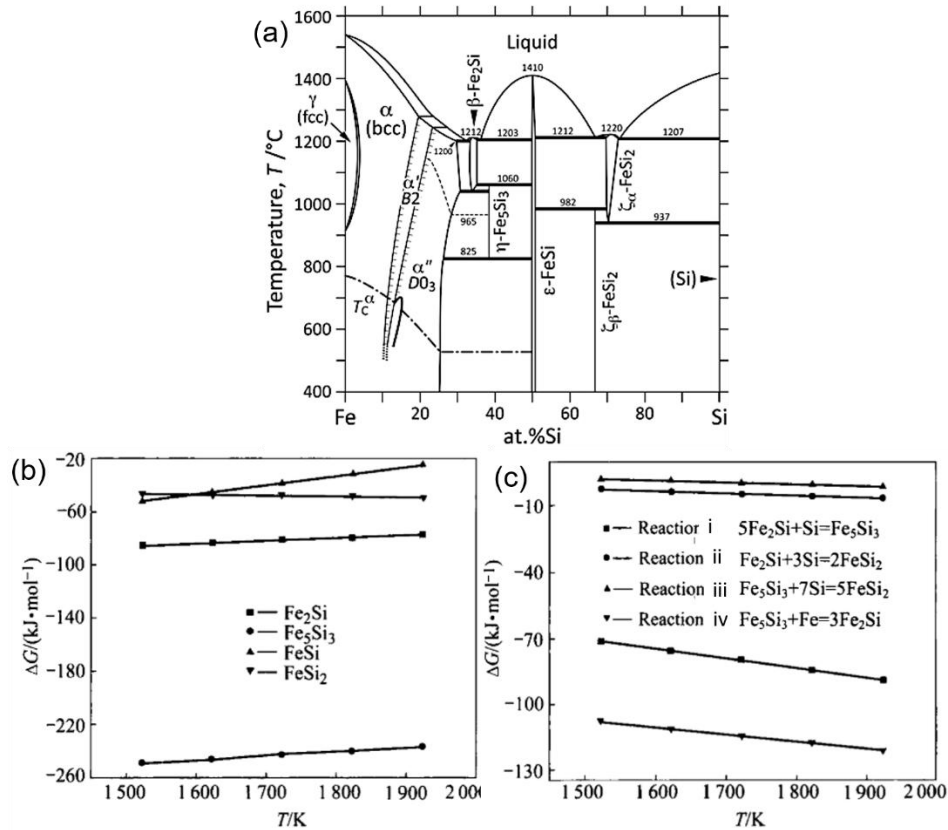


Fig. 2.12: a) Phase diagram of Fe-Si system, b) The formation ΔG of Fe-Si IMCs, and c) The reaction ΔG between Fe-Si IMCs in GMAB (Reconstructed from [96,97]).

The formation mechanisms of Fe-Si based IMC layer with flat and dendritic morphologies in WB process are well documented in the literature and align with stages 1) to 3) of the IMC formation sequence reported by Yu et al. [96]. As seen from the Fe-Si phase diagram (Fig.

2.12a), various compounds form sequentially with increasing silicon content. The IMCs classified so far include η -Fe₃Si₃ (~33 at.% Si), ϵ -FeSi (~50 at.% Si), $\zeta\beta$ -FeSi₂ (~66 at.% Si), and $\zeta\alpha$ -FeSi₂ (~66 at.% Si), with the latter being stable at lower temperatures [24,97]. However, the IMC fragmentation mechanism responsible for the formation of fine granular Fe-Si IMC (stages 4) to 6) of IMC formation sequence reported in [96]) contradicts with the mechanisms reported in more recent studies.

Khan et al. [98], investigating a similar brazing process, GMAB, reported that the granular IMCs are remnants of IMC nuclei within the braze, formed when heating time is too short to allow for further growth. Later, Sarmast-Ghahfarokhi et al. [99] identified the same Fe-Si-rich granular phases, including β -FeSi₂ and α -FeSi₂, classifying them as precipitates rather than IMCs, resulting from the dilution of the steel substrate into the molten Cu filler. Since Fe and Si have limited solubility in the Cu matrix, their accumulation beyond the solubility threshold leads to precipitation. However, such Fe-Si compounds are predicted to form as IMCs based on the Fe-Si binary phase diagram, as reported by Hao et al. [97].

In a recent study by Zhao et al. [100], the sequence of IMC formation during LWB was explored using thermodynamic models. They found that the sequence depends on interactions among Fe, Si, and Cu, with additional influences from Mn and Zn where applicable, resulting in the formation of Fe-Si-based and Cu-based IMCs, as summarized in Fig. 2.13. Mn and Zn influence phase stability by affecting diffusion behavior and reaction kinetics. Zn preferentially diffuses into the Cu-based bead, while Mn segregates at grain boundaries and contributes to the formation of Mn-Si compounds.

Regarding IMC layer formation at the interface, Si initiates the interfacial reaction by diffusing from the molten braze into the steel substrate due to the chemical potential gradient between the Cu-based filler and the Fe-rich base material. Thermodynamic calculations indicate that Fe has a strong affinity for Si, leading to the stabilization of Fe-Si reaction layers. As a result, the first phase to form is typically α -Fe(Si), which solidifies around 1410°C, as predicted by equilibrium calculations based on the Scheil–Gulliver solidification pathway [100].

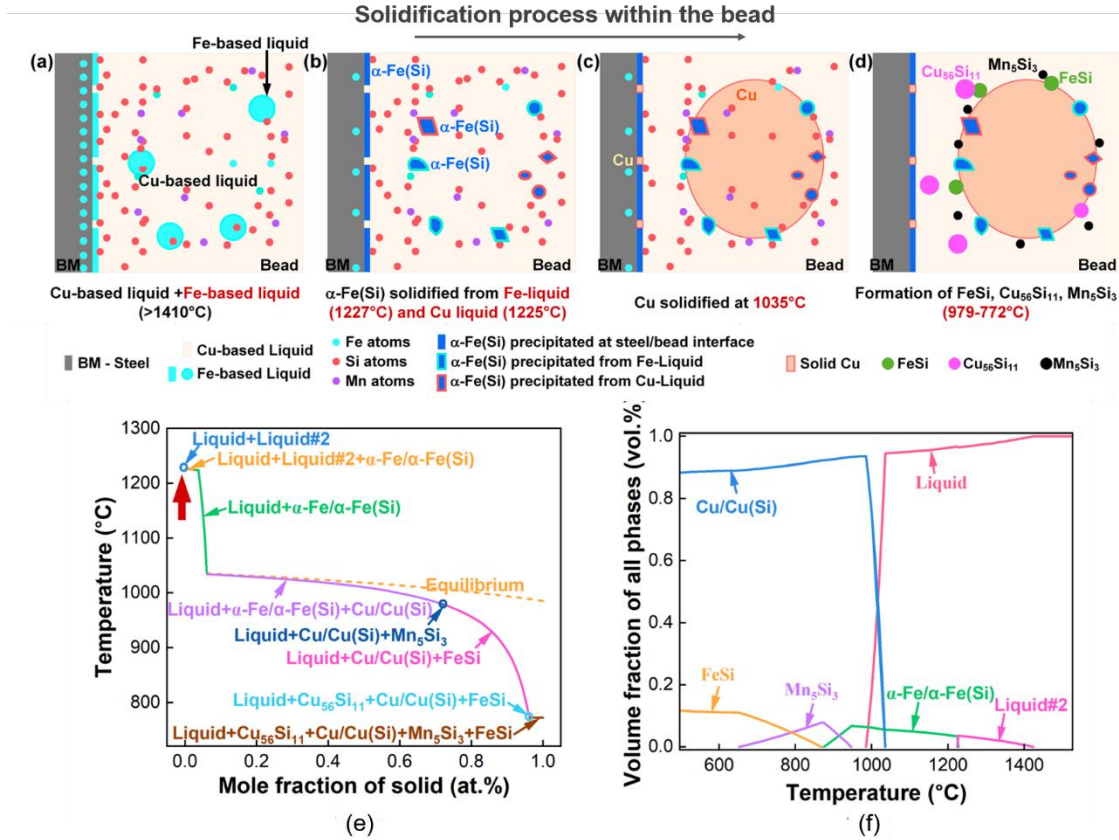


Fig. 2.13: (a-d) A schematic for the joining mechanism in a LWB condition favorable for high Fe diffusion into the braze. ThermoCalc curves indicating the phase formation in the bead: (e) Scheil–Gulliver solidification pathway, and (f) one-axis calculation [100].

For IMCs within the braze bead, Fe dissolves into the Cu-based liquid, influencing the solidification sequence. A high Fe concentration in the molten braze pool leads to the formation of a secondary Fe-rich liquid phase (Liquid #2), altering phase stability and promoting the early precipitation of α -Fe/a-Fe(Si) around 1227°C. As cooling progresses, Cu-based phases (Cu-Si, $\text{Cu}_{56}\text{Si}_{11}$) solidify around 1035°C, followed by late-stage Fe-Si (FeSi) and Mn-Si (Mn_5Si_3) IMCs at lower temperatures (~772–979°C). The presence of Mn in the braze bead supports Mn-Si compound formation at these later stages, while Zn remains primarily in the Cu-rich phase, influencing segregation and diffusion kinetics. The non-equilibrium Scheil-Gulliver calculations confirm that Fe-rich phases nucleate first, followed by Cu- and Mn-Si-based compounds as the system cools [100].

Recent studies, conducted in parallel with this research, have significantly contributed to understanding IMC formation, highlighting the complex interplay of thermodynamic and kinetic factors. These concurrent efforts underscore the dynamic nature of this research field, where multiple perspectives continue to refine the knowledge of IMC evolution. While substantial progress has been made in elucidating phase stability and formation sequences, aspects such as the potential for IMC morphologies beyond granular structures to develop within the braze area, the effects of increased Fe diffusion into the braze, the texture of IMCs, and their influence on the overall braze texture remain areas for further exploration.

2.4.2 Effect of Zn-coating on Wetting Behavior and IMC Formation

The Zn-coating on steel sheets, used for corrosion protection, can influence the wetting process and interfacial reactions contributing to the IMC formation during WB. Khan et al. [101] reported that a tail-like region containing a high amount of Zn can form at the bead edge of GI-coated steel during GMAB in a BoP configuration, while it was not observed in the case of GA-coated steel (Fig. 2.14). The Zn accumulation at the bead edge has a direct correlation with improvements in wettability, observed as a decrease in contact angle at the wetting profile; taken as a sign of good wettability. The tailing-out that occurred during the GMAB of GI-coated sheet, can be due to the surface morphology (i.e. surface roughness) and surface chemistry (i.e., available Zn content). The GI-coating, with its lower surface roughness than GA-coating, reduces heat absorption at the surface, leading to slower Zn evaporation. The lower Zn evaporation in GI leads to its accumulation at the periphery of interaction zone, forming a tail-like region at the bead edge reducing the wetting angle; this improves the braze appearance by smoothening the seam edge compared to GA with higher wetting angle and no tailing-out effect [101].

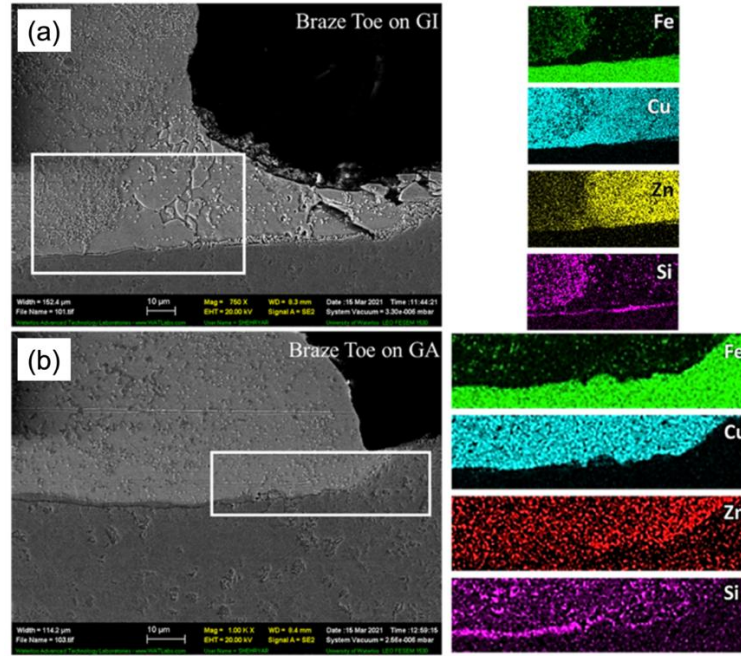


Fig. 2.14: Elemental maps showing Zn accumulation at the bead edge in GMAB of a) GI-coated steel and b) GA-coated steel [101].

In Khan et al.'s [102] study on lap-joint GMAB, it was found that tail formation occurs in both GA- and GI-coated steels, but with different composition. GA-coated steel formed a less brittle Cu-Zn alloy ($\sim$ Cu-25 wt% Zn) at the tail region compared to the Zn-rich tail of GI ($\sim$ Cu-50 wt% Zn). This difference was attributed to the interaction between the Zn coating and the molten Si-bronze filler, where Zn dissolves into the Cu matrix, and its concentration at the tail of the root varies depending on the coating type. In GI-coated samples, the higher Zn content ($\sim$ 50 wt%) resulted in a brittle IMC with pre-existing cracks, whereas in GA-coated samples, the lower Zn content ($\sim$ 25 wt%) led to a more ductile Cu-Zn alloy.

Zn contribution to wetting and interfacial reactions was also reported by Cao et al. [9] during GMAB of GI-coated steel with Al- and Mg-based filler materials. It was found that the effect of Zn on the wetting and bonding mechanisms differs between GI/Al joints and GI/Mg joints. In GI/Mg joints, the Zn-rich tail is formed, owing to the accumulation of Zn driven from the center during the Zn evaporation. While, in GI/Al joints, the diffusion of liquid Zn from the outer region of Zn-enriched periphery into the molten Al is responsible for Zn-rich tail

formation. In fact, in GI/Mg joint, Zn coating reacted to form the IMC layer between Mg and GI, while in GI/Al, Zn coating was non-reactive to IMC formation between Al and GI. Thus, the effect of Zn on the wetting process and interfacial reactions at the bead edge can vary case by case, and should be considered as an important factor influencing braze quality and IMC formation, equally significant as the interfacial reactions occurring at the center of the joint. In fact, the Zn-rich tail itself may be a type of IMC containing Cu, Fe, and Si, in GI/Si-bronze joint which has not yet been directly mentioned this region has IMC nature. The literature on elucidating the formation of Zn-rich tail IMC type is so limited and not being investigated in-depth for LWB.

2.4.3 Effect of Heat Input on IMC Formation

In LWB of GI steels using Si-bronze filler metal, it has been frequently observed that Si segregation is the main reason for forming IMCs [103]. The formation of IMC can occur when transferred energy and interaction time are sufficient for Si atoms to diffuse from the molten filler and Fe atoms to dissolve from the substrate surface (which should be slightly melted) to the interface to create stable Fe-Si phases. Li et al [103] investigated how changes in power input, which directly affect the $HI_{Nominal}$, influence Si segregation at the interface and promote IMC formation [106]. Fig. 2.15 shows the elemental distribution across the braze/substrate interface of a function of power. Note that low power (850 W) a narrow plane or flat Fe-Si IMC layer formed by Si segregation and Fe diffusion across the interface. At high power, 2 kW, IMC growth occurs, transforming the morphology from a flat layer to a dendritic structure as well as promoting formation of granular particles. As can be seen in Fig. 2.15b the IMC layer became thicker with dendritic morphology along with dispersion of some granular IMCs adjacent to the IMC layer. Due to the higher processing temperature resulting from using 2 kW power, more Fe atoms entered the braze to form Fe-Si IMCs. However, in a short heating time, compounds forming further into the braze could not grow up and the IMC shape remained granular [106]. Notably, Zn persisted at the interface of both conditions, exhibiting that Zn may remain even within the braze/substrate interface at low heat input.

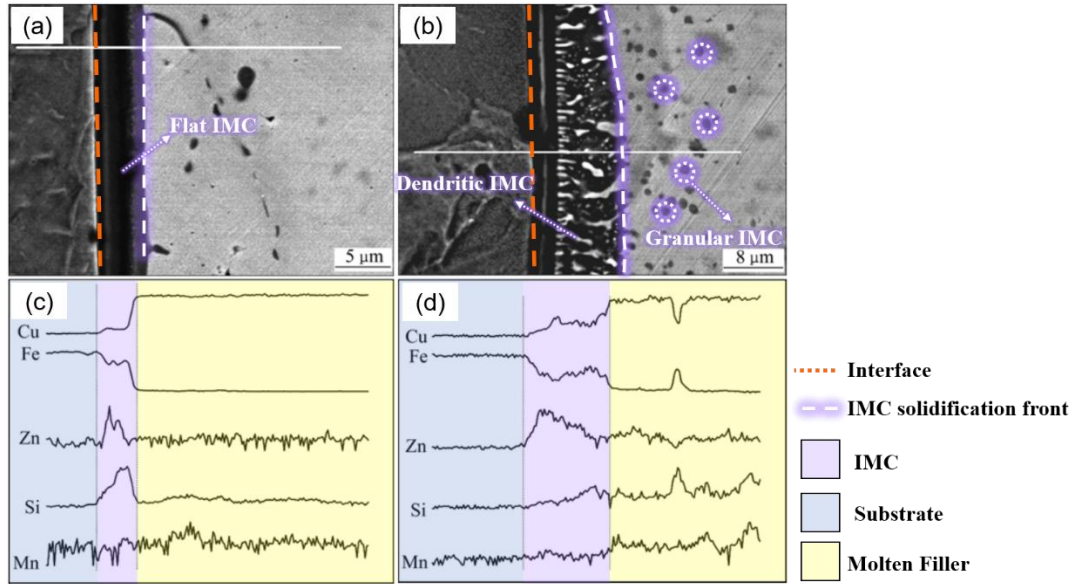


Fig. 2.15: Different IMC morphologies of LWB joint at the power of a) 850 W and b) 2 KW with corresponding elemental line profile are depicted in c and d [103].

The $HI_{Nominal}$ describes the thermal energy delivered during the brazing process but does not account for the influence of filler wire deposition. There is only one contribution in the literature that consider the role of WFS into the IMC creation. In a study by Zhao et al. [100], it was found that variations in WFS at a constant P/TS, $HI_{Nominal}$, altered the braze microstructure and IMC evolution. Although their work did not explicitly analyze the effect of heat input, the results indicate that WFS affected thermal cycle, playing a crucial role in controlling interfacial reaction and phase formation. The interfacial reaction layer primarily consisted of α -Fe(Si) solid solution, forming a thin and uniform layer with total thickness of 3.28–3.40 μm (Fig. 2.16). Despite variations in WFS, the thickness of the interfacial layer remained nearly constant due to rapid cooling and thermodynamic constraints on Fe-Si-Cu interactions.

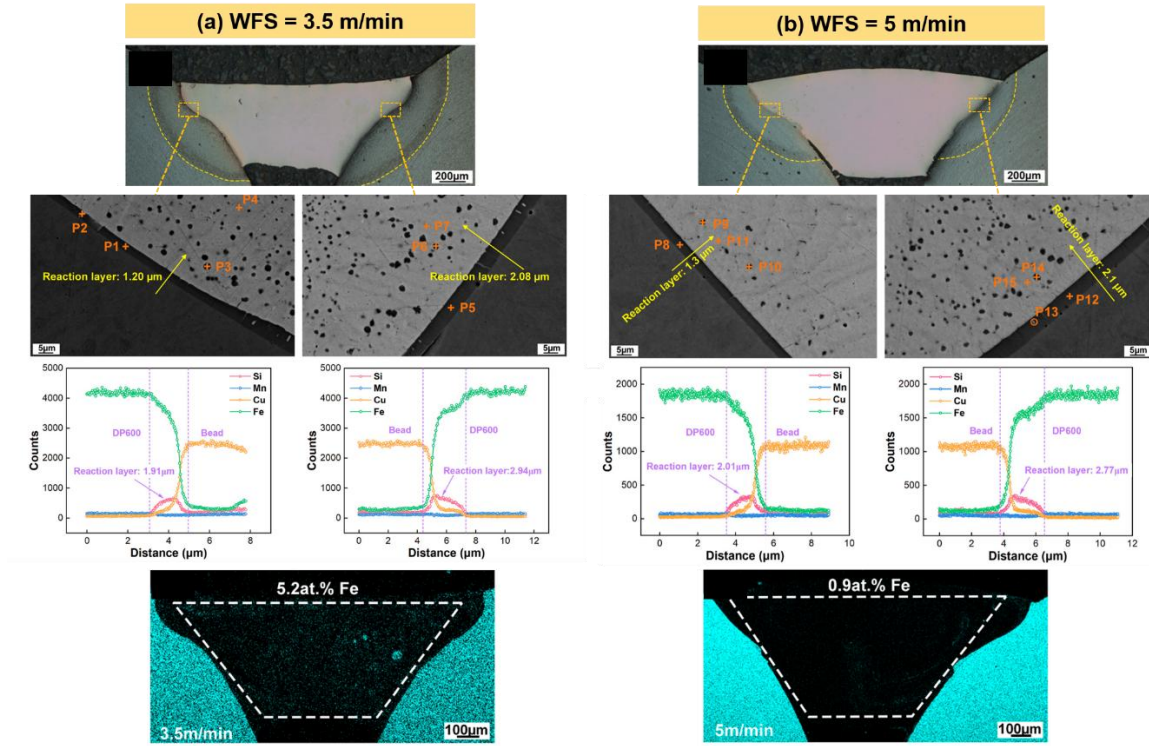


Fig. 2.16: Interfacial reaction and Fe dilution in two LWB conditions with various WFS (a) 3.5 m/min and high heat input, and (b) 5 m/min indicators of high and low heat input regimes, respectively [100].

Although WFS did not affect the IMC layer thickness, a low WFS (3.5 m/min) led to dilution of Fe in the braze bead, directly affecting the precipitation of Fe-rich phases (Fig. 2.16). This is because reducing the WFS at constant P/TS increases the heat input per unit length of wire, promoting greater Fe melting and diffusion across the interface, which leads to its mixing with the braze matrix and precipitation within the braze [100]. High Fe dilution into the braze resulted in the formation of various Fe-Si-based and Cu-based IMCs as discussed in Section 2.4.1. Presence of a Fe-based liquid phase at lower WFS (3.5 m/min), which solidified at a higher temperature of 1227°C, contrasted with the absence of this phase at higher WFS (5.0 m/min), where α -Fe/ α -Fe(Si) precipitated directly from Cu-based liquid at 1051°C [100]. This shift in solidification behavior suggests that localized temperatures were higher at lower WFS due to increased Fe dilution, providing indirect evidence of thermal differences associated with changes in WFS. It should be noted that the effect of WFS on thermal conditions during laser

brazing was not quantified through direct temperature measurements. Instead, the inferred temperature variations were estimated from solidification pathways and phase transformation analyses using ThermoCalc. This temperature analysis provided a comprehensive understanding of IMC formation and phase evolution, offering valuable insights into solidification behavior. However, to optimize the LWB process, a more direct and quantifiable measure of heat input, based on adjustable process parameters, is still required. This would enable more accurate predictions of IMC formation and interfacial reactions, facilitate parameter selection before experimentation.

Sarmast-Ghahfarokhi et al. [99] examined the influence of the heat source by comparing GMAB and LWB in BoP configurations. They found that in both processes Fe-Si precipitates formed. However, the distribution and density of these particles differed between GMAB and LWB (Fig. 2.17). In GMAB, granular IMCs were densely concentrated at the center of the braze seam. Conversely, in LWB, Fe-Cu-Si particles were primarily localized near the interfacial zone and had a significantly lower density. The higher fraction of precipitates in GMAB was attributed to the mixing of Fe from the molten substrate with the molten filler material, driven by Marangoni and vortex forces during the GMAB process. Deng et al. [104] further suggested that Marangoni flow, which also occurs in LWB, could facilitate the migration of IMCs from the interface into the Al-based filler/steel braze seam. However, it remains unclear whether similar granular particles can form in steel/Si-bronze LWB joints. Additionally, the precise influence of heat input on the fraction and morphology of these precipitates, in relation to process parameters of GMAB as compared to LWB, has yet to be fully investigated.

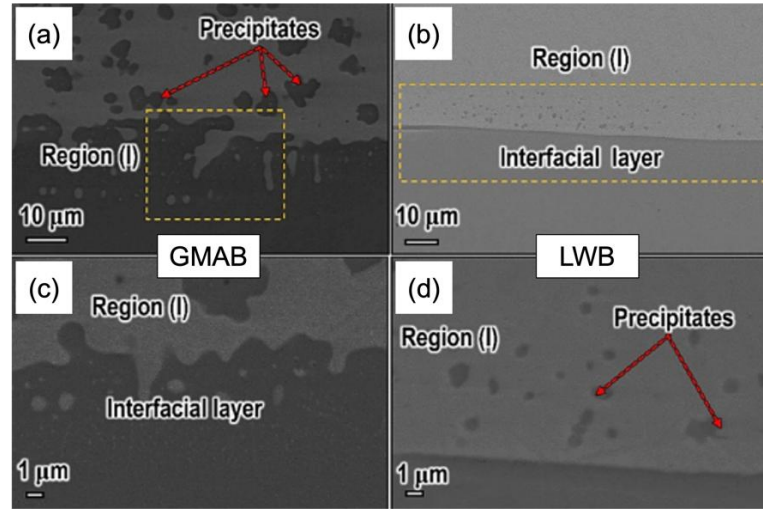


Fig. 2.17: Comparison of interfacial reactions of Zn-Al-Mg coated steel/Si-bronze joint in (a, c) GMAB and (b, d) LWB [99].

2.4.4 Effect of Thermal Gradient and Cooling Rate

Depending on where the filler wire, substrate, and laser are interacting, the temperature gradient and cooling rate may vary [105]. Fig. 13a shows multiple IMC morphology along the interface in a flanged butt joint (as described in [103]) with a variant interface. At position A, in Fig. 13b, the IMC layer is thicker with IMC particles formed adjacent to it due to melting of the base metal resulted from the longer time of laser exposure to heat compared to the lower areas of the joint (position B and C). At position B, in Fig. 13c, IMC layer shape is flat while IMC grows locally into the braze with cellular crystals across the interface and at the presence of IMC particles. A large thermal gradient exists between the molten filler wire and the cold substrate at the early stage of contact providing the driving force to promote IMC growth with flat shape. As laser heating continues, the temperature gradient decreases, and local Si depletion, due to Fe-Si compound formation, leads to constitutional undercooling near the interface. This undercooling allows intermittent IMC growth into the molten filler. However, rapid cooling limits their full development before solidification completes, resulting in the final flat interfacial shape with cellular tips and remained granular IMCs [103]. At position C, in Fig. 2.18d, where the filler metal is at low peak temperature but high cooling rate, Flat IMC

still forms with lower layer thickness and less dispersed granular IMCs. This phenomenon is due to the higher speed of filler metal solidification [103].

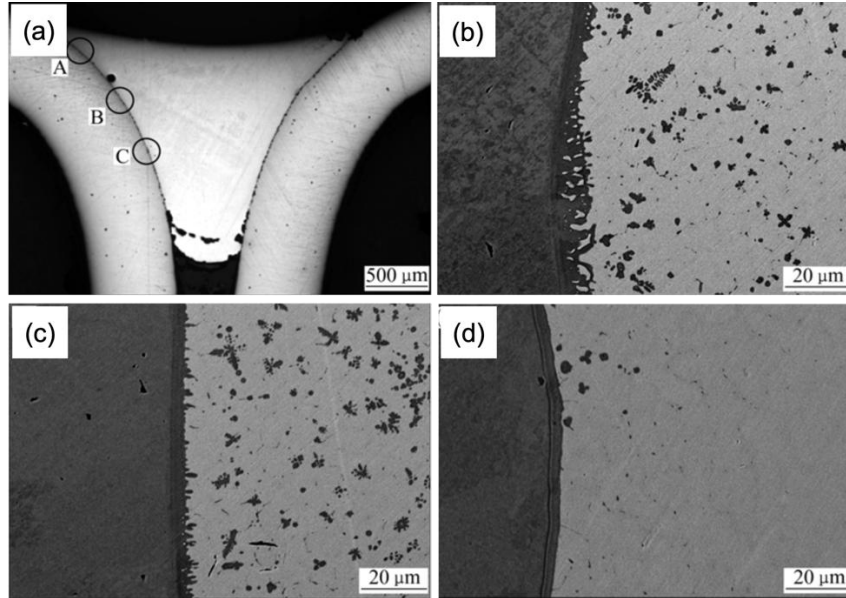


Fig. 2.18: IMC formation in double-flanged braze joint. (a) cross-section view, (b) position 'A', (c) position 'B', and (d) position 'C' from Fig. 2.18a [103].

2.4.5 Correlation of IMCs with Joint Mechanical Properties/Fracture Behavior

The joint interface is the weakest region, primarily due to the brittle nature of IMCs and their influence on bonding strength. However, studies indicate that brazed joints lacking interfacial reactions at the braze/substrate interface exhibit the lowest strength, highlighting the critical role of IMC mechanical and fracture behavior. Joint strength is governed by the composition, morphology, and thickness of IMCs, along with their homogeneity along the interface. Extensive research on Al/steel interfaces has shown that IMC heterogeneity can lead to microcrack initiation and propagation, ultimately compromising joint strength and reliability [106–110]. This underscores the necessity of forming a uniform and cohesive interfacial layer in Zn-coated steel/Si-bronze filler wire brazed joints to ensure robust bonding [111,112].

In GMAB of Zn-coated steels in a lap-joint configuration, failure predominantly occurs near the top sheet interface, where mechanical properties vary based on lap distance and Zn coating

type, both of which influence interfacial reactions [113,114]. Cho et al. [114] reported that increasing the lap distance from near zero to 0.5 mm significantly improved ultimate tensile strength (UTS) as seen in Fig. 2.19. However, further increasing the gap distance beyond 0.5 mm resulted in property degradation, with failure occurring at the interface. Despite these findings, the specific role of IMC types at the interface and the braze microstructure in joints with zero as compared to 0.5 mm lap distance was not further investigated. It remains unclear whether the observed strength variations were primarily due to geometric factors or if microstructural changes within the braze played a significant role.

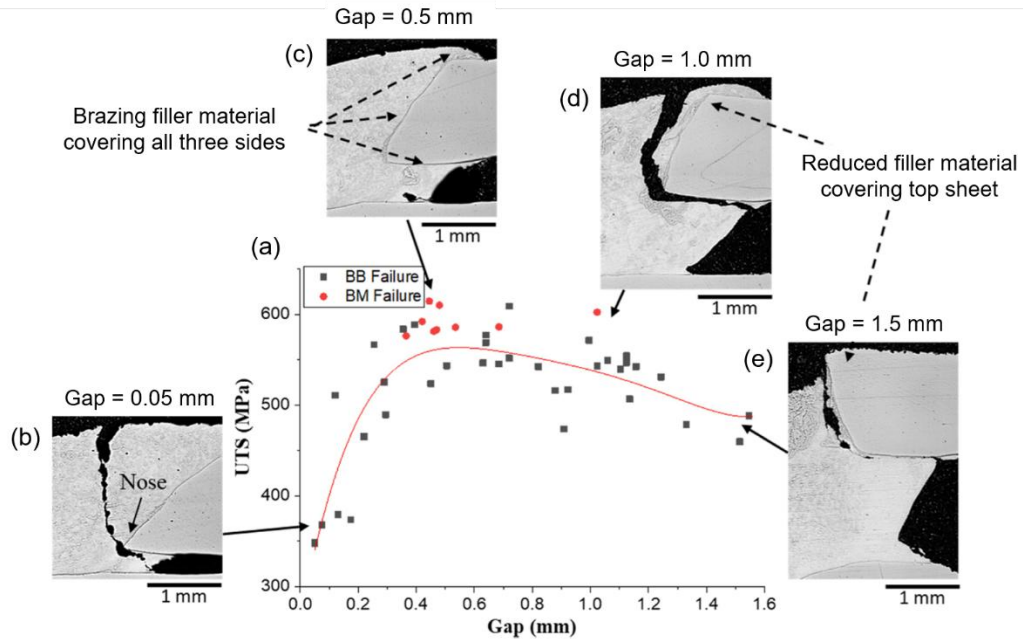


Fig. 2.19: (a) Scatter plot showing the distribution of UTS to the gap size; (b-e) cross-section of the fractured sample with the gap size of 0.05 mm, 0.5 mm, 1.0 mm, 1.5 mm, respectively [114].

Khan et al. [113] reported that joints on GI-coated steel exhibited lower mechanical performance compared to those made on GA-coated steel as seen in Fig 2.20a. Crack initiation occurred at the root region and then propagated into the fusion zone, as shown in Fig. 2.20b and c. Two primary factors contributed to crack initiation at the tail region. First, the Cu–50 wt% Zn alloy formed at the root region of GI-coated joints was significantly more brittle than

the Cu–25 wt% Zn alloy in GA-coated joints, making the GI-coated joints more susceptible to premature failure (see section 2.4.2). Second, loss of cohesiveness at the interface near the edge region further promoted cracking from the tail of the root (Fig. 2.20d and e). In brazes on GI-coated material, this was evident by the formation of a dendritic, non-uniform, and incohesive IMC layer, indicating incomplete bonding, which negatively affected mechanical properties. Notably, the formation of Fe–Si granular particles with a hardness of 711 HV in the GA/Si-bronze seam compared to 586 HV in the GI/Si-bronze seam (Fig. 2.20f and g) was reported as a contributing factor in enhancing the mechanical performance of GA-coated joints. However, it remains unclear whether the overall mechanical performance is predominantly governed by interfacial bonding characteristics or the distribution of these precipitates within the braze.

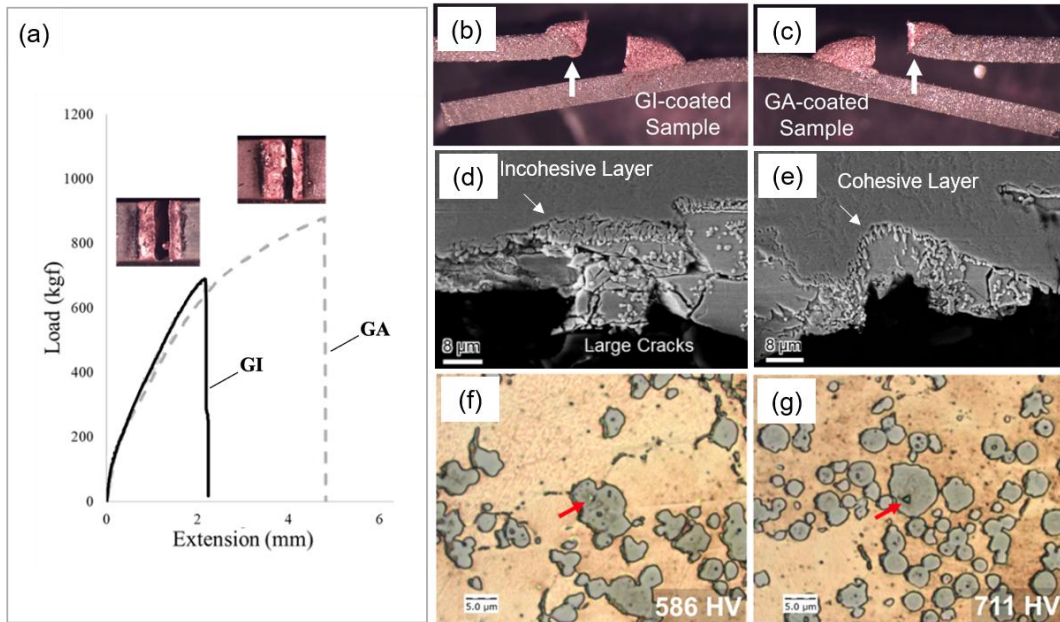


Fig. 2.20: (a) Flow curves of brazed samples with GI and GA coatings. (b, c) Stereoscope images the cross section of the fractured samples. (d, e) scanning electron microscope (SEM) micrographs from the root region of brazes, and (f, g) the bead microstructures showing the microhardness of Fe-Si precipitates within the GI- and GA-coated brazed samples [113].

The mechanical behavior of brazed joints in the presence of Fe-rich precipitates and/or IMCs within the braze region, away from the interfacial area, remains inconclusive in the literature due to the limited number of studies. Zhao et al. [115] recently investigated the impact of distributed IMCs and precipitates, on the mechanical performance of brazed joints in DP steel (Fig. 2.21). Their findings indicate that joint strength is influenced not only by the interfacial reaction layer but also by phases within the braze bead. A high fraction of Fe-rich nodules within the braze can act as crack initiation sites, degrading mechanical strength. Conversely, joints without these phases tend to fracture at the interface, achieving higher tensile load and displacement. Evidence supporting the role of Fe-rich precipitates and IMCs in crack propagation is their presence along fracture paths, aligning with the brittle nature of IMCs that facilitate crack initiation [104,116]. Some aspects remain unexplored in determining whether degraded mechanical performance is solely due to Fe-rich particles formed in the braze. For instance, the influence of a pre-existing hot crack beneath the braze bead at a wire feed rate of 3 m/min, acting as a crack initiation site, warrant further consideration. Moreover, it has been noted that the concave bead geometry observed at lower WFS (e.g., 3 m/min) suggests insufficient wire deposition. However, the smaller bead size compared a joint made with higher WFS might have contributed to mechanical performance degradation. Additionally, further investigation is needed to optimize wire deposition to minimize geometry-related effects, while maintaining a high fraction of granular IMCs is present within the braze.

Zhao et al. [100] also reported that an increased fraction of IMCs and other precipitates within the braze region increases microhardness. The microhardness of the bead decreased with decreasing WFS, from 117 ± 2.3 HV at a WFS of 5 m/min to 143 ± 2.5 HV at a WFS of 3.5 m/min. This increase in hardness is attributed to dispersion hardening, where hard-phase particles reinforce a softer matrix, a known strengthening mechanism expected to enhance joint mechanical properties [117–119]. The reason why a high fraction of precipitates within the braze degrades mechanical performance remains unclear, as dispersion hardening is generally expected to enhance strength in other applications. This effect may be attributed to specific types of precipitates within the braze that contribute to reduced mechanical properties rather

than improvement. Further investigation is needed to determine whether the dispersion of other types of IMCs would similarly degrade mechanical performance or potentially offer strengthening effects.

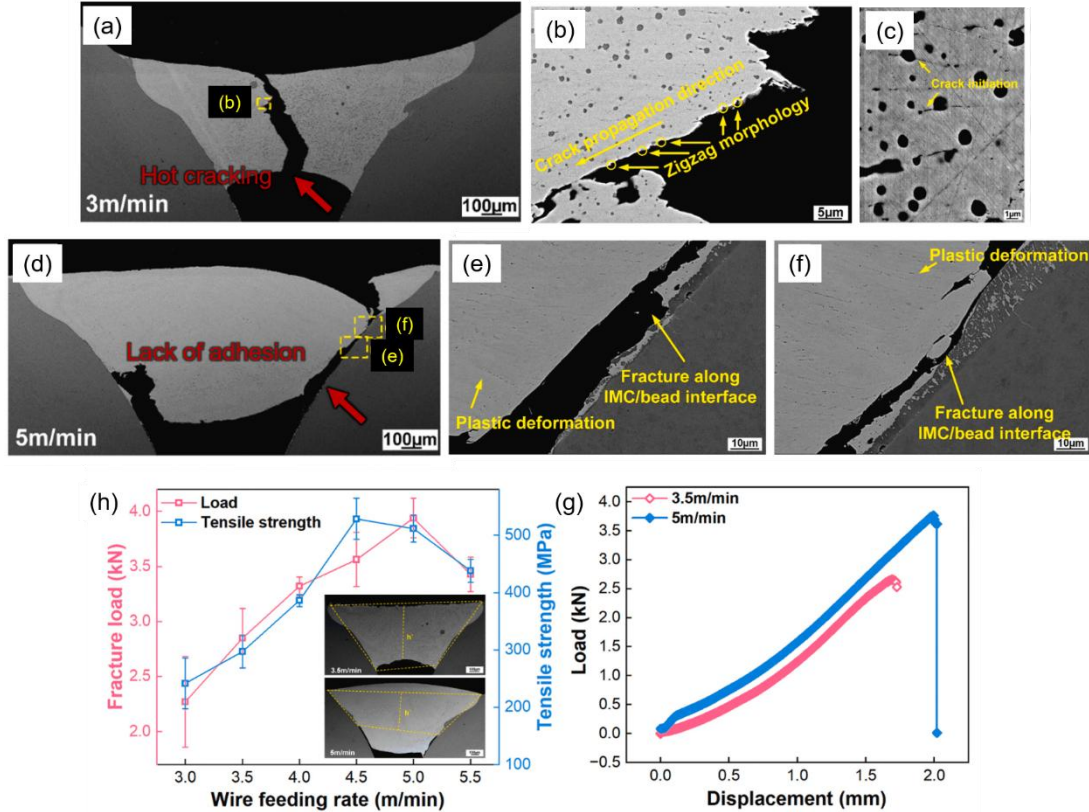


Fig. 2.21: Fracture path and tensile properties of the joints brazed with different WFS: SEM micrographs of joints with WFS of (a-c) 3 m/min and (d-f) 5 m/min. (h) Average fracture load and the tensile strength. (g) Load-displacement curves of the joint brazed with the WFS of 3.5 and 5 m/min [115].

Sarmast-Ghahfarokhi et al. [120] reported improved mechanical properties in the presence of Fe-Si(Cu) precipitates within the braze bead of GMAB brazed joints in IF steel (Fig. 2.22). They found that using a precision pulse arc wave module (high heat input), which is a modified spray metal transfer, form brazes had higher tensile peak load and extension due to the higher density of precipitates with irregular morphologies within the braze compared to surface tension transfer module (low heat input), that is a control short circuit metal transfer. From

instrumented indentation testing they explained these precipitates lead to local strengthening by 30% increase in strength. However, given the presence of complete bonding in the precision pulse samples and incomplete bonding in the surface tension transfer samples (as seen in Fig. 2.22a and b), along with the significantly larger braze bead in the precision pulse condition, it remains uncertain whether these factors also contributed to the observed mechanical performance improvements. Further investigation is needed to isolate the specific role of precipitates in strengthening mechanisms.

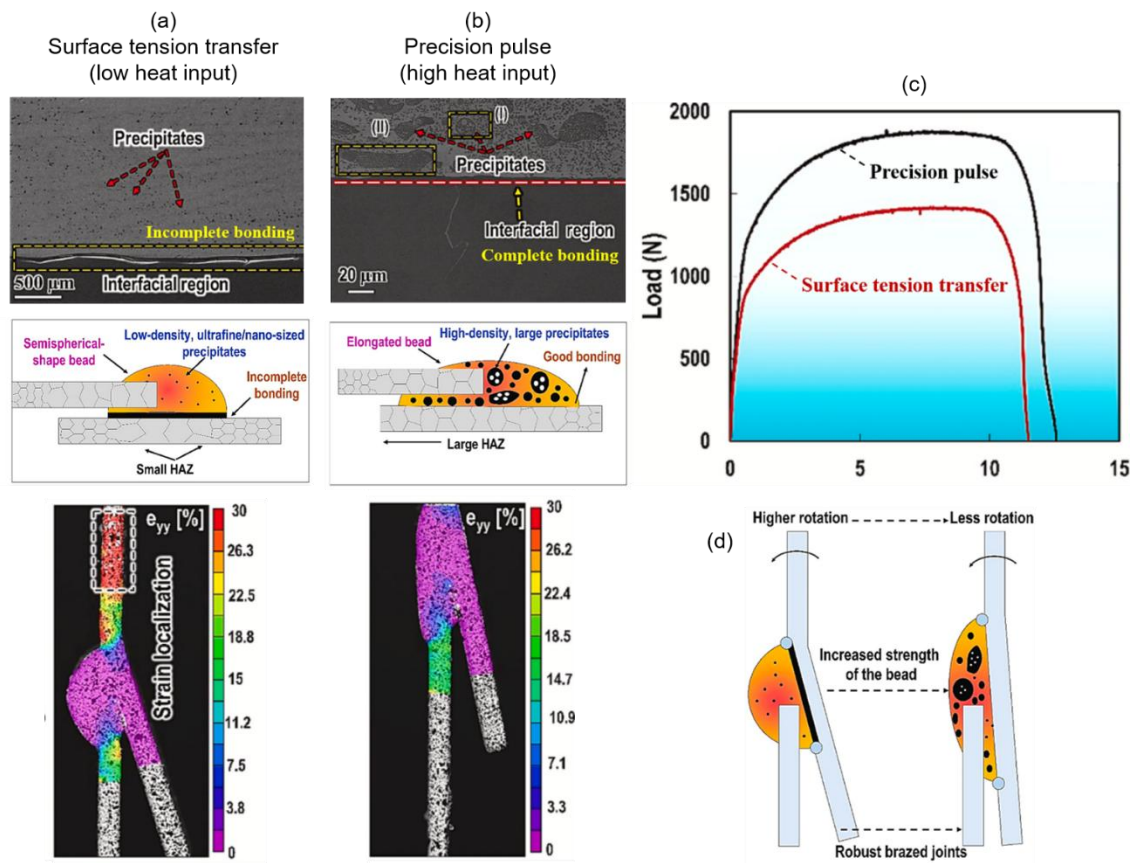


Fig. 2.22: Comparison of the braze microstructure before the tensile (SEM micrographs and schematic) and DIC imaging of strain distribution during tensile testing before the failure tensile properties between two GMAB modules (a) surface tension transfer and (b) precision pulse. (c) Load-extension curves during the shear-tensile test. (d) schematic illustration of the tensile response of the braze joints [120].

In a subsequent study [24], Sarmast-Ghahfarokhi et al. investigated GMAB brazed joints in IF steels, produced at different brazing speeds, 15 in/min and 30 in/min, representing high and low heat inputs, respectively. Contrary to the findings from [120], the joint with lower heat input, which had a lower fraction of Fe-rich precipitates, exhibited superior mechanical performance (Fig. 2.23). Shear-tensile testing of the joints revealed that, despite differences in bead geometry, microstructure, and interface characteristics, the peak load remained nearly identical between the two heat input conditions. However, extension increased as heat input decreased, with both samples failing within the base metal as seen from Fig. 2.23. Both conditions exhibited a ductile failure mode, as failure occurred within the base metal (Fig. 2.24), which is consistent with the inherent ductile fracture behavior of IF steel.

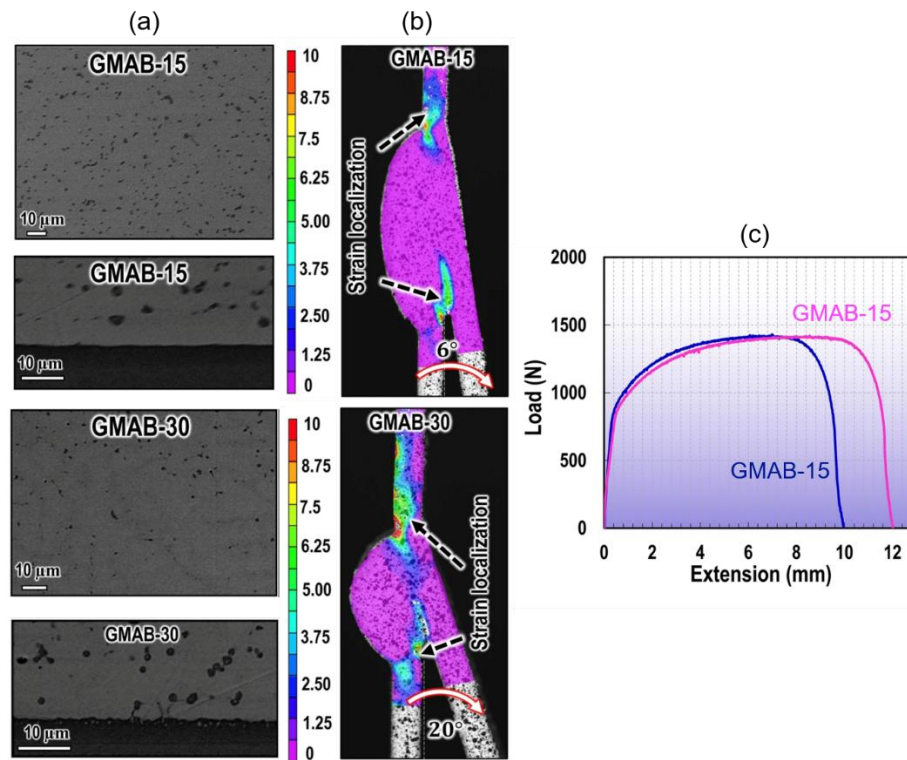


Fig. 2.23: Comparison of the braze mechanical performance of two GMAB conditions with high heat input (GMAB-15) and low heat input (GMAB-30) showing: (a) SEM micrographs before the tensile, (b) DIC imaging of strain distribution before the failure, (c) fracture tensile samples showing failure from the base metal along with its corresponding fracture surface morphology, and (d) tensile load-extension curves [24].

Finite element analysis (FEA) was further incorporated in [24] to simulate the tensile testing as seen in Fig. 2.24. It was found that beyond bead geometry and microstructure, loading conditions, particularly bending and rotation of the bottom sheet during shear-tensile testing, significantly influenced mechanical properties and failure behavior. The degree of rotation during lap-shear testing was identified as a critical factor influenced by braze geometry, with GMAB-30, which had a smaller bead size, exhibiting a higher degree of rotation. This, in turn, affected strain distribution and contributed to the observed mechanical performance trends, whereas in the precision pulse sample [120], the lower degree of rotation was associated with improved joint robustness during tensile. They reported that bead characteristics (precipitates) alone do not directly correlate with joint mechanical performance. Despite differences in brazing conditions, the peak load remained comparable across samples, and the ability of the braze coupon to bend and high degree of is the reason of higher displacement.

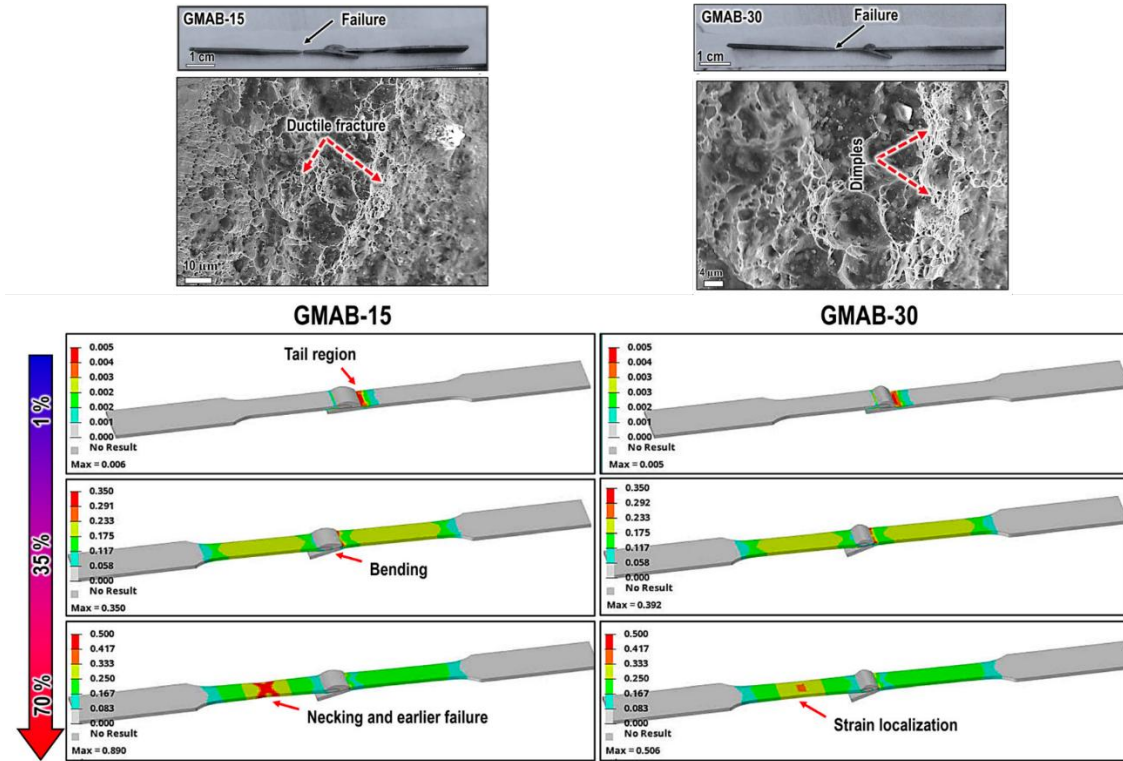


Fig. 2.24: FEA analysis at different loading stages of shear-tensile loading test for both GMAB-15 and GMAB-30 joints simulated the tensile behavior of braze samples indicating that GMAB-30 with a certain braze geometry despite of its smaller bead size and lower fraction of precipitates experienced delayed strain localization compared to GMAB-15 leading to its higher extension. FEA results also aligned with the base metal failure and its corresponding ductile fracture morphology of both samples [24].

In a separate study [121], the mechanism of precipitation hardening in a braze with a high density of precipitates was further examined using ball indentation testing, along with transmission electron microscopic (TEM) and electron backscattered diffraction (EBSD) analyses. The presence of nano- and micro-sized precipitates significantly contributes to shear banding as seen from Fig. 2.25a and b. It was reported that micro-sized precipitates promote shear banding by affecting strain distribution, while nano-sized precipitates contribute to dislocation interactions that facilitate nano-twin formation. These nano twins further refine the shear bands, influencing strain distribution and strengthening mechanisms [121]. The study also observed that the presence of precipitate particles results in an increase in the strength of

the Cu matrix. Additionally, nano twins lower the critical stress necessary for deformation, further influencing strain accommodation. The inverse pole figure (IPF) and kernel average misorienting (KAM) analyses in Fig. 2.25c-f confirm that these structural changes influence localized strain accumulation, reflecting deformation heterogeneity within the material during indentation. The dispersion of precipitate particles alters strain distribution, as evidenced by the localized plastic strain observed in the KAM maps, highlighting their role in strengthening mechanisms under indentation loading [121].

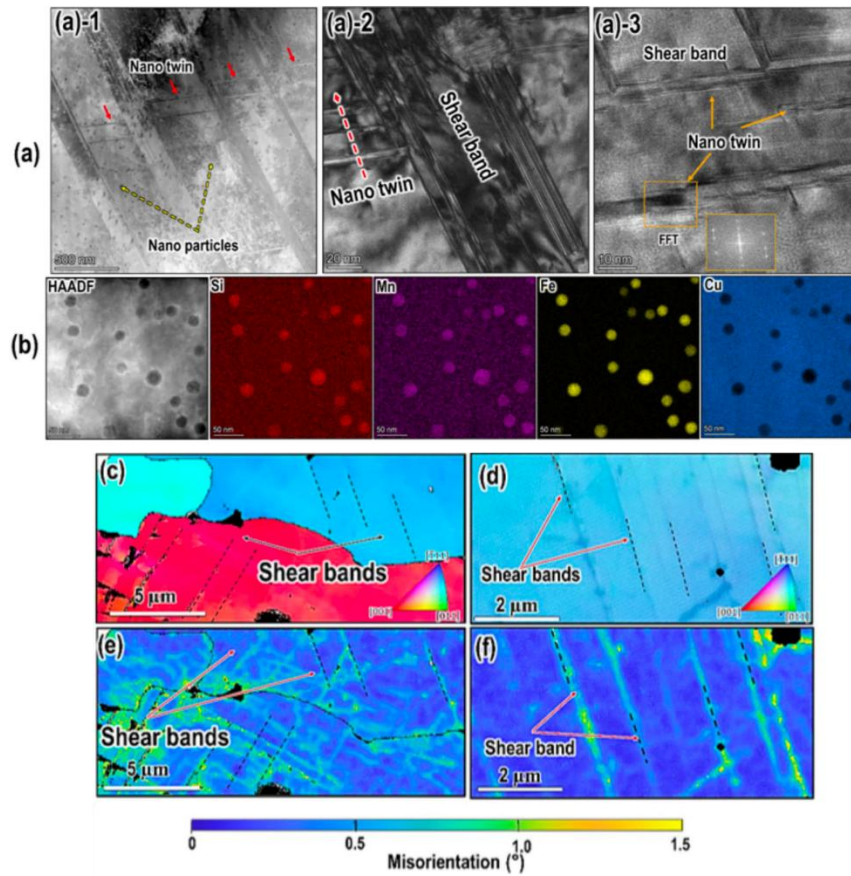


Fig. 2.25: (a) TEM micrographs of the nano twins formed within the shear bands and the matrix in presence of precipitates. (b) elemental maps of the nano-sized precipitates show that the particles have the same composition as the micron-size particles. The figure shows that the. (c,d) IPF map of in-grain shear bands and (e,f) the corresponding KAM of the shear bands [121].

The strengthening effect of precipitates and/or IMCs within the braze has been identified in [121], and improved mechanical performance has been reported for brazes with a high precipitate density in [120]. In contrast, other findings suggest a different perspective. Specifically, [24] indicates that bead geometry characteristics have a more dominant influence on mechanical performance than precipitation density, and [115] reports a degrading effect of IMC particles. These conflicting observations introduce further uncertainty regarding the exact role of IMCs and precipitates in overall mechanical performance and fracture behavior.

Another aspect to consider, deformation under indentation is primarily compression-driven, differing from tensile loading, which influences twin activation depending on grain orientation. The activation of twinning observed in a work of Sarmast-Ghahfarokhi et al. [121] occurred at a nanoscale under compression loading conditions during indentation testing. Whether this behavior translates to tensile testing for these samples remains unexplored. According to Schmid's law, plastic deformation, including slip and twinning, occurs when the resolved shear stress on a particular system reaches a critical value, which depends on the loading direction relative to the crystal orientation. As shown in Fig. 2.26, Schmid factors for slip and twinning vary under tensile (a) and compressive (b) loading, where grains favoring twinning in tension may not do so in compression [34,122,123]. This is because the angles defining the resolved shear stress φ (the angle between the loading direction and the slip/twin plane normal) and λ (the angle between the loading direction and the slip/twin direction) change with loading conditions, influencing which deformation mechanism is activated. Additionally, since the Cu-based braze has face centered cubic (FCC) crystal structure, twinning is expected if the stacking fault energy (SFE) falls within the low to medium range, as lower SFE reduces cross-slip and increases dislocation pile-up, making twinning more likely. However, since deformation under indentation is compression-like, the activation of twinning under tensile loading may be different due to grain orientation effects. The extent of twinning activation during tensile loading could be higher or lower, requiring further investigation to accurately assess the effect of precipitates and/or IMCs at a larger scale, considering more grains.

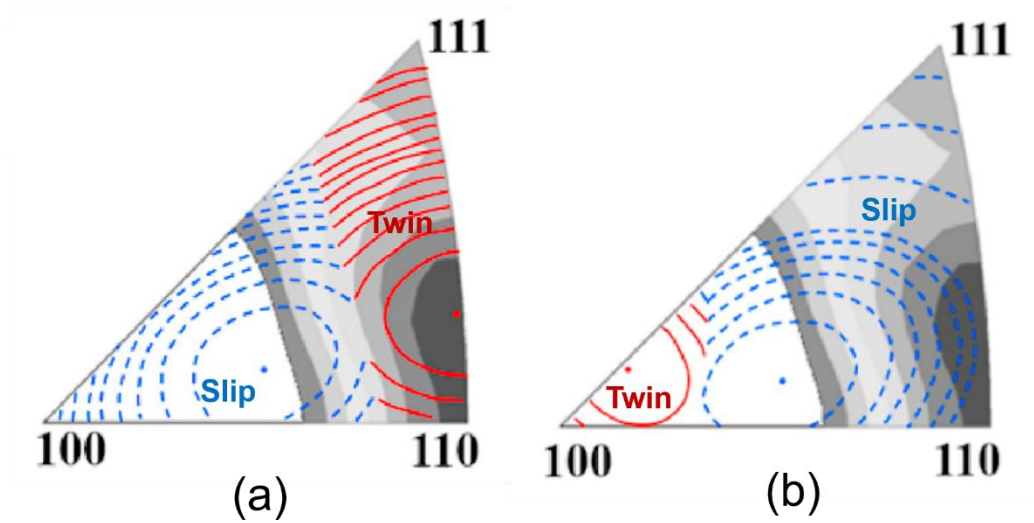


Fig. 2.26: Orientation distribution stereographs of a twining-induced-plasticity steel under (a) tensile and (b) compression loading modes. Schmid factors corresponding to the slip and twin are highlighted by the blue and red lines, respectively, showing the favorable regions for slip and twinning [34].

Despite extensive research on the chemistry of Fe-Si IMCs, their impact on joint mechanical performance remains a subject of debate, with conflicting findings in the literature. The role of IMC fraction and morphology, whether beneficial or detrimental, has yet to be clearly defined. To provide a comprehensive overview of these discussions, Table 2.1 summarizes key findings from existing studies.

Table 2.1: The most recent studies on effect of IMC/precipitate/Fe-rich phases dispersion on mechanical properties of the braze joints.

Year, Ref.	Base material / brazing method	Braze characteristics	Reported correlation between IMC and mechanical properties	Unexplored aspect
2024, [115]	DP steel / LWB	- Braze with high fraction of precipitates with smaller bead. - Braze without precipitates and larger bead.	- Failure from braze region with precipitates led to lower tensile load and displacement. - Failure in braze without precipitates occurred from interfacial IMC led to higher strength.	- Effect of bead size. - Effect of the differences between hardness and chemistry of interfacial IMC versus precipitates.
2025, [100]	DP steel / LWB	Braze containing high fraction of precipitates.	Increase in microhardness of the braze led to degraded mechanical properties.	Increase in hardness of braze may be due to dispersion hardening.
2023, [120]	IF steel / GMAB	- Braze with high fraction of precipitates with larger bead. - Braze with low fraction of precipitates and smaller bead.	- Instrumented ball indentation testing showed that precipitates enhance strength through localized strengthening - Tensile peak load and displacement increased by precipitates.	- Effect of bead size (explored in [24]). - Failure occurred from the base metal, softer matrix, in both, make it hard to understand if the braze played a decisive role in overall strength.
2024, [24]	IF steel / GMAB	- Braze with high fraction of precipitates with larger bead. - Braze with low fraction of precipitates and smaller bead.	- High fraction of precipitates resulted in lower displacement in tensile, while the peak load remained unchanged. - Bead characteristics do not directly correlate with mechanical properties. - An ability to bending and high degree of rotation were found to be the reason of higher displacement.	- It remains unclear how a higher degree of rotation and bending were resulted from a smaller braze. - It contradicts with the finding in [120], that a low degree of rotation correlates with robustness of the joint.
2024, [121]	IF steel / GMAB	Braze with precipitates within its coarse grains.	Instrumented ball indentation showed that nano-sized precipitates promote nano-twin formation within shear bands, enhancing deformation behavior.	Effect of precipitate strengthening on tensile properties needs further investigation.

2.5 Summary of research gaps in the literature and direct pathways to address

LWB has emerged as a promising alternative to fusion welding, particularly for automotive exterior panels, offering high-speed joining with high precision heat delivery and minimal operating heat input, thereby reducing Zn burn-off and preserving surface integrity. The widespread use of Zn-coated steels in automotive applications adds complexity to the joining process. Variations in coating type and steel composition affect how the base material reacts with the laser beam and the Si-bronze filler material, influencing joint formation. At the same time process parameters, affecting heat input, and joint configuration play a critical role in determining joint quality.

While GI-coated steels have shown improved wettability in GMAB compared to GA-coated steels, their behavior in LWB, where a laser beam melts the filler wire instead of an electric arc, remains unexplored. Understanding how laser-wire interactions with different coatings influence substrate surface morphology and braze segregation, compared to arc-wire interactions in GMAB, is crucial for establishing fundamental knowledge of LWB's distinct wetting behavior. This gap is systematically investigated in Chapter 3 based on Objective 1.

A key challenge in LWB is optimizing process parameters due to the complex interactions between principal laser parameters such as P , TS , and WFS , which are not yet fully understood. The widely used $HI_{Nominal}$ oversimplifies thermal interactions by overlooking variations in beam absorption efficiency caused by filler wire deposition. As a result, process optimization largely depends on empirical trial-and-error methods rather than a predictive framework that can anticipate brazing defects, IMC formation, and microstructural evolution. To improve process efficiency and joint reliability, a more precise optimization strategy is required, one that quantifies heat input using adjustable process variables to enhance control over interfacial reactions and microstructure evolution. The present study directly addresses this need by establishing a wire-adjusted HI approach that more reliably reflects the thermal conditions during LWB, enabling accurate prediction of different features within the braze including sub-surface defects and control over IMC formation, as outlined in Objective 2 and detailed in Chapters 4 and 5.

Despite advancements in AI-assisted UT for real-time inspection, its adoption in LWB, particularly for Zn-coated steels, is limited due to challenges such as wave attenuation, IMC-induced microstructural complexity, and seam geometry variations. The industry continues to rely primarily on optical and X-ray inspection, while efforts to enhance UT-based solutions are ongoing. A major obstacle to perform AI-driven in-line UT is the absence of a structured LWB optimization framework that integrates adjustable process parameters to systematically control defect formation. Without a methodical approach to inducing and categorizing defects under controlled conditions, establishing clear correlations between process parameters and defect characteristics remains challenging. This hinders AI model training, reducing defect prediction accuracy and preventing the seamless integration of AI-driven inspection systems into real-time quality control frameworks. For effective implementation, AI models require diverse, high-quality datasets that accurately capture defect formation under varying LWB conditions. Each brazed seam produced under specific parameter settings must demonstrate unique defect characteristics, such as type, size, and distribution, directly influenced by LWB variables. A robust AI-driven inspection system highly depends on systematic curation of datasets that reflect these variations while maintaining consistency in defect representation within the braze. Chapter 4 of this study addresses these challenges by developing a robust process window that enables controlled defect formation, ensuring a diverse, reliable, and consistent dataset for AI-enhanced UT inspection as outlined in Objective 3.

In addition to brazing sub-surface defects, IMC formation plays a crucial role in LWB joint integrity, yet several aspects of their formation remain uncertain, which further hinders enhancement of braze joints mechanical performance. Recent studies have provided significant insights into phase stability and transformation mechanisms, but critical gaps persist regarding a formation of IMCs under a higher extent of Fe diffusion into the braze, the formation of IMC morphologies beyond granular structures. Examining the effect of forming Fe-rich phases within the braze bead is critical, since it may affect the joint mechanical performance as well. In LWB at a higher heat input, there is a risk of thickening the interfacial layer degrading the mechanical performance. Achieving high mechanical performance in brazed joints and

minimizing the risk of interfacial fracture typically requires a thin IMC layer at the interface of defect-free braze seam. However, no further efforts were made to explore methods for modifying braze microstructure to improve its mechanical performance. This study aims to address this gap by maximizing braze joint's strength and ductility through promotion and dispersion of additional phases within the braze microstructure, inspired by the idea of dispersion hardening. The potential for achieving such a microstructure can arise from in-situ alloying; however, this approach may not be viable due to the complexity of selecting new elements and the uncertainty of their interactions with existing constituents. A more practical alternative can be to leverage the intrinsic dilution of Fe atoms into the braze, which naturally occurs during WB of steels, containing high amount of Fe, with Si-bronze filler material, sharing its Si atoms to form Fe-Si IMCs. IMC formation was found to be detrimental to joints mechanical performance, while a crucial gap is the lack of efforts in strategically modulating the Fe dilution and IMC formation to modify the braze microstructure and mechanical performance. This gap shaped the foundation of Objective 4, which is addressed in Chapter 5 by developing a novel pathway to manipulate IMC formation using the wire-adjusted HI approach that promotes a higher fraction of IMCs within the braze interior region under controlled elevated heat input. It unveils completely new insights into IMC formation, which have not yet been explored in the literature, and examines their subsequent impact on mechanical performance.

The optimization criteria for enhancing mechanical performance, particularly regarding the interfacial IMC layer, remain consistent across literature, with broad agreement that reducing interfacial IMC thickness and hardness is crucial for minimizing embrittlement and preventing interfacial failure. However, the literature on the impact of Fe-rich IMC particles or precipitates, often referred to interchangeably in some works, on mechanical performance under higher absorbed heat input is limited, with findings across different works being inconclusive. Some studies suggest these IMC particles act as crack initiation sites, leading to mechanical degradation, while others indicate that dispersion hardening mechanisms contribute to strengthening. The increase in hardness associated with IMC dispersion should,

in theory, enhance joint performance, yet observations linking IMC presence to crack propagation under tensile loading raise questions about their overall effect. Whether the reduction in peak load and displacement is solely attributable to IMC particles or influenced by additional factors such as joint geometry and strain distribution remains unclear. Some studies report that Fe-rich precipitates promote shear banding and nano-twin formation, leading to improved mechanical properties. In contrast, some studies suggest that bead geometry characteristics, including bead size and precipitation, have no direct impact on peak load in tensile tests. They indicate that tensile displacement increases primarily when the braze coupon can bend and rotate, though further clarification on the underlying mechanisms remains a gap. Thus, the effect of dispersion hardening by IMCs or precipitates on mechanical performance remains unclear, as the concept is immensely complex and has not yet been universally concluded as beneficial or detrimental. Additionally, the reliance on indentation-based hardness testing to assess local mechanical properties, which operates under compressive loading conditions, does not fully capture the material's response under tensile stress. A more comprehensive evaluation of IMC behavior across a wider range of IMC morphologies and distributions is necessary to establish a clearer correlation between IMC dispersion and mechanical performance. To address this gap, Objective 5 needs to be addressed. For this purpose, Chapter 6 of this study advances the understanding of IMCs' effect on the fracture behavior of LWB joints by systematically characterizing new IMC classes and thoroughly examining their contribution in response to the tensile loading and crack propagation. Key to understanding how these IMCs contribute to deformation is to characterize how the presence of IMCs alters crack propagation path, fracture modes, underlying deformation mechanisms, and texture of the braze, rather than only relying on their chemical composition. This aligns with Objective 5's goal of advancing the understanding of IMCs' effect on fracture behavior in LWB joints.

Chapter 3. Effect of Heat Source and Zn-coating Type on the Geometry and Morphology of Laser Weld Brazed Thin-gauge steels²

3.1 Overview

LWB offers significant advantages over other types of traditional seam joining methods: the ability to join dissimilar materials without significant melting of the substrates, reduced Zn burn-off in Zn-coated steels, and minimal HAZ formation. However, the edge (or the tip) of laser-brazed beads can be one of the potential cracks initiations sites which results in catastrophic failure of the parts. The existing literature on LWB offers no insights into the geometry and elemental segregation at the edge of the brazing bead as a result of the interaction between the laser beam and the substrate. To bridge this gap in knowledge, LWB was conducted in the bead-on-plate configuration on two different types of Zn-coated hot-dip GI- and GA-coated AHSS using a Si-bronze filler wire. LWB provided an improved wettability because of the formation of a tail-like geometry at the bead edge of both coatings showing a high Zn concentration, which is opposed to what has been observed in similar studies on GMAB. LWB had a significant effect on modifying the surface morphology of the GA-coating, which has previously not been reported in the literature. Similar Zn distribution of LWB samples at the bead edge of both coatings clearly showed that the wettability of molten filler material in LWB is improved relative to GMAB. A graphical abstract summarizing this chapter is presented in Fig. 3.1.

² This chapter is composed of a published research paper as outlined in the statement of contributions. **S. Akbarian**, M. Shehryar Khan, O. Sherepenko, M. Lee, P. Wanjara, E. Biro, Effect of heat source and Zn-coating type on the geometry and morphology of laser-brazed thin-gauge steels, J. Manuf. Process. 107 (2023) 356-367.

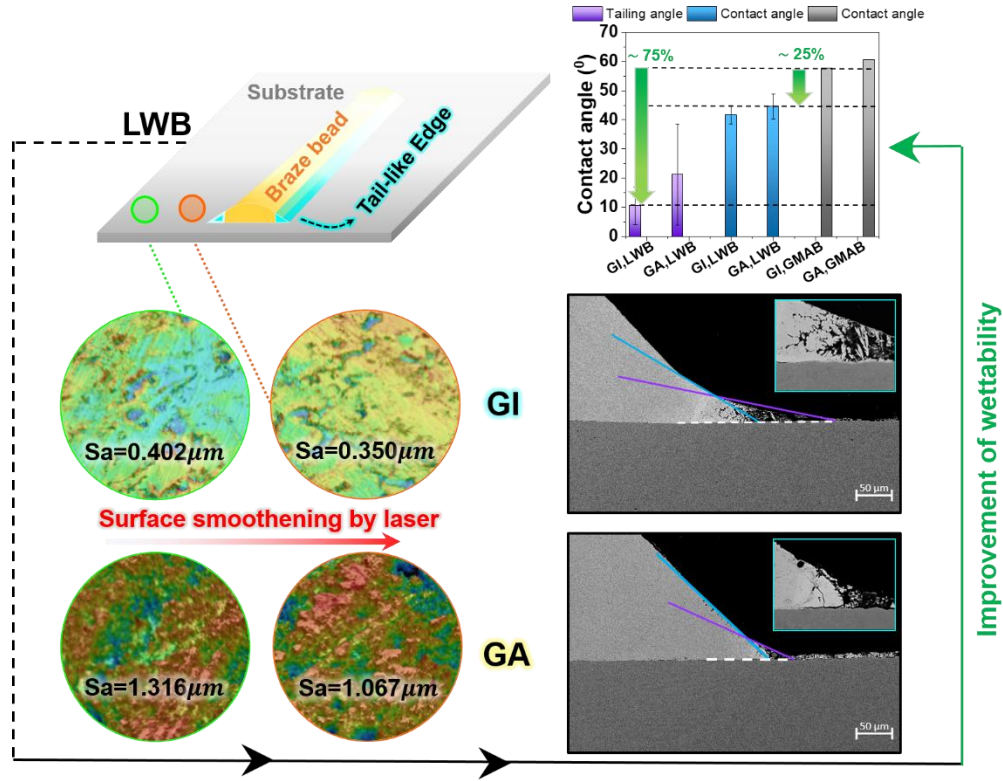


Fig. 3.1: Graphical abstract illustrating an overview of Chapter 3.

3.2 Background

DP steels are one of the most commonly used grades of AHSS in the automotive industry for the production of BIW structures, owing to their excellent combination of high strength and high ductility [48,124]. However, similar to other common steel products, these thin-gauge steel sheets exhibit poor corrosion resistance properties in service. Thereby, they are conventionally protected using a Zn-coating [33], which offers a desirable degree of galvanic protection [125]. During the assembly of different parts of the BIW structure, the appearance of the joint may be of critical importance, particularly when joining the rooftop and decklid to the side members, because the joints may be visible to the end user [126]. These types of joints require a “seamless” appearance with minimal reworking required before painting [45]. Conventional joining methods, such as roof molding on the spot-welded joints, are becoming obsolete in the joint design of the rooftop and decklid in modern vehicles. A suitable alternative

that has gained popularity in recent years includes fusion welding [127–130] or weld-brazing [101,113,131–134] techniques as two options to achieve a slit-less joint design.

Fundamental differences exist when comparing fusion welding and weld-brazing with each other. In welding, the creation of a bond between the substrates and filler wire requires melting and re-solidification of both metals. This, in turn, requires a processing temperature higher than the T_m of both the substrate and filler materials [17]. The main drawback associated with the fusion welding of Zn-coated steels is Zn burn-off due to the severe vaporization of the Zn element at the surface of the substrate as it is exposed to temperatures above T_b , the boiling temperature of Zn. Specifically, the T_b of Zn is far lower than the melting temperature T_m of the steel substrates ($T_b(\text{Zn}) = 907^\circ\text{C}$ and $T_m(\text{Steel}) = 1425\text{--}1540^\circ\text{C}$), which eventually causes Zn vaporization [135]. Zinc vaporization results in welding defects; for instance, damage to Zn protection, lack of adhesion, and porosity are the main concerns that can act as crack initiation sites, deteriorating the mechanical performance [41,46]. During weld-brazing, on the other hand, only the filler metal is melted; this means that the processing temperature only needs to exceed the T_m of the filler metal, which is typically selected to be lower than the T_m of the substrates [42]. By avoiding substrate melting, Zn vaporization, which is one of the main detrimental concerns of fusion welding, can be minimized near the periphery of the molten wire/substrate interaction zone [136–138]. Furthermore, the lower T_m associated with brazing reduces the necessary heat input and mitigates drastic phase changes in the microstructure of the substrates by limiting the formation of undesired phases or changes to the grain structures [139]. In this context, WB is also referred to hereafter as brazing for short.

Considering the issues that arise from fusion welding, brazing can be a viable alternative for joining BIW structures fabricated from AHSS [118]. However, optimizing the brazed assembly and the creation of a strong bond, are dependent on the quality of the brazing bead, which is affected by the wetting mechanism at the interface between the molten filler and the substrate(s). The successful bonding of two dissimilar metals at their interface is closely associated with the formation of IMC phases [89]. The thermodynamic stability of IMC phases

determines which compound would form within the interface based on the phase diagram corresponding to the existing elements [90]. However, these IMC phases are brittle and can act as crack initiation sites, posing a considerable challenge to brazing applications [91]. One approach to modify IMC formation and the brazing bead quality is the manipulation of the heat input and distribution, which are both dependent on the process parameters and the type of heat source [49]. Arc and laser are the most common heat sources used for wire-fed welding and brazing within the automotive industry [136,140]. Although the use of an arc in brazing is economical and conventional, the benefits of using a laser heat source outweigh the advantages of an arc process [141]. Compared to GMAB, LWB enables a higher joining speed by providing high energy density using a laser heat source, which is more consistent and controllable compared to an electric arc [47,126,142]. Designing intricate details in BIW structures with high precision and less thermal damage to the substrate is possible through the enlargement of the laser beam spot size by altering the working distance, which refers to the distance between the laser focal point and the targeted joining surface. By increasing or decreasing the working distance, the spot size can be modified, resulting in a finer braze geometry due to lower heat input. This capability provides a significant advantage in achieving precise and intricate design features [143]. Concerning the IMC issues in joining applications, many studies in LWB have shown that by optimizing factors such as IMC thickness [94,110,144], morphology, distribution [145–147], and chemical composition [104,148,149] at the interface, the harmful effects of the IMC layer can be minimized.

Despite the rich body of literature on brazing bead brittleness at the interface, studies on the wetting profile along the edge of the joint during brazing of Zn-coated steels and its significant effect on the subsequent mechanical performance have been neglected [133,150]. This region can potentially act as the first crack initiation site during loading of overlapped brazed joints [65,114,151]. Khan et al. [113] showed crack initiation and propagation occur through the root region of arc-brazed lap joints due to the presence of brittle Fe-Zn compounds present at the tail of the root. The brazing bead edge properties depend on the wettability of the molten filler metal on the substrate, which in turn is dependent on the surface morphology of the substrate

and the chemical composition of the materials engaged in the process. The rougher the surface morphology the lower the chances of achieving good wetting. Fundamentally, the wettability can be determined by measuring the contact angle, θ , at the wetting profile edge, or the bead edge in brazing. A smaller contact angle shows higher wettability of the molten filler onto the substrate [49,152]. As the roughness increases, more surface irregularities are created acting as energy barriers against molten filler to spread on the surface leading to the formation of a brazing edge with the contact angle increasing as the wettability decreases [153,154]. Regarding the chemical composition of the substrate/filler, primarily the incompatibility of the chemical composition of the substrate with that of the filler and subsequently the chemical reactions during heating/solidification cycles of the LWB process can cause elemental segregation affecting the wetting behavior and brazing geometry [155–157]. Achieving better wettability leads to an improved bead edge, which enhances resistance to cracking. Among Zn-coated steels, GI- and GA-coatings are the two most widely used coating types in the automotive industry [40]. GI consists of pure Zn with a very thin layer of Fe-Al IMCs located on the substrate surface. However, GA is the annealed version of the GI, containing a variety of Fe-Zn compounds at the surface to enhance corrosion resistance [158,159]. Khan et. al. [101] conducted GMAB in the BoP configuration with Si-bronze filler wire on GI- and GA-coated DP600 steel substrates. Their work showed that the wettability of a GI-coating was better than a GA-coating. The surface roughness of these substrates was found to be a major contributor to the wettability of the molten filler metal, which is lower in the GI- than GA-coated steel. This is due to the existence of pure Zn, which reduces the surface roughness by soaking the sheet of steel in a bath of molten Zn. In GA-coating, however, the formation of Fe-Zn IMCs at the surface causes an increase in roughness [160]. As a result, the contact angle at the brazing bead edge was smaller in the case of GI- than GA-coated steel, which is associated with a “tailing out” effect by an extension of the wetting profile beyond the apparent end of the brazing bead edge, taken as a sign of good wettability. In LWB, according to a study by Reimann et. al. [133], reflectivity of the surface was observed to affect heat absorption when comparing different types of Zn coatings with different reflectance. They found that the surface

of hot-dip GI Zn-coated substrates had lower heat absorption caused by higher surface reflectivity than electro-galvanized steels when interacting with the laser radiation [133]. In fact, surface roughness and reflectivity are two complementary factors resulting from the surface morphology with an inverse relationship [161].

Current, no work has been done to understand the effect of surface morphology and surface chemistry of Zn-coated steels on the brazing bead edge properties in LWB. Therefore, this study investigated the effect of laser heat source on the wetting process, bead edge morphology, and diffusion behaviour of the laser brazed joints of two different substrate coatings. As a way to understand the role of using a laser heat source in LWB as opposed to an electric arc in GMAB, the substrate/filler materials used in LWB were selected, similar to the study on GMAB of GI- and GA-coated steels by Khan et al. [101], to offer a direct comparison between LWB and GMAB in reacting with various surface morphology and chemistry.

3.3 Experimental Procedure

3.3.1 Materials Selection

This study was conducted using two industrially-produced DP600 sheets of steel. The steels were industrially hot dip coated, one with a GI and the other with a GA coating. The braze material used in the study was a copper-based filler metal, commercially called Si-bronze or CuSi3Mn1 (AWS ERCuSi-A) with T_m of 1000 °C as reported by the manufacturer. The coating weight of the GI- and GA-coated steel sheets is 51.0 and 51.3 g/m² at the top and 51.6 and 50.1 g/m² at bottom of the sheets. The total thickness of the steel sheets was 1.4 mm, and the wire diameter of the Si-bronze filler was 0.9 mm. The chemical composition of the substrates with details of their coating and filler wire is given in Table 3.1, which is identical to what was used for the GMAB study reported in [101].

Table 3.1: Chemical composition (mass%) of the substrate, coating, and filler wire

Substrate	Fe	Cu	Zn	Si	Mn	Al	C	Traces (Cr, Ti, B, Nb, Ni, Sn, Mo, V, P, S, N)
GI DP600	Bal.	0.130	-	0.160	1.930	0.030	0.094	0.143
GA DP600	Bal.	0.032	-	0.150	1.986	0.025	0.091	0.113
Coating								
GI	0.49	-	Bal.	-	-	0.25	-	-
GA	8.30	-	Bal.	-	-	0.20	-	-
Filler wire								
CuSi3Mn1 (Si-bronze)	0.02	Bal.	-	2.90	0.80	-	-	-

3.3.2 LWB Process

An IPG Ytterbium-doped fiber laser with a maximum power capacity of 10 kW and a laser beam wavelength of 1070 ± 10 nm was used to braze the steel substrates in the BOP configuration, as indicated in Fig. 3.2. The processing parameters for LWB were optimized to produce an ideal braze bead in terms of bead appearance and elimination of any internal defects, such as porosity, lack of adhesion and cracking. The methodology was selected to make the results relatively comparable in the same qualitative manner as the previous study that used GMAB in Ref. [101]. The LWB parameters were selected while adhering to the constraint of using a nominal heat input ($HI_{Nominal} = \frac{P}{TS}$) that closely matched the heat input ($\sim 80 - 86$ J/mm) previously used in GMAB. LWB was conducted with an output power of 3 kW with a nominal spot size diameter of 0.6 mm and TS of 2 m/min, providing a 90 J/mm $HI_{Nominal}$. To be comparable to GMAB that the heat is directly generated from the electrode filler wire, the laser spot size was chosen to be equal to the filler wire diameter in LWB. The beam was defocused 12 mm from the focal point to increase the laser beam spot size diameter

to 0.9 mm on the workpiece. The wire was fed at a horizontal angle of 45° relative to the substrate using a leading wire configuration, with the wire being fed from the direction of laser travel at a *WFS* of 2.5 m/min. LWB was undertaken using a coaxial configuration to deliver argon gas shielding at a rate of 20 l/min. It is important to note that the processing parameters were selected based on what is currently being employed in the industry based on recommendations from industry partners.

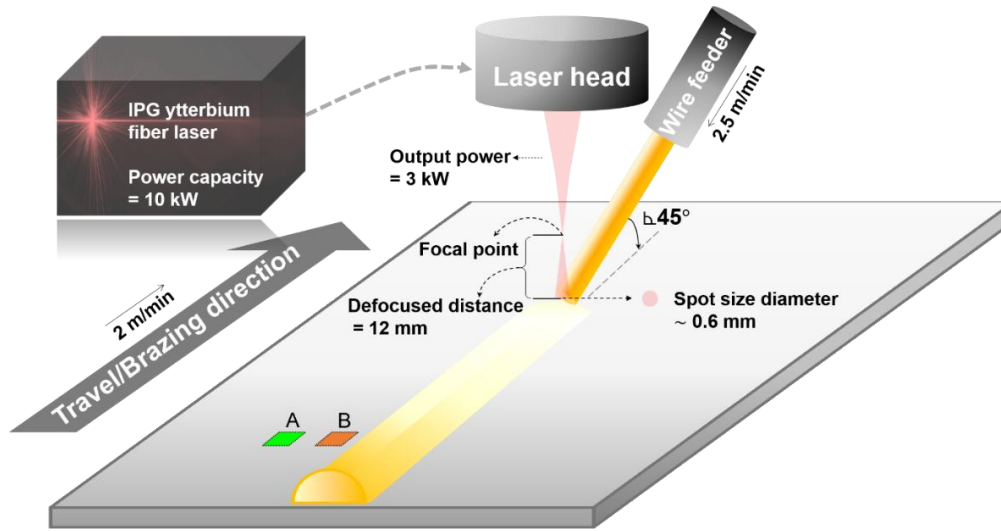


Fig. 3.2: Schematic of LWB in BoP configuration along with the processing parameters. “A” and “B” indicate regions of interest for surface roughness analysis.

3.3.3 Microstructural Characterizations

To reveal the microstructure, the laser-brazed samples were cross-sectioned perpendicular to the brazing direction. A highly controlled polishing procedure was developed to attain a scratch-free surface on the examined area using four stages of automated grinding and polishing on a Struers Tegramin-20 system with commercially available consumables [162] as detailed in Table 2. It is noteworthy that to ensure every part of the edge at the brazing bead was revealed during the polishing steps, a maximum polishing force of 5 kN was used, 20% of the typical force used to polish welds in DP600. To compensate for the reduction in polishing force, the polishing time (as shown in Table 3.2) was increased for each stage compared to the GMAB sample preparation reported in [101]. To observe the HAZ using an

optical microscope (OM), the polished surfaces were etched with two etchants, specifically ferric chloride (120 ml C₂H₅OH (80%) + 30 ml HCL (20%) + 10 g FeCl₃) and a 2% Nital solution in two rounds of etching for 20, and 8 seconds, respectively, as shown in Table 3.3. Geometrical features of the brazing bead were measured on four cross-sections and averaged for statistical relevance and accuracy. Error bars in the resulting graphs represent deviations from the mean value of the measurements, based on 95% confidence intervals.

Table 3.2 Metallography sample preparation parameters

Polishing components	Stage 1	Stage 2	Stage 3	Stage 4
Disc	MD-Piano 220	MD-Allegro	MD-DAC	MD-NAP
Suspension	Water	Diapro Allegro 9 µm	Monocrystalline diamond 3 µm	Monocrystalline diamond 1 µm
Speed (RPM)	300	150	150	150
Force (kN)	5	5	5	5
Time (min)	10	5	15	15

Table 3.3. Procedure of etching the metallography samples

Etching stages	Round 1	Cleaning	Round 2	Cleaning
Solution	Ferric chloride	Ethanol	2% Nital	Ethanol
Time (s)	~20	-	~8	-

Further investigation of the microstructure on the brazing bead edge was conducted on the above-mentioned highly controlled polished metallographic samples without etching the surface to avoid the occurrence of corrosion in this area. To distinguish the edge morphology and identify the differences in chemical compositions, microstructural characterization was conducted using electron microscopy with backscattered imaging (BSI) and energy-dispersive (EDS) detectors. A Zeiss LEO 1530 SEM equipped with a field emission gun was used for this study. The selected voltage, beam current, and aperture size were 20 kV, 2.46 A, and 60, respectively. Elements of Fe, Cu, Zn, Si, Mn, and Al were chosen for EDS point and line profile scans with a scanning time of 10 minutes for each zone.

Microstructural characterization of the geometry was complemented with surface profilometry. Surface morphology and roughness of the substrate surfaces were measured far from the bead edge (region “A” in Fig. 3.1) and at the region adjacent to the bead edge (region “B” in Fig. 3.1) using a Keyence VK-X250 3D confocal laser scanning microscope (CLSM). The surface roughness is generally defined by an absolute value, S_a , the arithmetic mean distance of the peak height to the surface [163–165].

3.4 Results and Discussion

3.4.1 Influence of Heat Source and Coating Types on Wettability

The laser-brazed beads on GI-coated and GA-coated DP steel substrates can be seen in the stereoscope images, shown in Fig. 3.3a and b, compared with arc-brazed beads from previous work in Fig. 3.3c and d [101]. Two noticeable differences between LWB and GMAB were the uniformity of the brazing bead, as well as the significant reduction in spatter. Almost no spatter was observed around the laser-brazed bead in both cases of GI-coated and GA-coated substrates. From the heat source point of view, this was likely due to the consistency of heat transfer through LWB, which enabled a more stable fluid flow at the wetting surface compared to GMAB. In fact, during an arc process, the power fluctuates instantaneously due to changes in the arc/wire interaction as the wire melts [166].

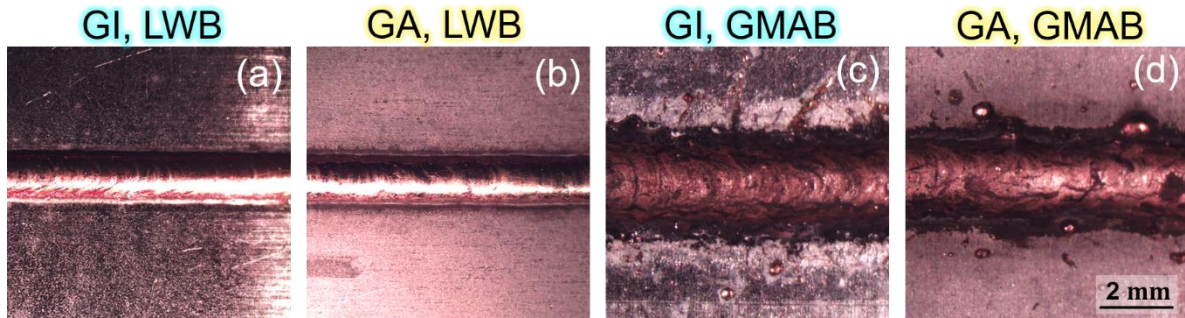


Fig. 3.3: Top view of the LWB bead in BOP configuration on (a) GI- and (b) GA-coated DP600 steel. (c) and (d) are depicted from GMAB work reported in [101].

As mentioned in Section 3.3.2, the brazing beads were manufactured to have a fully-optimized bead appearance by both LWB and GMAB. It is important to note that using a laser as the heat source allowed for increases in travel speed to 2 m/min without decreasing braze quality, while the brazing travel speed was 1.02 m/min in GMAB. As a result, the LWB processing time was less than half the GMAB time; also, less heat per unit length of the laser braze was transferred to the substrate, which in turn reduced damage to the substrate surface. The decreased microstructural damage may be seen from an examination of the braze cross-section, shown in Fig. 3.4. The LWB results showed complete bonding at the interface, without any sign of substrate melting, as desired in brazing. Whereas, in the case of GMAB [101], there was a considerable partially melted zone (PMZ) in the substrate beneath the interface, measured at approximately 0.6 mm^2 . Although the substrate surface may have melted in LWB as well, the PMZ was negligible. The region between the brazing interface and the orange dashed line in Fig. 3.4 is the HAZ revealed by etching the samples. Unlike the PMZ, the HAZ microstructure did not consist of severe phase changes as there was no chemical composition gradient in the corresponding BSI maps of the HAZ (Fig. 3.4b and d).

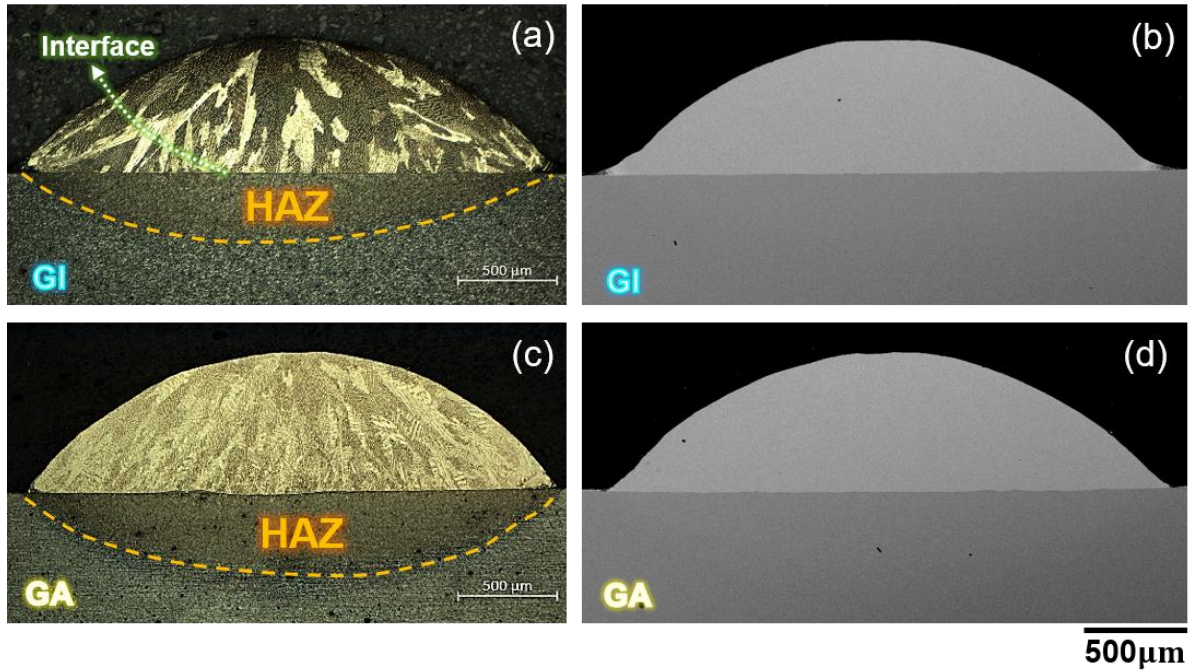


Fig. 3.4: Cross-sectional view of the LWB samples in etched and unetched conditions captured by OM and SEM, respectively, (a and b) GI-and (c and d) GA-coated.

To show how the brazing bead geometry was affected by the laser, geometrical features including contact angle, bead width, and HAZ depth of the brazing bead on GI- and GA-coated substrates were measured from the etched microstructures and compared with the GMAB samples studied in [101], as illustrated in Fig. 3.5 and tabulated in Table 3.4. As may be seen in Fig. 3.4b, the bead width and bead cross-sectional area for LWB were reduced by 40% and 70%, respectively, compared to GMAB, indicating that the brazing bead was narrower in LWB. The contact angle of the laser-brazed samples was 25% smaller than the arc-brazed bead (as seen in Fig. 3.5c). As the contact angle may be considered a representative measure of wettability, it can be confirmed that the wettability of both cases in LWB, GI- and GA-coated substrates, was improved by a reduction in the contact angle.

LWB and GMAB processes can also be compared in terms of their corresponding interaction volumes. For LWB, the interaction volume consists of 1) the laser beam, 2) the non-melted wire, 3) the solid substrate, and 4) the melt pool. In the case of GMAB, the

interaction volume includes 1) the electric arc + plasma, 2) the non-melted electrode and/or filler wire, 3) the solid substrate, and 4) the melt pool. Among all components in the interaction volume, the 1st (heat source) component has a major impact to shape the bead geometry. The bead width and bead area are both functions of input energy and heat density (derived from the 1st component which is different in LWB and GMAB) required to melt the brazing wire (2nd component) and spread at the surface of the solid substrate (3rd component). The 2nd component, brazing wire is a function of wire chemical composition, wire diameter, and wire feeding rate. The effect of 3rd component, solid substrate, is manifested by different types of coating in this work, GI- and GA-coatings. After the filler wire melted it must spread on the solid substrate surface. The coating which is a function of surface morphology and surface chemistry should interact with the first two components. Eventually, the interaction between the first three components results in the creation of the melt pool, 4th component.

To realize how the 1st component impacted the bead geometry, the heat generation during each process, LWB and GMAB can be compared considering a constant 3rd component. In LWB, the heat is delivered by irradiation of the laser beam on the filler wire and substrate. In GMAB, however, the heat is generated by transition of an electric current into the filler wire electrode in which the electric arc and its surrounding plasma can be excited once the electrode touches the workpiece or substrate. In fact, in GMAB as reported in [101], the heat input is generated from the filler wire which act as electrode, while, in LWB, the heat is sourced from a separate component. Owing to the higher concentration of laser beam heat delivery and higher heat density, a laser beam with the spot size of as small as 0.9 mm which is equal to the wire

diameter at a double brazing speed could provide a sufficient heat input for brazing. As a result, in LWB, a smaller area of the substrate surface was affected by heat compared to arc + plasma forming at the periphery of the molten wire electrode in GMAB. Therefore, in LWB, a narrower melt pool (4th component) could be formed which directly resulted in 40% smaller bead width.

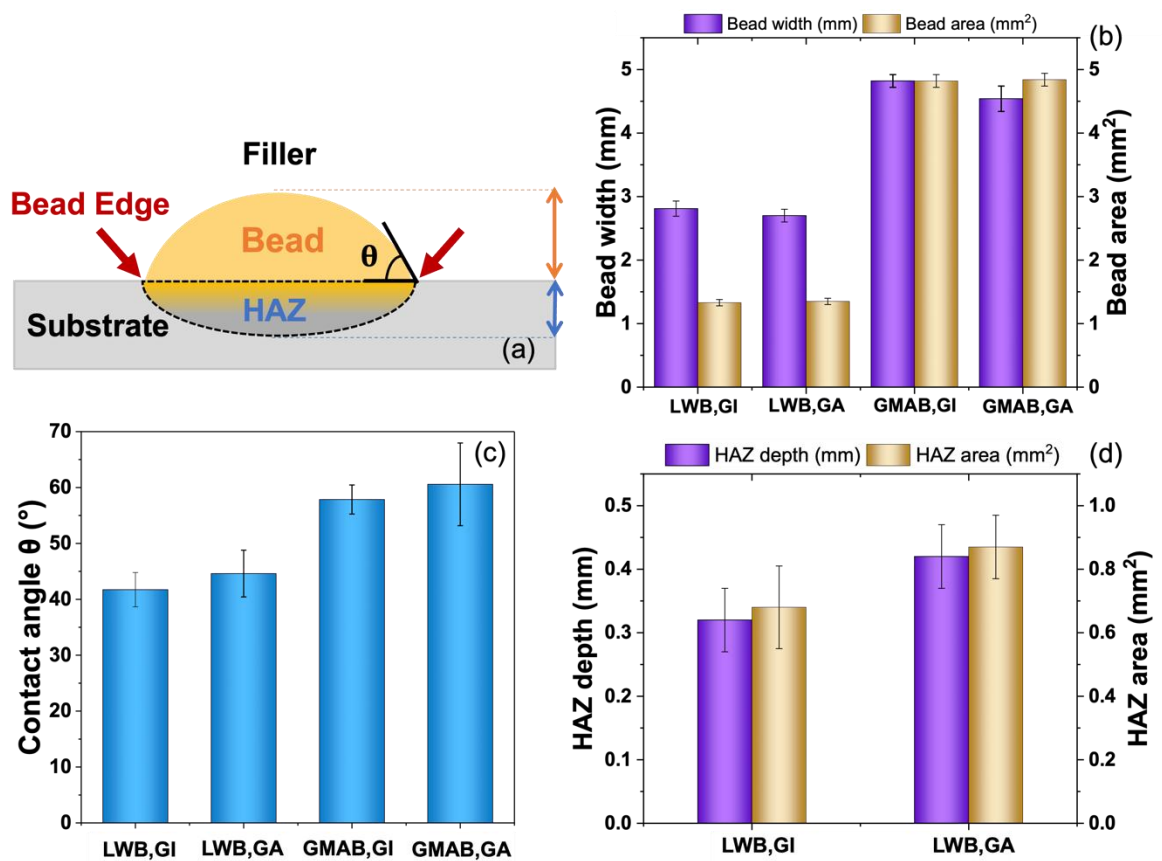


Fig. 3.5: (a) A schematic representation of the brazing bead cross-section, and geometrical measurements of the brazing bead including (b) average contact angle, (c) bead width and area for the LWB GI- and GA-coated samples in the current work relative to GMAB data from previous work [101], and (d) HAZ depth and area of LWB samples. The GMAB data does not have error bars as none was provided in the original reference.

Table 3.4. Summarized geometrical measurement for the brazing bead of LWB samples in the present work in a comparison with Ref. [101].

Samples	Contact angle θ (°)	Bead width (mm)	Bead area (mm ²)	HAZ depth (mm)	HAZ area (mm ²)	PMZ depth (mm)	PMZ area (mm ²)
LWB_GI	41.7 ± 3.1	2.8 ± 0.1	1.3 ± 0.1	0.3 ± 0.1	0.7 ± 0.1	~0	~0
LWB_GA	44.6 ± 4.2	2.7 ± 0.1	1.3 ± 0.1	0.4 ± 0.1	0.9 ± 0.1	~0	~0
GMAB_GI [101]	57.9 ± 2.6	4.82 ± 0.1	4.82 ± 0.1	N/A	N/A	0.33	0.55
GMAB_GA [101]	60.6 ± 7.4	4.54 ± 0.2	4.84 ± 0.1	N/A	N/A	0.35	0.58

The 3rd component (solid substrate) is a function of surface morphology and chemistry of the coating. Comparing bead width of the laser-brazed GI- and GA-coated substrates (according to Table 3.4), the difference is only 0.1 mm. While comparing bead width of GI and GA in LWB with GMAB, the difference is almost double. It shows that the influence of 1st component (heat source) on bead width is stronger than 3rd component. Comparing the contact angles of GI and GA samples in LWB, in Table 3.4, GA-coated substrate had a larger contact angle (~7%) than the GI-coated substrate indicating the effect of 3rd component in wettability, which was observed in GMAB results as well. In LWB, a smaller HAZ area (~28%) and shallower HAZ depth (~25%) in the GI sample relative to the GA sample may be an indication that lower heat was absorbed by the GI substrate; this linkage between HAZ depth and heat absorption by the substrate has been observed previously for GMAB in [160] and LWB in [133,167] and attributed to surface roughness. This observation agrees with studies showing that surface roughness and heat absorption have a direct relationship [133,167].

As the surface morphology of the substrates can influence the interaction volume of LWB, the roughness characteristics of GI- and GA-coated substrates were also examined in the present

study to understand its effect on the wettability of the molten filler material within the interaction volume in LWB (interaction between 1st and 3rd components). Using 3D topographical mapping, the surface morphology was characterized as shown in Fig. 5 for two regions: “A” (far from bead edge, i.e., coated substrate) and “B” (adjacent to bead edge, i.e., heat-affected surface) as given in Fig. 3.2. The Sa values on each map were the evaluation results of the surface roughness. The surface roughness of the GI-coating was less than the GA-coating, $Sa_{GI} = 0.402 \mu m$ and $Sa_{GA} = 1.316 \mu m$, respectively (Fig. 3.6a and c). Khan et. al. [101] proved that lower surface roughness can improve wettability by increasing the surface tension or SFE of a solid substrate which increases the contact area between the substrate and the molten droplet in a sessile drop test experiment. This correlates well with the better wettability of the GI-coated substrate with lower surface roughness compared to the GA-coated substrate in LWB as well.

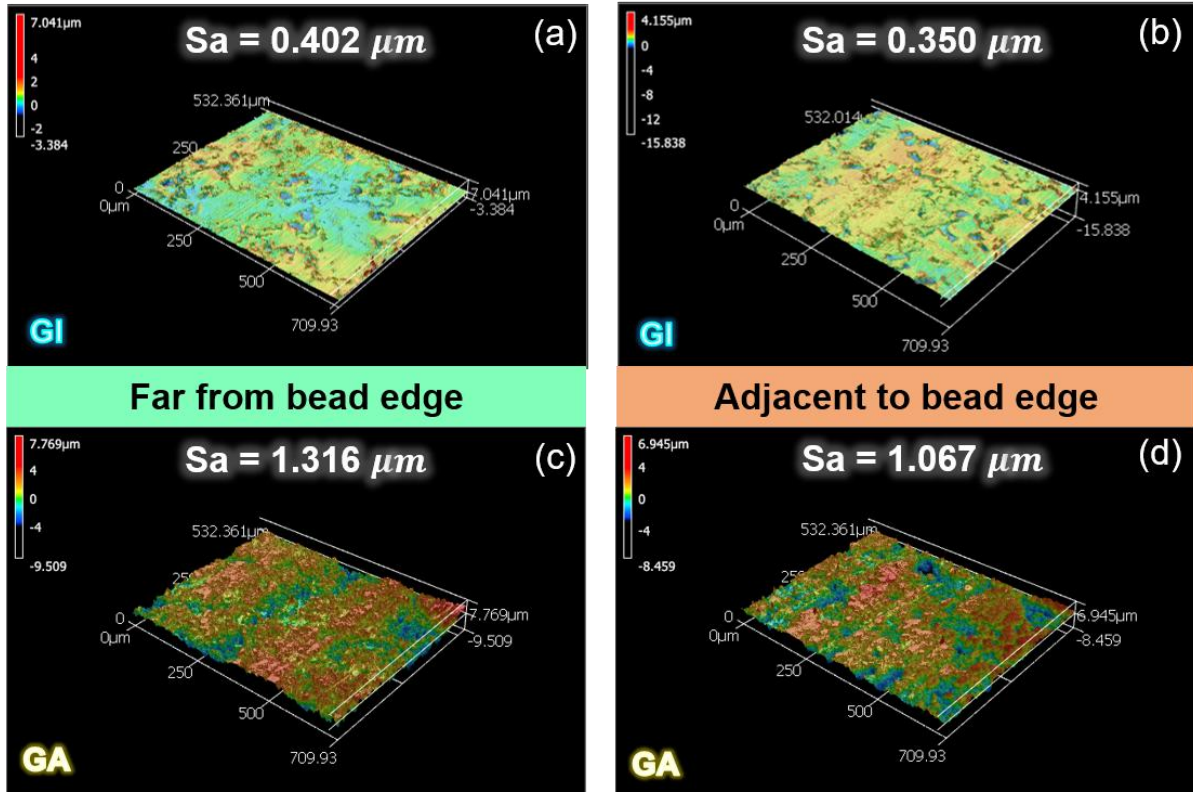


Fig. 3.6: 3D maps of the surface roughness characterized by CLSM for (a) GI and (c) GA coatings far from the bead edge (region “A” in Fig. 1), and (b) GI and (d) GA coatings adjacent to the LWB bead edge (region “B” in Fig. 1).

It is important to note that differences in roughness were not the only contributing factor, since the interaction of these substrates with the laser (i.e., laser-solid/liquid substrate interaction) needs to be considered as well. Comparing the effect of LWB on GI- and GA-coated substrates, the S_a measured at the bead edge was $0.350 \mu m$ and $1.067 \mu m$ for GI- and GA-coated samples, respectively (Fig. 3.6b and d). Although the change in roughness (i.e. from region A to B) was different for both substrates, the roughness in both cases decreased. The results showed that heating of the substrate near the interaction area smoothed the surface morphology of both coatings. Considering the laser beam was Gaussian with the peak temperature at the centre, it is reasonable to assume that the heat at the periphery of the interaction volume can still have significant effects on the region close to the edge as supported by previous findings in [168,169]. The local increase in temperature at the interface between the molten wire droplet

and the substrate was sufficient to alter the surface morphology. This was attributed to the heat transfer by laser radiation, beam reflection, and heat conduction being absorbed by the substrate. One desired consequence of the morphology alteration was the reduction in the contact angle of the brazing bead of the LWB samples compared to GMAB samples, as an indication of improved wettability of the filler wire. The interaction of the laser with the material at the surface, similar to other laser surface modification treatment techniques reported in the literature [143,170,171], was effective in improving surface morphology.

To further understand the effect of coating type on roughness evolution due to surface morphology alternation by laser, “smoothing ratio” can be defined as $Sa_{\text{adjacent to bead edge}} / Sa_{\text{coated substrate}}$. The smaller the ratio, the greater the effective decrease in surface roughness for the same processing parameters. The results showed that in the case of the GI-coated substrate, this ratio was 0.87, while for the GA-coated substrate, this value was 0.81. This means that the GA-coating had a higher propensity to improve its surface roughness compared to the GI-coating, although its overall roughness was much greater. For the GI-coating, the surface was already smooth; this led to a lower capability for further smoothing. Another point of consideration is that the effectiveness of heat transfer by the laser was diminished when interacting with such a smooth surface with high reflectivity. Surface reflectivity has an inversely proportional correlation to surface roughness. Theoretically, if the number of surface imperfections is smaller than the wavelength of the incident light, almost complete reflection can be expected from the surface as reported in [172]. By assuming that the surface roughness is a measure of surface imperfection, almost complete reflection can be expected from the surface under irradiation of the laser beam with a wavelength 1070 ± 10 nm, which was used in this study. To this effect, the 25% shallower HAZ in the GI-coated sample confirmed lower heat absorption due to higher reflectivity, which resulted in a less pronounced effect on surface morphology. Meanwhile, it can be realized that there were two effects arising from the substrate as a component contributing to the LWB interaction volume. The surface roughness of the substrate impacted the heat conduction within the interaction volume. The surface reflectivity of the substrate also impacted the heat absorption capacity of the substrate.

Recognizing the contribution of each one needs further investigation, which is a subject for ongoing study.

3.4.2 Importance of Brazing Bead Edge Geometry

High magnification imaging of the unetched cross-section was required to understand the importance of brazing bead edge geometry in determining wetting behaviour. The GI and GA samples were analyzed using an SEM in BSI mode, as depicted in Fig. 3.7 and 3.8, respectively. Firstly, it was seen that a tail-like region formed at the edge of the brazing bead on both the GI- and GA-coated substrates. This is referred to as the occurrence of a “tailing-out” effect, which is the association of the apparent contact angle with the flow of the brazing bead to create a tail at its edge. Fig. 3.9 demonstrates the difference between the apparent contact angle and the contact angle at the tail region, called “tailing angle”. This geometrical parameter was defined to distinguish between the apparent contact angle and the contact angle at the tail. The tailing-out formation occurs with a distinct mechanism that is different from the common spreading of the molten filler wire on the substrate. Although the tailing angle for LWB on the GI samples is 50% that of the tailing angle of the LWB on the GA sample, the tailing angles of both samples were smaller than their apparent contact angles. In comparison, the tailing-out effect for GMAB was only observed in the GI-coated substrate [101]. In fact, the tailing-out feature on the brazing bead edge represented higher wettability, while the brazing bead edge without a tail refers to a substrate with lower wettability [173]. Secondly, from the OM images and SEM-BSI micrographs, a sharp colour contrast existed between the tail region and the braze bead, accompanied by a gradual colour contrast from the tail interface (blue dashed line in Fig. 3.7b and 3.8b) to the braze bead. These colour contrasts confirmed the occurrence of unique elemental transportation reactions due to the chemical segregation in this region.

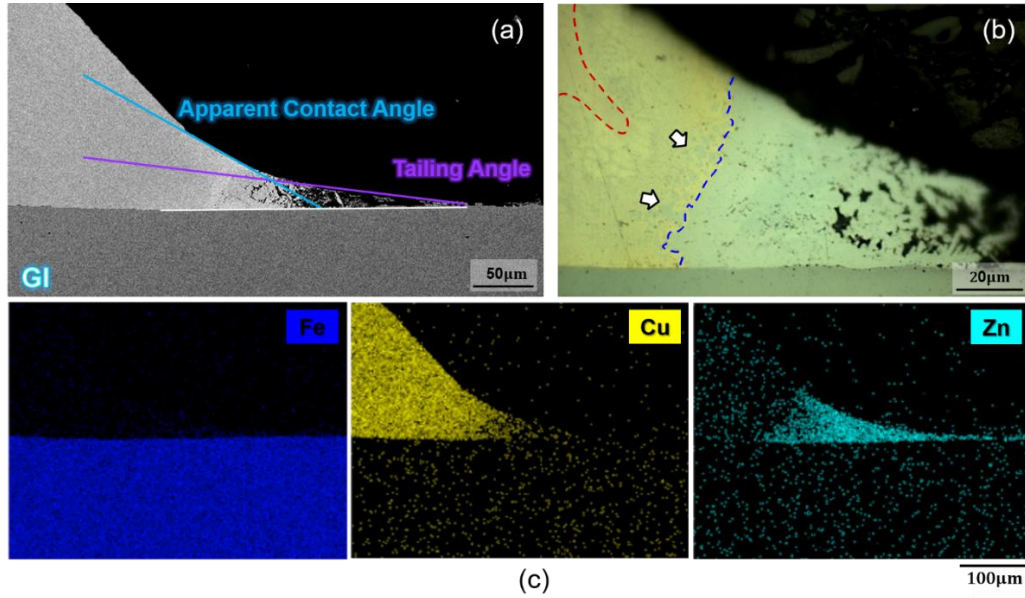


Fig. 3.7: Brazing bead edge microstructure in as-polished GI sample captured by (a) SEM and (b) OM along with the corresponding (c) EDS maps of Fe, Cu, and Zn elements.

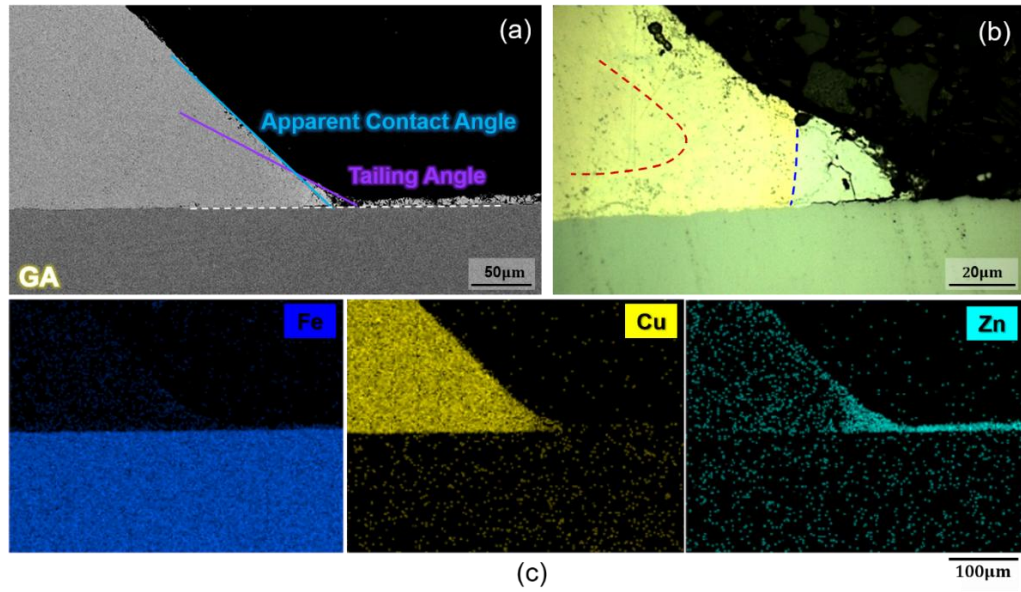


Fig. 3.8: Brazing bead edge microstructure in as-polished GA sample captured by (a) SEM and (b) OM along with the corresponding (c) EDS maps of Fe, Cu, and Zn elements.

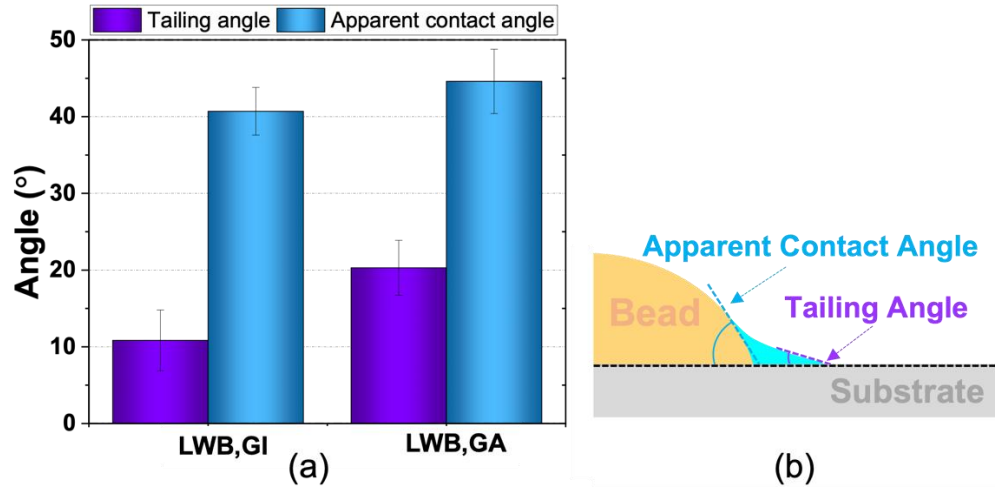


Fig. 3.9: Graphing the apparent contact angle and tailing angle for GI and GA coated samples, referring to the schematic in (b) (Color figure online).

According to the elemental distribution maps, shown in Fig. 3.7c and 3.8c, the Zn concentration was higher than Cu at the tail region in both cases of the GI- and GA-coated substrates. Not only the Zn but also the Cu from the matrix of the brazing bead and Fe from the matrix of the substrate were detected in the bead edge developing a Zn-Cu alloy at the tail. The formation of this Zn-rich tail is a phenomenon, and it can be attributed to various factors. Considering the lower heat input at the periphery of the braze bead, the rate of Zn vaporization was reduced at the bead edge as compared to the bead centre along the interface. The high affinity of Zn to form an alloy with Cu in brazing leads to Zn-rich tail formation as reported previously in [136,158]. Another reason for Zn accumulation at the brazing bead edge can be the turbulent movement of the melt pool and fume emissions, which might lead to pushing the Zn towards the edge as related in [44,133]. Previously, a high concentration of Zn at the tail was only seen for GMAB of GI-coated samples [101]. In the present work, there was a correlation in the Zn distribution at the tail region in the cases of both the GI- and GA-coated substrates of the LWB samples, which is in contrast to GMAB findings for the GI-coated substrate in [101] that indicated Zn accumulation only within the tail region. In GMAB of the GA-coated substrate with no tail region [101], however, Zn was distributed on both sides of the wetting interface, close to the bead edge and within the substrate, showing that Zn further

diffused into the substrate from the coating. The difference in the Zn distribution of samples with and without the tail region can be related to the lower heat input in LWB compared to GMAB. A lower amount of Zn vaporizes at the braze edge of LWB samples, which affects interfacial reactions at the edge of the brazing bead to promote tail formation [49]. Although the Zn distribution of GI- and GA-coated substrates in LWB both demonstrate the same pattern resulting in the tail formation and improvement in wettability, the size of the tail region was smaller in the GA-coated substrate. Considering the roughness of the GA-coating was about 2.3 times higher than the GI-coating, higher heat absorption and Zn vaporization were more expected in the GA- than in the GI-coated substrates, decreasing the tail size. The next section describes how interfacial reactions leading to tail formation were affected by interaction volume components in LWB.

3.4.3 Segregation Behaviour at Brazing Bead Edge

The melt pool, 4th component, which is a function of the chemistry of the filler wire and substrate coating as well as the available heat in the interaction volume, impacts the segregation behavior at the brazing interface. The elemental concentration of each element found in the tail region of the GI- and GA-coated brazing bead edges were analyzed to understand how the segregation behaviour of elements in this tiny region correlated with the wetting process. Noteworthy, although by following the highly controlled polishing procedure designed for the LWB samples (according to Section 3.3.3) the braze edge could withstand the forces during sample prep, still local breakage was occurred at the bead edge. Therefore, some parts of the edge were not revealed leading to the appearance of rough area as an unavoidable artifact from mechanical polishing; it doesn't affect the accuracy of the results. Using EDS data acquired from selected points/spots on the tail region of the GI sample, Zn, Cu, and Fe were revealed as the main elements found at the tail (Fig. 3.10a and b). From points 1 to 5, the atomic fraction of Cu dropped from 26.41 to 4.91 (at.%), due to the diffusion of Cu from the molten filler wire into the tail region; by contrast, Zn and Fe increased by +11.63 and +9.29 (at.%) to reach to 83.00 and 10.16 (at.%), respectively, which points to these elements flowing from the substrate to the braze bead – that was mostly composed of Cu and eventually led to tailing out formation.

In fact, it can be realized that the apparent contact angle was indicative of the filler metal flow on the substrate with Cu as a main content, while the tailing out occurred as a result of the formation of a build-up containing Zn and Fe atoms from the substrate towards the braze at the edge.

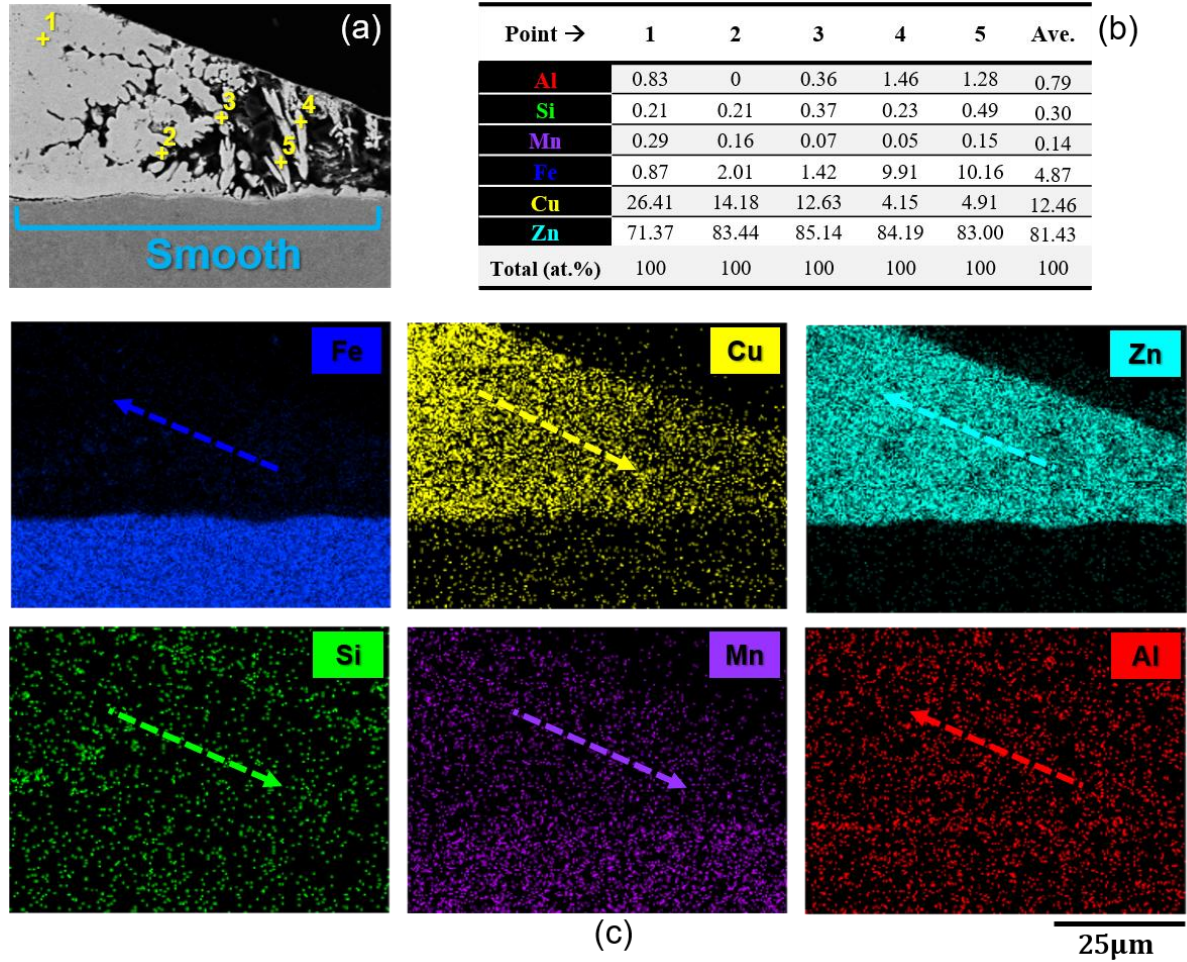


Fig. 3.10: (a) High magnification SEM BSI micrographs of the tail region of the GI-coated sample at the LWB bead edge, (b) the atomic fraction of substantial elements from point analysis on (a), and (c) corresponding EDS maps. Dashed arrows show the expected diffusion direction of each element (Color figure online).

A similar trend was observed in the GA-coated sample (Fig. 3.11a and b), but the atomic fraction of each element differed from that of the GI-coated sample; specifically, in the GA-coated sample, Fe was +3.13 at.% higher near the end of the tail and Zn was slightly lower (-

0.38 at.%) at point 7. Arrows on the EDS maps in Fig. 3.10c and Fig. 3.11c represent the expected diffusion direction of the elements, which migrated from a region of high concentration to low concentration based on the initial chemical composition of filler wire and substrate and the final segregation profile. To better compare the diffusion behaviour of the GI- and GA-coated samples at the brazing bead edge, the atomic fraction of each point was averaged for points 1 to 5 in GI and 3 to 7 in GA, along the tail for each element (in Fig. 3.10b and Fig. 3.11b, respectively). The tail of the brazing bead edge contained higher fractions of Fe, Cu, Si, and Mn in the GA-coated sample, while Zn and Al were higher in the case of the GI-coated sample. As for Fe and Zn, the difference between the GI- and GA-coated samples was connected to the initial composition of these coatings and Si-bronze filler wire (Table 1). The filler and initial substrate in both cases were similar, while the chemical composition of the coating was different. The GI-coating had 0.49 at.% Fe and 99.27 at.% Zn, while the GA-coating had a much higher amount of Fe (8.3 at.%) with significantly lower Zn (91.5 at.%), which is the main reason for the +3.13 at.% higher Fe content and -0.38 at.% lower Zn content in the tail of the brazing bead edge in the GA-coated sample. As for the remaining elements of interest, including Cu, Si, and Mn, the analysis was relatively more complicated. Although the pattern of elemental distributions for Zn and Cu within the tail was similar for the GI- and GA-coatings, the results showed that the distribution of Fe and Si elements had formed a curve-shaped build-up in the GA-coated sample, which was different from the GI sample.

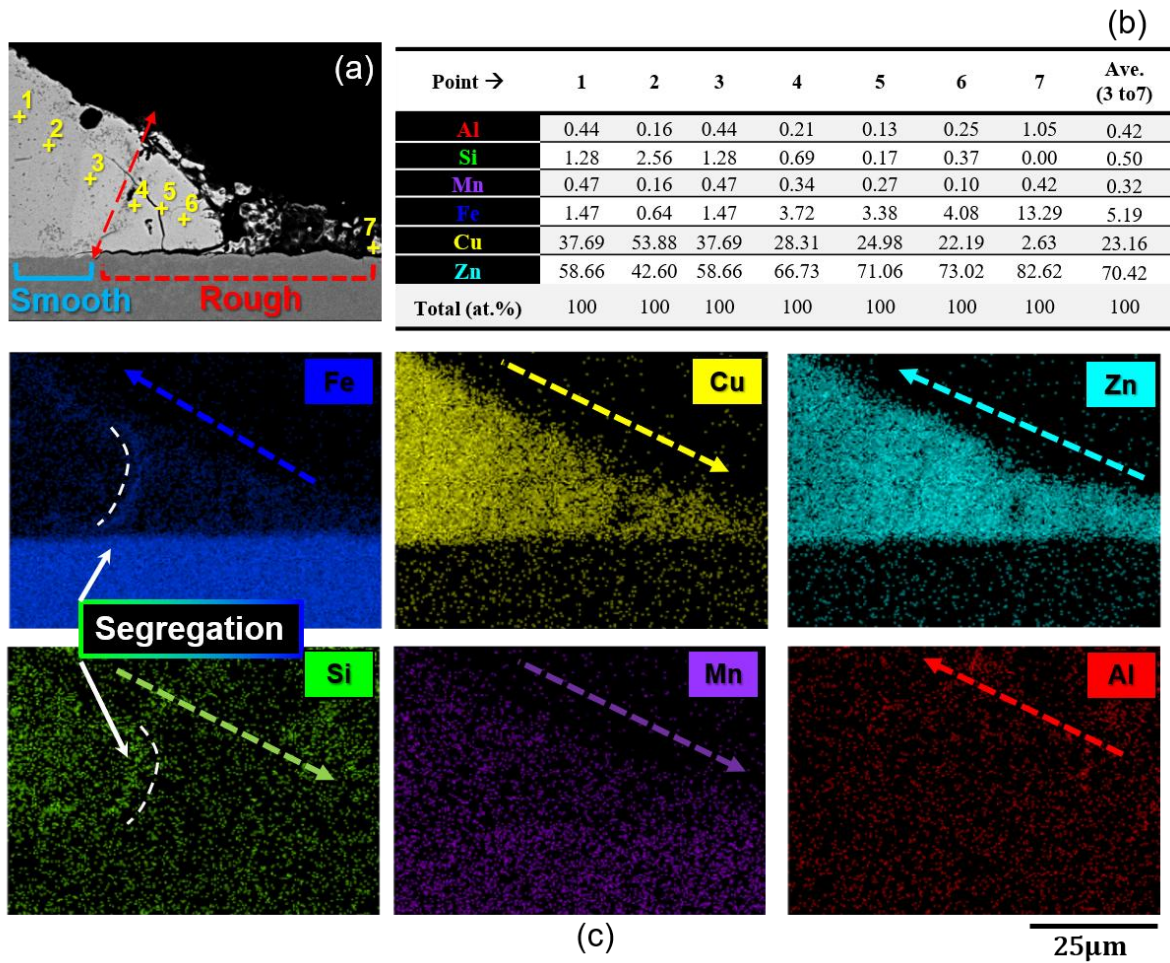


Fig. 3.11: (a) High magnification SEM BSI micrographs of the tail region of the GA-coated sample at the LWB bead edge, (b) the atomic fraction of substantial elements from point analysis on (a), and (c) corresponding EDS maps. Dashed arrows show the diffusion direction of each element. White curve-shaped dashed arrows on Fe and Si maps indicate elemental segregation. Two-sided dashed red arrow in (a) indicates the hypothetical boundary where segregation happened (Color figure online).

In the line scan taken from the segregation region as delineated by the dotted line in Fig. 3.12, it can be observed that the Fe and Si elements had a sharp peak in their concentrations (dark blue and green lines, respectively) visible in the curve-shaped region, indicating segregation of Fe and Si elements. It was also notable that the Cu and Zn contents had elemental drops within this region.

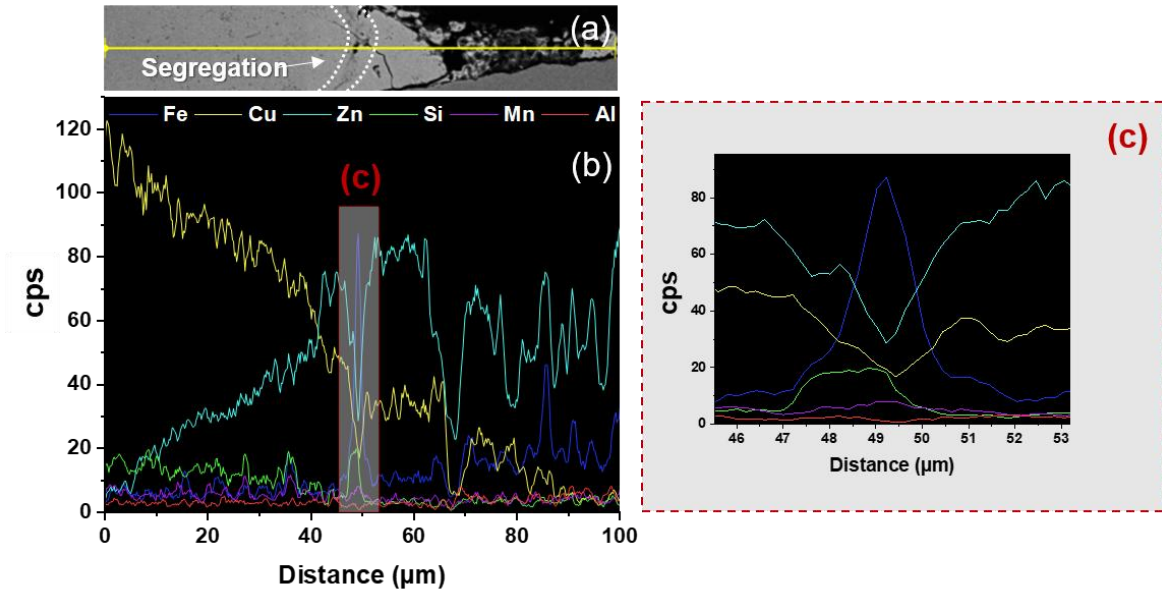


Fig. 3.12: EDS line scanning on the brazing bead edge region of the GA-coated sample along the (a) yellow line, (b) line profiles showing elemental change along the 100 μm length, and (c) magnified region corresponding to the rectangle in (b) (Color figure online).

Fe and Si segregation near the tail region can be explained as follows: 1) Si had the highest concentration at the interface of the substrate and wire, as it was available from both the substrate and the filler. Fe also existed in the vicinity of the GA coat and substrate. 2) Compared to the other elements present, Fe and Si were in the highest abundance, showing the driving force at the curve-shape region was sufficient to thermodynamically stabilize these elements leading to the appearance of upward peak in the elemental profile as opposed to Cu and Zn at this region as supported by [174–176]. The reason why this segregation behaviour was only observed in the case of the GA-coated sample during LWB can be identified by comparing the tail/substrate interface between the two samples. This region was continuously smooth in the GI-coated sample (blue dashed line labelled as “Smooth” in Fig. 3.9a), while in the GA-coated sample, there was a rough surface beneath the segregated region at the interface close to the end of the brazing bead edge (red dashed line labelled as “Rough” in Fig. 3.11a). In the GI-coated substrate, due to the lower surface roughness, the molten braze metal wet the surface of the coating easily without any restriction to elemental migration. Whereas in GA-

coated substrate, the flow of material was restricted beyond a hypothetical boundary (red two-sided arrow in Fig. 3.11a); so, in the elemental maps of Fe and Si there was a curvature.

Aside from the observations concerning the interface between the braze and underlying steel, it should also be considered that segregation, in the case of the GA-coated substrate, can be process-related during LWB. Using the laser instead of an arc had an influence on the cooling rate after the brazing process. Due to the smaller size of brazing bead by LWB compared to GMAB, the rate of cooling right after the process was much higher. Meanwhile, the GA-coating with 2.3 times higher roughness compared to the GI-coating, which was a potential heat absorber in the interaction volume of LWB, can experience a higher diffusion rate at the interface of the brazing bead and the substrate. On one hand, the combination of the high cooling rate in the LWB process and the high diffusion rate of Fe and Si at the interface of the GA-coated substrate and braze bead resulted in a localized contrast in temperature distribution stabilizing Fe and Si more than Cu and Zn. On the other hand, the presence of the remained rough interface might have further stimulated this favourable condition for Fe and Si segregation by restricting the homogenous flow of the molten metal. Thus, the temperature gradient from the state of the molten pool and a solidified braze bead can act as a driving force of segregation in this case. As compared to the GI-coated substrate in LWB, due to the smaller temperature gradient, the driving force was not enough to cause segregation.

3.5 Summary

This work presented a detailed study on the wettability of a Si-bronze filler wire on GI- and GA-coated DP600 steel using the LWB process in BOP configuration. The brazing bead geometry and diffusion behaviour at the edge region were investigated and compared with a correlated work from the literature on GMAB [101]. The key findings of this study can be summarized as follows:

1. LWB, compared to GMAB, produced a more uniform brazing bead with a reduction in spattering due to higher heat production consistency and fluid dynamic stability, which allowed doubling the travel speed and resulted in less heat absorption by the substrate.

2. By LWB, domination of the HAZ instead of the PMZ and about 25% reduction in the contact angle of the brazing bead were achieved. Wettability in LWB was improved by tail formation due to Zn accumulation from the GI- and GA-coatings into the brazing bead edge that could only occur in the GI-coated substrate for GMAB. In LWB, although the GI-coated sample compared to the GA-coated sample showed a slightly smaller (25%) wetting angle, the tailing angle was 50% smaller in GI.
3. Laser-solid/liquid substrate interaction and subsequent heat conduction within the solid substrate and filler metal, the molten filler metal, and the melt pool were found as the major components contributing to the difference in the wetting process during LWB compared with GMAB.
4. The GA-coated substrate with higher surface roughness showed a higher capacity to smoothen when interacting with laser heat conduction due to surface modification via phase transformation. So, the negative effect of the rough surface on wettability could be diminished by the laser. The smoothening capacity for the GI-coated substrate was found to be less than the GA-coated substrate since the former was already adequately smooth to wet the surface.
5. The smaller size of the tail formed on the GA-coated substrate was correlated with the segregation of Fe and Si, the two most thermodynamically stable elements in this joint area, close to the edge of brazing bead. The main reason for restricting further flow of the molten wire on the substrate was the existence of a rough surface and a higher temperature gradient of this condition right after the process.

Chapter 4. AI-friendly Optimization of Robotic Laser Weld Brazing Process Advancing Industry 4.0 Real-time Inspection³

4.1 Overview

This study establishes a custom-designed process window for a robotic LWB process to control internal defects in braze seams used in automotive applications. This window is essential for producing braze seams that meet custom specifications, enabling the development of an industrial-level UT defect detection system. Supported by AI, this system automates post-processing of UT data and measurement extraction. In order to advance such developments, a novel wire-adjusted heat input approach was defined that expounded clear correlation between the heat input and defect formation, unlike the traditional nominal heat input method, which lacked consistency under varying wire feed speeds. With this novel approach, the heat input-related boundaries could be delimited to enable LWB optimization and distinguish defect-free seams from those exhibiting defects, including LoA, pores, BMM, and non-conforming joint geometry. An optimal process window was validated through an unprecedented application of full-scale 2D and 3D X-ray μ CT to scan and analyze the braze seams for the pore distribution using a semi-automated procedure for pore segmentation, which facilitated statistical analysis of the pore evolution within the braze. This approach ensured reliable classification of the defect conditions within the established AI-friendly process window. The UT response to both the defect conditions and seam geometry in the brazes produced in this process window was also evaluated and reliable signal feedback for defect detection was substantiated. Also, AI predictions of the seam geometry from the UT data were developed using 2D extraction of the μ CT data as ground truth of the braze volume, covering a variety of joint geometries and sub-surface defect types. The process window studied effectively establishes the groundwork for

³ This chapter is composed of a research paper as outlined in the statement of contributions. **S. Akbarian**, R. Scott, A. Oullette, A. Salandari-Rabori, S. Sarafan, P. Wanjara, J. Gholipour, E. Hetric, B. Ghaffari, R. Maev, E. Biro, AI-friendly optimization of robotic laser weld brazing process advancing Industry 4.0 real-time inspection, to be submitted to Opt Laser Technol.

both UT detection and AI-based prediction of the joint geometry from the UT data, and AI-based detection of the defects warrants further investigation. A graphical abstract summarizing this chapter is presented in Fig. 4.1.

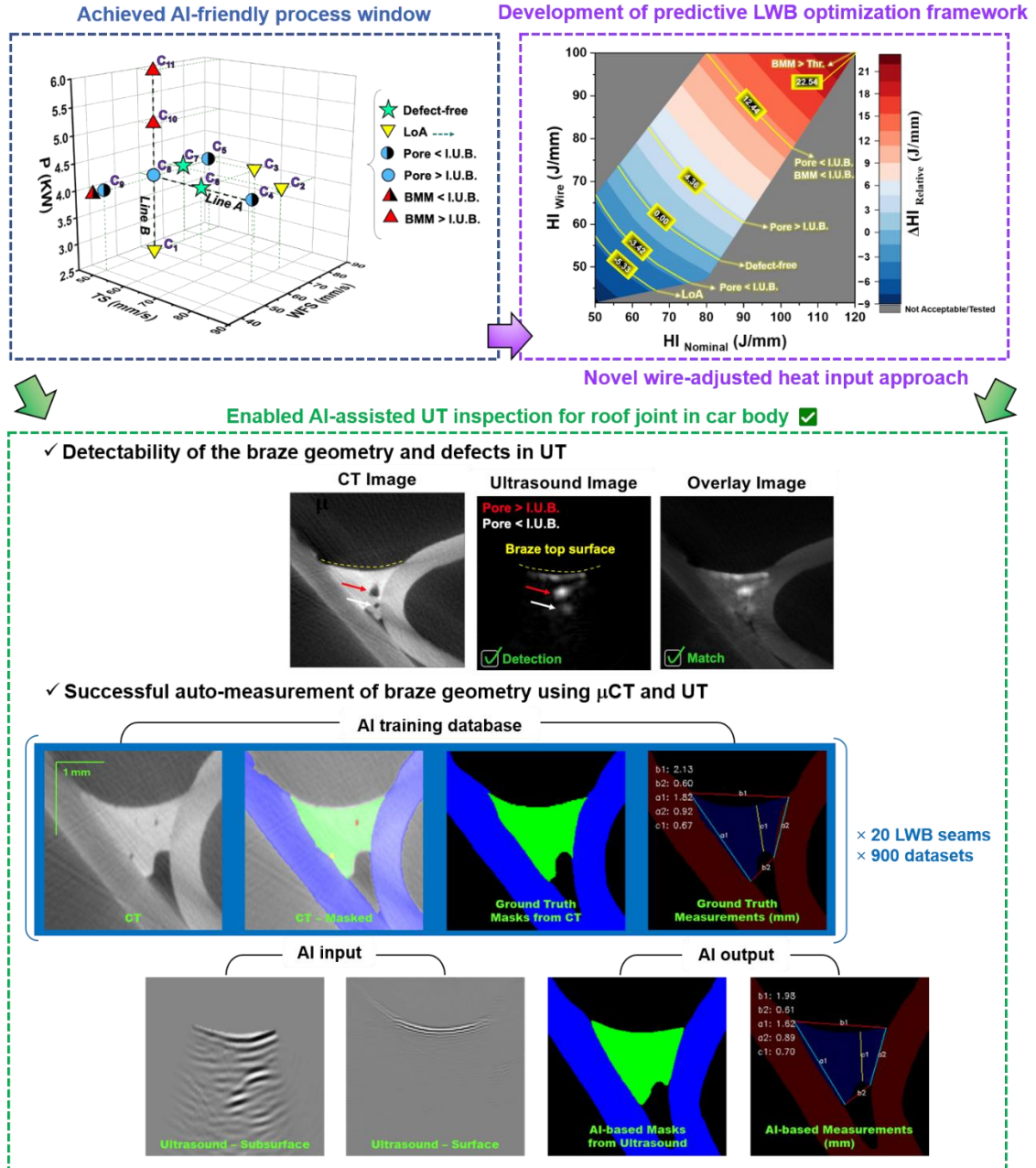


Fig. 4.1: Graphical abstract illustrating an overview of chapter 4.

4.2 Introduction

Robotic LWB has received significant attention in joint design in the realm of modern automated manufacturing [177]. According to Koltsov and Cretteur, this method has mostly been employed for joint assembly, particularly in the joining of roofs and hatchbacks, in luxury lines of well-known car companies such as Ford, General Motors, BMW, Volkswagen, and Renault-Nissan [13]. These applications have been typically brazed because substrate melting in weld brazing processes, in general, are minimized by limiting the input energy to only melt of the brazing filler metal, often in the form of wire, to join two or more similar/dissimilar substrates. The processing temperature only needs to exceed the T_m of the filler metal, which is typically selected to be lower than the T_m of the substrates. The use of laser processing for brazing offers further advantages. Narrow LWB offers aesthetically pleasing seamless joint appearance, which makes it an ideal joining method for regions visible to the customer, particularly in assembly of exterior body panels. Moreover, due to the precise energy input and the localized heat generation achieved by the laser heat source, LWB can increase productivity compared to traditional brazing methods, such as GMAB [49,178], by increasing travel speed. These factors make LWB a more energy-efficient choice for automotive brazing. Compared to GMAB, LWB offers nearly twice the joining speed along with a superior precision for intricate details, attributed to the higher energy density of the laser beam and its stability compared to the heat generated by the electric arc in arc welding processes [130,179].

Producing high-quality LWB seams requires strategies to minimize the formation of brazing defects during the assembly process. Brazing defects resulting in poor joint appearance are surface defects including cavities, surface-breaking pores, blow holes, spatter, insufficient wetting, or lack of adhesion, skipping, undercut, and wavy appearance seam [180]. In addition to surface defects degrading joint appearance, sub-surface defects, which cannot be seen from the surface, can significantly limit joint integrity. Porosity, lack of adhesion, incomplete fusion, insufficient penetration, base metal melting, and cracking are common types of sub-surface defects [180]. While surface defects are typically related to issues at the joint's outermost layer,

sub-surface defects in laser brazing are often tied to the precise control of heat distribution and penetration within the joint, as well as material interactions and joint design [15].

To inspect sub-surface defects in LWB seams at a manufacturing scale, various in-line NDT methods are employed. These methods include X-ray and μ CT scanning, eddy current testing, and UT [49,74,181,182]. Over the last decade, NDT technologies have undergone remarkable advancements through the adoption of AI [74,80,81,183–185]. For NDT applications, modern AI, particularly ML and DL, revolve around the use of data-driven algorithms and model architectures designed to replicate or augment human expertise in interpreting NDT data. The convergence of AI and NDT has ushered a new era by aiming toward zero-defect manufacturing by enabling 100% inspection coverage and traceability, real-time monitoring, and predictive maintenance strategies to diagnose the root causes of a detected defect and adjust the variables to resolve the issue instantly without shutting down the process line [82]. This can reduce human error, improve inspection reliability, achieve cost savings, and enhance efficiency. Defect detection using ML and DL often involves framing a signal/image processing stage as anomaly detection [83], classification (identification of a category for a given input) [84], object detection (classifying regions of interest and tightly enclosing them in bounding boxes) [85,186], and/or semantic segmentation (classification of pixels/voxels in inputs) [86,187]. Though these strategies have proven powerful in NDT applications, including in UT [188], the development and deployment of modern AI approaches has many significant challenges, especially in NDT of parts being made where safety is critical such as automotive assembly.

Besides the significant challenges in ML/DL model development and deployment – for example estimating deployed NDT system performance, estimating ML/DL prediction confidence, AI trustworthiness, and data drift – serious complications can arise even earlier in the modeling lifecycle in NDT applications. In particular, as the ML/DL approaches are data-driven, dataset size, dataset coverage, dataset balance, and biases are major concerns that can lead to poor performance and generalization of the AI in NDT systems. Thus, many researchers and developers opt to synthesize data via simulations, which can provide greater control over

dataset generation [189]. However, this approach can also yield models that fail to generalize to real applications, because simulations may produce biased or unrealistic datasets if they are not carefully developed and used. It is therefore far more preferable, when possible, to leverage real NDT data from real products, but this comes with significant challenges itself. In particular, many manufacturing processes are extremely expensive to perform, time-consuming, or difficult to replicate – even to produce nominal samples. To achieve sufficient dataset size for NDT defect detection, samples must represent the full range of real production defects. However, creating and verifying such samples through alternative NDT or destructive tests is a significant and often insurmountable challenge. Training AI to enhance real-time NDT requires a robust and reliable database that aligns with manufacturer inspection criteria, such as geometrical factors and defect intensity. It must also provide sufficient dataset variability, quantity, and repeatability, ensuring that the braze geometry remains consistent along each braze seam.

Achieving zero-defect manufacturing in LWB through real-time quality control systems necessitates the production of braze seams with well-characterized defects for AI training. These defects must be distinctly identifiable across various braze conditions to enable a deeper understanding of their complex origins. Yet, there is no direct reference to optimize the formation of defects that are arising during the LWB process. The most common known approach is defining the heat input of the LWB process by the nominal heat input, calculated by dividing the power by the brazing speed—two readily accessible process parameters [65]. However, nominal heat input does not account for the remainder of laser parameters, such as wire feed speed, wire diameter, and laser beam spot size diameter, which affect the heat necessary to melt the brazing wire and the heat available to raise the temperature of the workpiece to ensure that the molten braze can flow on the joint. Measurement of the real heat input of the process requires sophisticated calculation models [190–193]. Process optimization requires a more direct approach to define the process heat input to select principal parameters, such as power, brazing/travel speed, and speed of wire feeding. As an approach to incorporating the wire feed rate into process optimization, the ratio between wire feed rate-to-

travel speed is considered as an indicator, where a higher ratio reduces substrate mixing due to increased wire deposition [115]. However, no explicit formula has been proposed to define heat input as a unified parameter that simultaneously incorporates power, brazing speed, and wire feed rate. A key knowledge gap is the lack of establishment of a robust correlation between the principal LWB parameters and brazing defects, instead of implementing complex heat input models. In most LWB studies [26,193,194], a correlation of heat input related metrics with microstructure uniformity and bonding is provided qualitatively rather than quantitatively.

This study addresses the critical challenges of achieving high-quality braze seams with various classes of defects and generating sufficient datasets for AI training, aiming to provide a smart solution quality control for LWB joints in automotive manufacturing. A novel LWB optimization strategy, leveraging heat input-related parameters, was developed to navigate the formation of various sub-surface defects during the LWB of ultra-thin gauges of GI-IF steels in roof joint configurations. To generate comprehensive datasets for AI training, μ CT analysis was employed to detect and characterize braze geometry and defects across the entire seam. The AI performance for predicting braze geometry from UT data, trained using μ CT scans, was also evaluated. While μ CT scanning has been predominantly applied to single-material samples with simple geometries, its application to characterize defects in multi-material brazed joints with complex geometries remains underexplored. Since braze seams often exhibit variations across different cross-sections, understanding the defect development is critical. The defect evolution was statistically analyzed from μ CT scans of LWB joints to ensure compliance with AI-friendly specifications and robustness within the process window. This study offers key insights into optimizing the LWB process in pursuits of AI-assisted zero-defect manufacturing aligning with Industry 4.0.

4.3 Experimental Procedure

4.3.1 Materials Specifications

The material used in this study was thin-gauge, GI-IF steel that was 0.65 ± 0.02 mm thick. The Zn coating had a weight of 50 g/m^2 on both the top and bottom surfaces. The chemical composition of the substrates is detailed in Table 4.1. The filler material for the LWB process was a Si-bronze, CuSi3Mn1, brazing filler wire with a diameter of 1.2 mm. Its chemical composition is also listed in Table 4.1.

Table 4.1: Chemical composition (wt.%) of the GI-IF steel (substrate) and CuSi3Mn1 (filler) wire.

Material	Fe	Cu	C	Si	Mn	P	S	Al
GI-IF	Bal.	-	0.06	0.50	0.40	0.025	0.020	0.010
CuSi3Mn1	0.02	Bal.	-	2.90	0.80	-	-	-

4.3.2 Roof Joint Coupon Geometry and Fixture Design

The joint geometry used in the present study was designed to simulate those currently used in automotive construction for roof-ditch applications; so a double-flanged lap joint configuration was used, as shown in Fig. 4.2. The two coupons used in the joints differ in geometry, closely resembling the actual roof joint configuration used by the manufacturer. There is a height difference of 7 mm between the surfaces of these parts. To maintain this distance and ensure high consistency in positioning these parts during LWB, a custom-designed fixture was used [195].

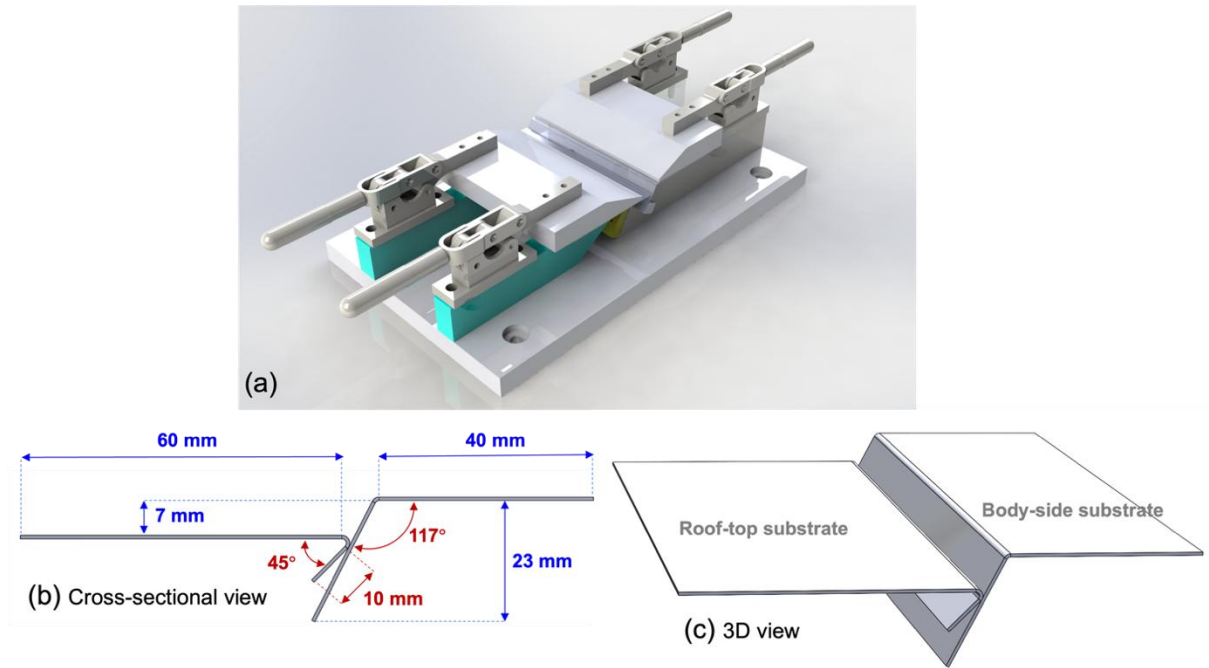


Fig. 4.2: SolidWorks® CAD models of the (a) custom-designed fixture to clamp the LWB coupons along with (b) 2D and (c) 3D views of the LWB coupons for the roof joint configuration.

4.3.3 LWB Process Setup

The LWB process was conducted to join the coupons, simulating the body-side and roof-top in a double-flanged joint configuration. This was performed using a multi-head robotic hybrid laser welder (Lincoln Wayne-Trail/IPG) at the Centre for Advanced Materials Joining (CAMJ) at the University of Waterloo. The facility is equipped with an IPG YLS-8000 fiber laser. The process parameters were selected and optimized to achieve industrially typical brazed joint geometries, while ensuring visually flawless seams on the top surface. For this study, all LWB experiments utilized a laser head emitting a beam with a wavelength of 1070 ± 10 nm. The laser head was equipped with a cold wire feeder; the beam had a waist diameter of a $600 \mu\text{m}$ and a Gaussian-profile. Brazing was undertaken using a larger laser beam spot size of 3.2 mm, achieved by defocusing the laser so that the focal point was 42 mm above the workpiece. The laser beam was directed at an incident angle of 90° relative to the substrate (x-y plane), while the filler wire was introduced at a 5° angle, as depicted in Fig. 4.2a and b. This arrangement

for aligning the filler wire, laser beam, and substrate interaction was maintained consistently throughout the experiments.

In keeping with standard industrial practices, no shielding gas was utilized during the brazing process. The primary LWB parameters— P , TS , and WFS —were varied within the ranges specified in Fig. 4.3a and b, specifically 2–6 kW, 50–85 mm/s, and 40–80 mm/s, respectively. These operational ranges are representative of those commonly employed in manufacturing. The double-flanged joint configuration resulted in a trapezoidal geometry at the braze seam cross-section [195] after the filler wire melted and solidified, as illustrated in Fig. 4.3c.

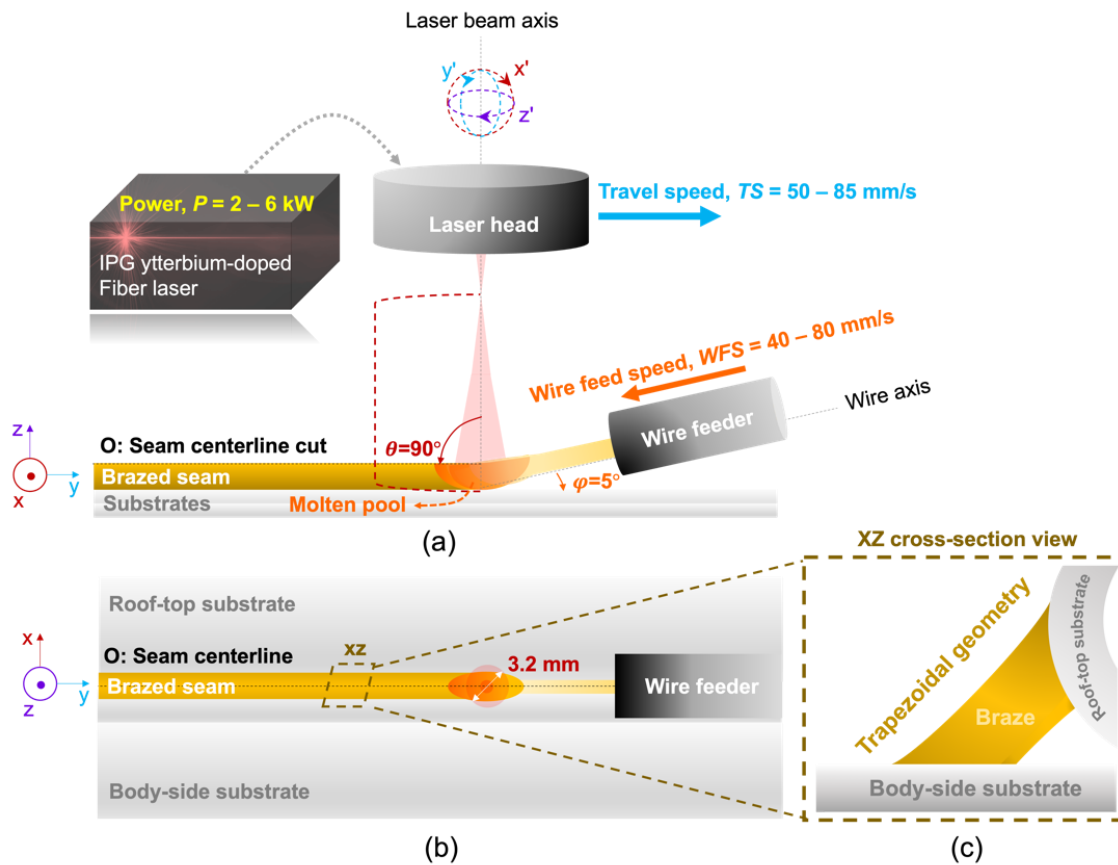


Fig. 4.3: (a,b) Schematics of the LWB process setups showing non-variables, such as spot size diameter, defocused distance, and the angles of incidence for the laser, wire, and substrate. The experimental domains for principal process variables, including P , TS , and WFS , are also presented; (c) representation of the double-flanged braze geometry.

4.3.4 AI-friendly Specifications

The LWB seam generated for enrichment of the AI development dataset must satisfy certain specifications, including (1) the manufacturer's braze geometry and the maximum acceptable imperfection/defect criteria; and (2) a low variation in the type of imperfections/defects along each braze seam. The manufacturer's criteria are shown in the tables in Fig. 4.4. Conforming to these factors enables AI recognition of the braze status for certain LWB conditions using UT. Brazing imperfection/defect conditions of interest are LoA, porosity, and BMM as proposed by the manufacturer. It is worth mentioning that imperfections were considered as brazing defects only when they exceed the imperfection upper bound (I.U.B) limits in AI-friendly specifications.

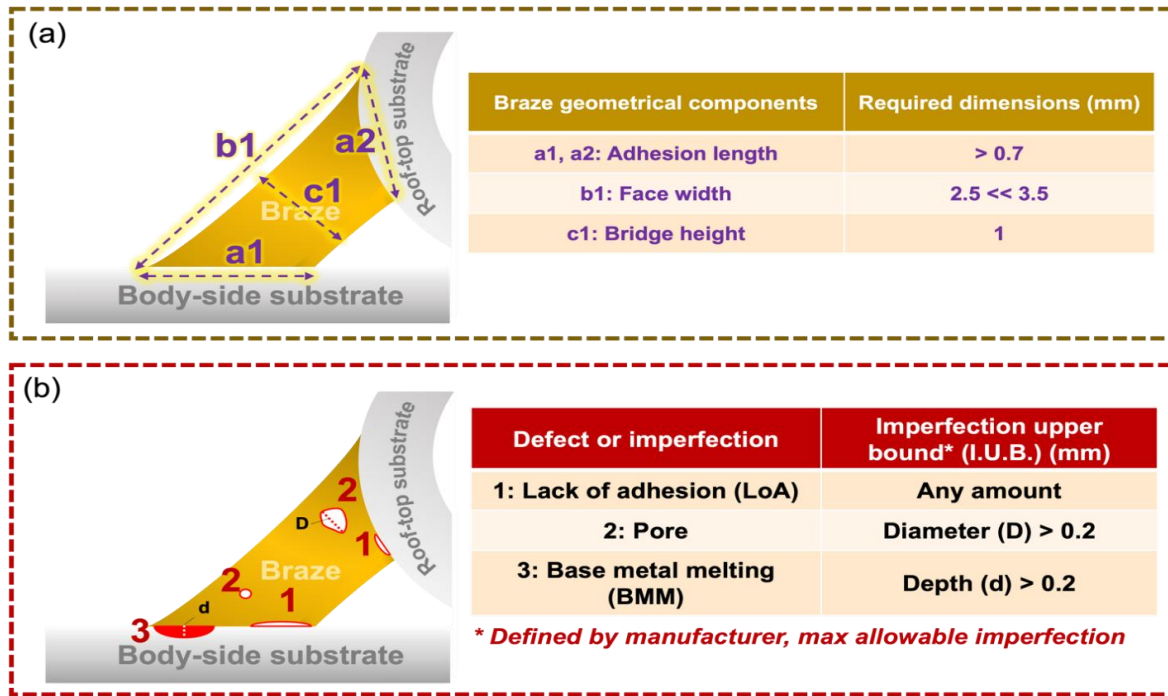


Fig. 4.4: AI-friendly specifications for (a) geometry and (b) defect or imperfections detected within the trapezoidal braze geometry of the roof joint.

4.3.5 Defect Characterization

4.3.5.1 Metallography Technique (MT)

The braze microstructure and geometry in each LWB condition were investigated by cross-sectioning the LWB seam perpendicular to the brazing direction. A highly-controlled polishing procedure was used on the four cross-sections extracted from each LWB seam, according to a metallographic method described in [179]. An optical Keyence VHX-7000 microscope was used to observe the polished brazed surfaces to provide a larger field of view for capturing the whole braze geometry of each cross-section.

4.3.5.2 X-ray Micro-Computed Tomography (μ CT) Scanning

To enable a comprehensive analysis of the braze geometry and sub-surface defects, such as pores, μ CT scanning was performed on the braze seams under selected conditions. This method plays a crucial role in generating extensive datasets suitable for integration into an AI database. The μ CT technology is a computer-based image reconstruction of a series of X-ray images obtained from specified cross-sections with different density and display in the form of high-resolution digitalized 3D images [196]. The density of the tested material is analyzed using the transmission of X-rays, where the reduction in X-ray intensity is quantified by the material's attenuation coefficient, which depends on the material's composition and density. The intensity of X-ray transmitted through an object shows an exponential decay respect to the density of the tested sample [197]. The μ CT outputs a volume of voxels that can be sliced into different gray scale images, with each gray scale number corresponding to the density of the material at a specific position. The greater the material density is, the higher the brightness on the μ CT image (white-prone image). By contrast, voids have the smallest density, which will be shown as dark black on the image. This density and structure correlation provide a quantitative method to analyze and reconstruct the whole structure based on stacking the μ CT images.

The scanner used in this study was a ZEISS Xradia 520 Versa 3D X-ray Nano-CT Scanner at the Multi-Scale Additive Manufacturing Laboratory, at the University of Waterloo. The system includes a 160 kV X-ray source manufactured by Hamamatsu Photonics K.K. with a maximum focal spot size of 4.5 μm . To detect pores of as small as 20 μm , a voxel size of 20 μm was selected to provide sufficient spatial resolution; μCT scans were performed at micro-scale. To allow for this voxel size, scanning volume was limited to $20 \times 20 \times 20 \text{ mm}^3$. To fully scan the seam of each braze, samples were first cut using a waterjet machine, as depicted in Fig. 4.5a to extract μCT samples with a size of $100 \times 20 \times 20 \text{ mm}^3$ that could fit into the X-ray chamber (Fig. 4.5b). Then the desired scanning volume was achieved by virtual cross-sectioning, followed by scanning 20 mm long sections, in-turn, from each braze (see Fig. 4.5c), and measuring the virtual volume as seen in Fig. 4.5d. After scanning, the projection image was reconstructed using the Zeiss Reconstruction software package, resulting in a series of grayscale images with 16-bit intensity ranges. Advanced image processing and porosity analysis were performed using Dragonfly Pro v3.1 [198]. Each scan consisted of 1018 2D slices, as exemplified in Fig. 4.6a. Fig. 4.6b shows the post-processed 2D slices to view the actual specimen in 3D, where the surface of the braze seam is also exhibited. For an accurate comparison between the two scans, μCT scans were all cropped to mainly focus on the braze region and its neighboring substrates.

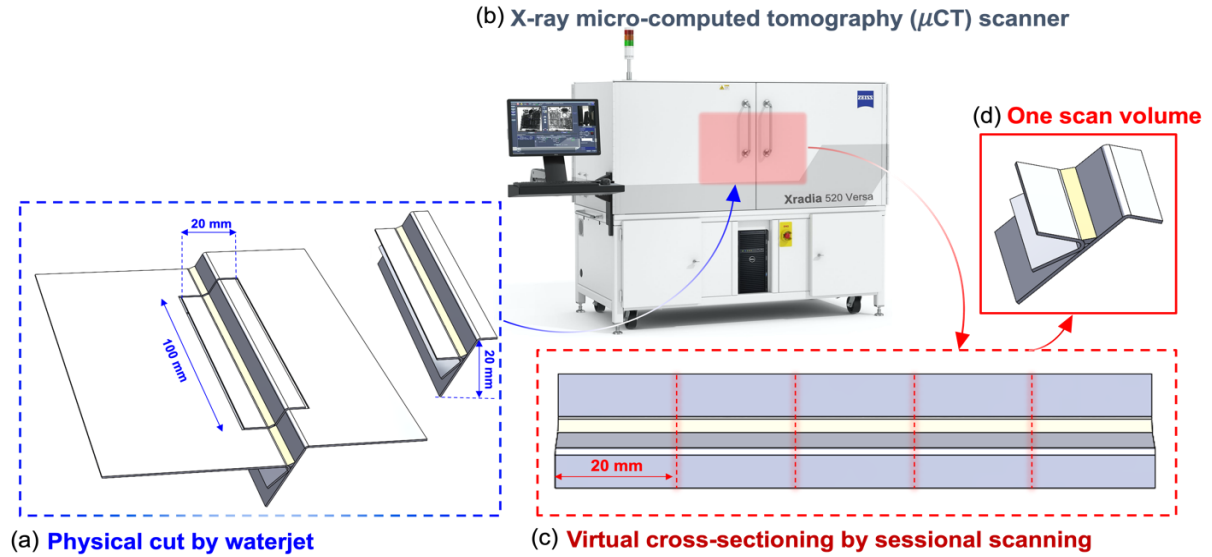


Fig. 4.5: (a) CAD model of the LWB seam coupon with a roof joint configuration showing the geometry of the extracted sample for μ CT scanning, (b) μ CT scanning device, (c) virtual cross-section from the full-length (100 mm) μ CT scanning sample for each 20 mm length during μ CT scanning by dividing into 5 parts, and (d) volume geometry per scan to achieve 20 μ m voxel size.

To evaluate the pore formation, a semi-automatic segmentation procedure was used to facilitate full detection of pores within the braze seam. Pore identification in the braze joints is challenging because the built-in segmentation modules in DragonFly software may confuse pores with geometry-related voids in the sample. Additionally, some pores may be missed, even if they fall within the predefined ranges. For example, as seen in Fig. 4.6c only a pore with size below 0.04 mm is detected, while the larger pore is missed. To achieve highly accurate pore detection, post-processing of μ CT data was initiated with training an automated pore segmentation module and followed by manual correction. Undetected pores were identified and painted within the region of interest (ROI) defined for pores in each 2D slice of the preliminary pore map to ensure that they were properly integrated into the final output. In the final stage, pore quantification, ROIs containing pores smaller than the voxel size used in the settings were removed from the output. The resulting ROI in Fig. 4.6d showed that the larger pore size with 0.36 mm diameter is detected, confirming a fully accurate pore detection.

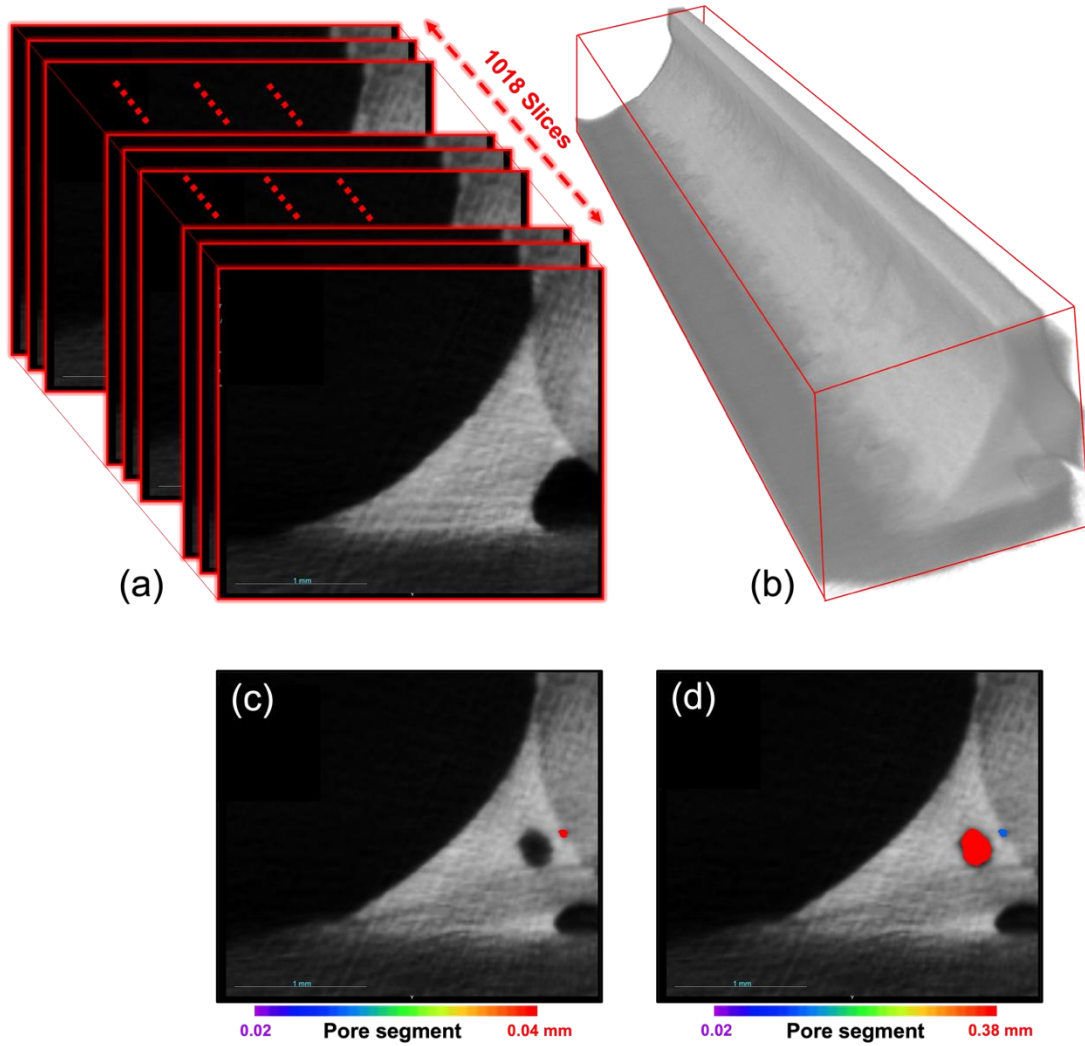


Fig. 4.6: Post-processing procedure of pore analysis in DragonFly software on μ CT accrued data. (a) stacked 2D μ CT slices of 1018 for each 20 mm length/portion of the braze seam, (b) 3D processed view of the braze in 20 mm length, (c) example of pore segmentation using an embedded AI module for pore detection in Dragonfly software showing poor pore identification, and (d) manual modification of pore segmentation showing corrected pore identification.

4.3.5.3 Ultrasonic testing (UT)

To carry out UT of the LWB samples, a methodology based on prior research was used where the samples were sequentially imaged, and the images were reconstructed (similar to what was done for the μ CT) [74,186,199]. Here, a linear array placed over the surface is coupled using water. For comparison to existing μ CT data, a 4 axis isel cartesian scanner was used to align the ultrasound probe with the sample and allow for an encoded scan to occur along the direction of the braze. The setup used to acquire the sequential ultrasound images, at a pitch of 0.1 mm, is depicted in Fig. 4.7a. For each point, a linear 64 element array, with a layout presented in Fig. 4.7b was used, with properties indicated in Table 4.2 to acquire the data. This array was chosen based on optimization from previous work [74,186,199]. Signals corresponding to each transmit-receive combination of the array were acquired for a total of 4096 signal combinations. This data, known as full matrix capture data, was acquired using the commercially available AOS OEMPA mini [200].

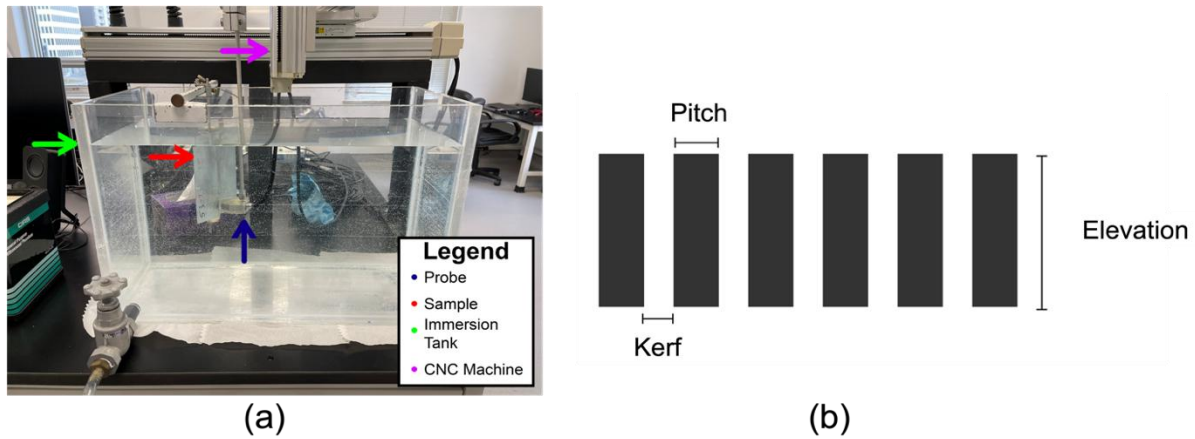


Fig. 4.7: (a) A diagram of the ultrasound acquisition system. Here, a water tank is used to scan along the laser weld brazed sample, acquiring data at each point. (b) A depiction of the linear transducer layout and properties.

Table 4.2: Properties of the ultrasound transducer used in the data acquisition process.

Parameter	Value	Description
Frequency	18.5 MHz	Central frequency at which the transducer pulse is sent
Bandwidth	65%	Frequency bandwidth of the transducer, with a higher bandwidth indicating a shorter pulse length
Pitch	0.15 mm	Size of each element in the array along the direction of phasing
Kerf	0.025 mm	Size of the gap between each element in the direction of phasing
Element Elevation	5 mm	Size of the elements in the non-phased direction
Number of Elements	64	Number of elements in the phasing direction
Elevation Focus	15 mm	Ideal imaging depth of the non-phased direction

This data acquisition method allows for the post processing of acquired data to create an image via a method known as the total focusing method using Equation 4.1 [188]. Here, the signal emitted at a point, x_n , at a time, t_0 , is allowed to propagate through a material using the time evolution operator, H , interact with an observable, O , propagate further, and be recorded at a later time, t , by a receiver located at a position $x_{n'}$. To generate an image at position x and time t , the summation in Equation 4.1 is applied. It was assumed material properties are temporally invariant during acquisition, so the signal is sampled at a point $t_q + t_{q'}$, corresponding to the most probable path between the emitter and receiver that passes through the point x . This signal is then weighted by the amplitude of the signal expected when the observable is a perfect, point source reflector. By summing all possible transmit receive pairs, and normalizing by the sum total amplitude, A_{tot} , of all possible paths, a likelihood measure of a reflector being present at a spatial position is computed.

$$\langle x, t | O | x, t \rangle = \frac{1}{A_{tot}} \sum_{n', n} A_{n'q'} A_{nq} \langle x_{n'}, t_0 + t_{q'} + t_q | HOH | x_n, t_0 \rangle \quad (4.1)$$

In this work, this process is performed in a multi-step process [74,199]. In the first step, the reconstruction is performed in the coupling material, allowing for an analytical computation of the corresponding amplitudes and travel times. In the second step, an estimate of the location of the braze surface is measured, forming a boundary through which sound must pass to interact with the internal structure of the braze. Next, a numerical simulation is performed to determine the corresponding travel paths from each emitter to each point in space, while accounting for the effects of the boundary and the acoustic properties of each medium. Using Helmholtz reciprocity, the return path to the receiver is simultaneously known. Combining the most likely propagation paths in each case allows for the time-migration function (TMF) to again be used to form a sub-surface image of the material. In regions where zero or near zero signals are expected, a value of zero is assigned to prevent singularities. The computation of multiple images along the braze seam allows for a volume of ultrasonic data to be obtained. In order to allow for comparison to μ CT data, a rigid body image registration algorithm was developed to allow for 3D alignment of the two data sets.

4.3.5.4 AI-based analysis of UT data

Extraction of the joint geometry and defect indications from UT data was framed as a supervised semantic segmentation task, wherein each pixel in the input ultrasonic data is designated a class and models are optimized given corresponding ground truth for each given input. To this end, brazed joints, substrate sheets, pore indications, and LoA indications were segmented in the high-resolution μ CT scan data from 20 LWB samples with 10 cm length to derive a set of masks of the different regions for each individual μ CT image. To enable the transfer of μ CT-based ground truth data to the UT dataset, μ CT scans, μ CT-based masks, and UT scans were aligned using cross-correlation of front and back surfaces of the brazed joint regions. The μ CT scans were acquired with 5x greater resolution than UT longitudinally along the braze, and the UT samples aggregate information within the scanned region of the joint; so a set of 25 μ CT scans were mapped to the corresponding central UT scan, and the UT ground

truth for semantic segmentation was the average of the masks for the corresponding 25 μ CT scans.

Semantic segmentation models were based on the U-Net [201] architecture. The models were optimized and trained using ultrasonic images as input and their corresponding aggregated ground truth masks as target outputs. A nested cross-validation approach was used, where the dataset was partitioned based on the joint sample rather than random image slices. Models were evaluated on joints that were held out from model training within the outer cross-validation folds. Both ground truth (from μ CT) and AI-based (from UT) mask outputs of the LWB samples were post-processed to enable the extraction of measurements that are often subject to acceptance criteria, as well as comparison between the approaches. Extracted measurements along the joints were smoothed using rolling median.

4.4 Results and Discussions

4.4.1 LWB Process Window Design using MT

To identify the optimal conditions for producing AI-friendly LWB seams, a comprehensive screening process was performed using OriginPro software. A full factorial design of experiments (DOE) generator with a randomized design was employed to maximize the degrees of freedom in selecting LWB parameter sets. Three factors— P , TS , and WFS —were considered, each with multiple levels: 3 for P , 3 for TS , and 4 for WFS . These levels were constrained within the ranges specified in Fig. 4.3. The blue spheres in the 3D plots in Fig. 4.7a illustrate the parameter combinations used in the study.

The blue spheres with green halos, conditions a-h in Fig. 4.8a, highlight the preliminary screening window of conditions that meet the AI-friendly specifications for geometry limits and defect classifications. Five classes of LWB conditions were identified by MT: defect-free, LoA, brazes with pore formation below and above 0.2 mm in diameter, and brazes with BMM that was less than 0.2 mm in depth. These conditions namely, “a” to “h”, are represented by corresponding symbols in the bracket tree in Fig. 4.8a.

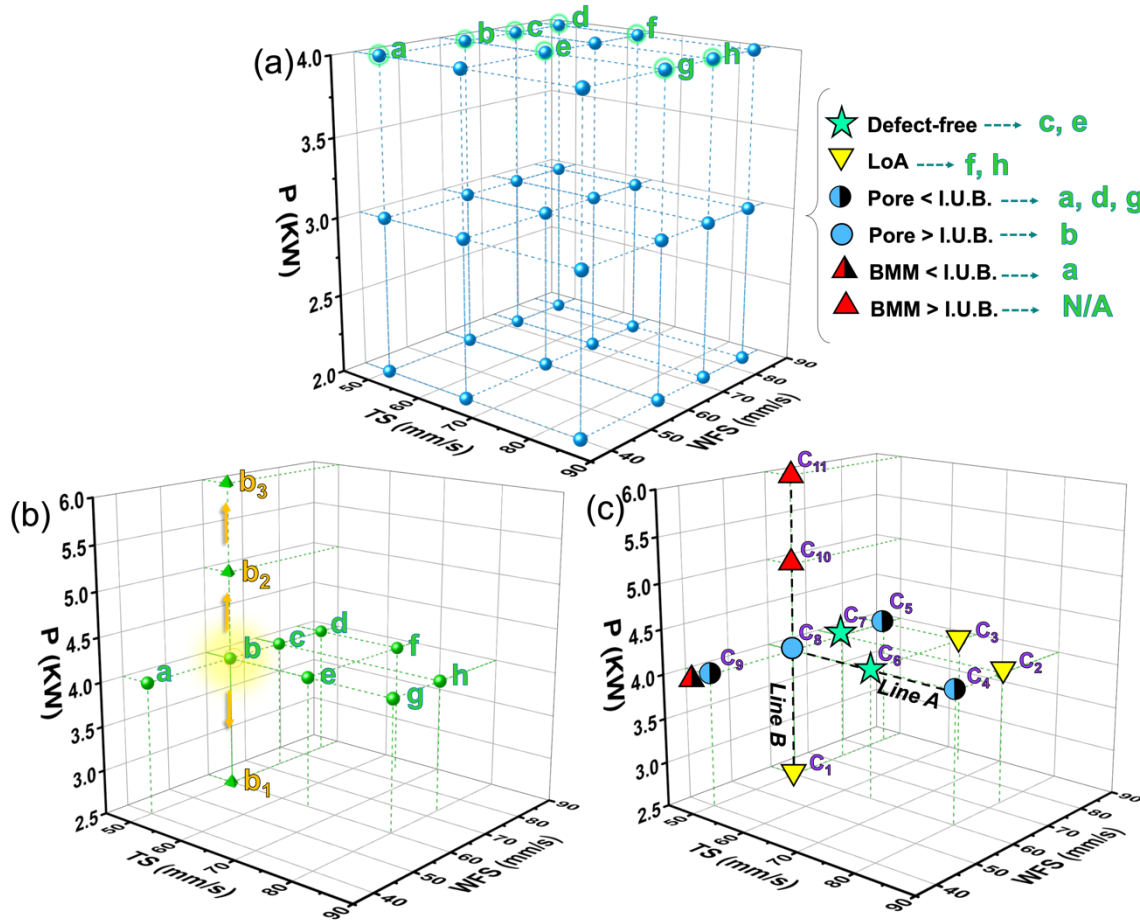


Fig. 4.8: 3D plots of LWB process window of principal variables, including P , TS , and WFS (a) preliminary screening window with partially achieved defect conditions as highlighted by the green halo, and (b) modified window by embedding new points “ b_1 ”, “ b_2 ”, and “ b_3 ”, (c) resulting LWB window which passed AI-friendly geometrical and defect specifications experimentally.

The LWB parameters causing BMM greater than 0.2 mm (BMM defect) has not yet been achieved from the preliminary screening window in Fig. 4.8a. To explore LWB conditions which accelerate melting of the base metal, condition “b” from the preliminary screening window was selected. This condition, “b”, contains the most severe defects such as large pores (above the pore I.U.B), which formed under an increased heat input. By further adjusting the heat input relative to condition “b”, moving to points above and below point “b” in Fig. 4.8b—where TS remains constant—a wider range of defects is expected. Condition “b” was selected

because the formation of large pores is a sign of high instability in the braze molten pool, increasing the likelihood of defect formation. It was observed that condition “b₁” with reduced P (2.5 kW), led to LoA defects in the brazing seam, even though the TS was the lowest level within the modified window. Conversely, increasing the P to 5 and 6 kW, conditions “b₂” and “b₃”, respectively, resulted in BMM exceeding I.U.B. for BMM, 0.2 mm.

The resulting window from MT, encompassing all required defect conditions satisfying the given AI-friendly specifications (as specified in Fig. 4.4), is depicted in Fig. 4.8c, in which the achieved conditions are indexed by C_i labels. To avoid presenting identical defect conditions, six points showcasing defect-free and defect conditions such as C_1 , C_4 , C_6 , C_8 , C_9 , and C_{11} , are considered for further investigation in this study and their corresponding micrographs are gathered in Fig. 4.9. It is important to emphasize that all achieved conditions exhibited no surface defects in the brazing seam and the braze geometry using MT measurements conformed to the limits specified for AI-friendly samples. A detailed summary on the braze geometry measurements of candidate conditions is provided in Table 4.3. Subsequently, μ CT analysis was conducted on the produced LWB seams within the resulting MT-defined process window, enabling full-scale characterization of the braze seam geometry and sub-surface defects from along the entire length of the seam.

Table 4.3: Braze geometry measurements (from MT) of candidate conditions showing AI-friendly geometry specifications passed successfully.

Candidate C_i	Bridge height, c_1 [mm]	Adhesion length, Body side, a_1 [mm]	Adhesion length, roof top, a_2 [mm]	Face width, b_1 [mm]	Passed?
C_1	0.73 ± 0.08	1.61 ± 0.07	0.87 ± 0.12	2.33 ± 0.11	Yes
C_4	0.59 ± 0.09	1.33 ± 0.28	0.77 ± 0.10	2.09 ± 0.25	Yes
C_6	0.72 ± 0.25	1.83 ± 0.23	1.11 ± 0.16	2.41 ± 0.22	Yes
C_8	0.83 ± 0.05	2.11 ± 0.17	1.16 ± 0.20	2.50 ± 0.34	Yes
C_9	0.73 ± 0.09	2.27 ± 0.43	1.17 ± 0.37	2.65 ± 0.51	Yes
C_{11}	0.78 ± 0.26	2.66 ± 0.34	1.28 ± 0.49	3.30 ± 0.30	Yes
AI-friendly specification	~ 1	> 0.7	> 0.7	$2.5 << 3.5$	

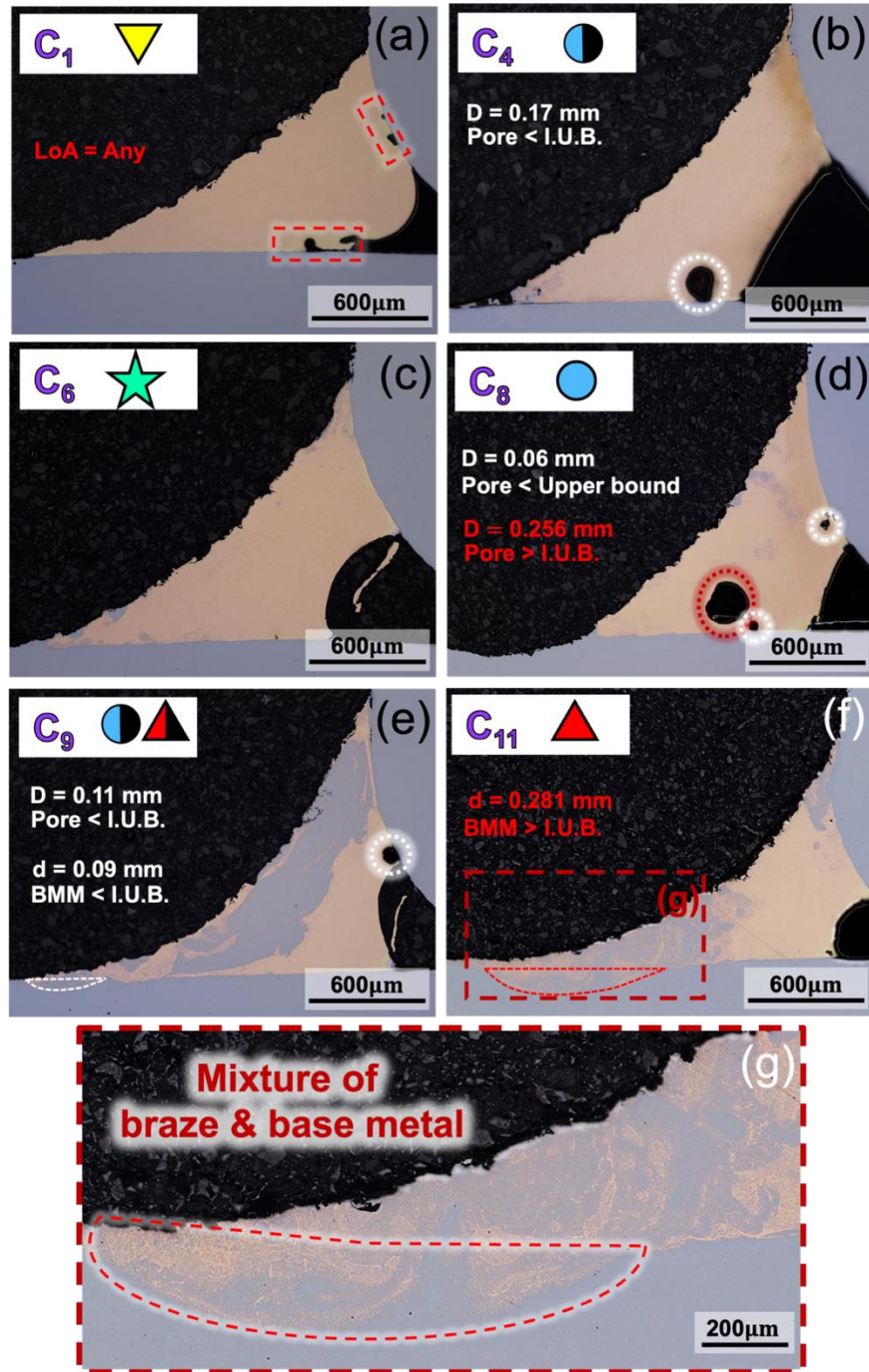


Fig. 4.9: Braze microstructure of LWB conditions satisfying AI-friendly specifications on geometry and defects. Condition indexes are with respect to Fig. 4.7c. Imperfections with a size below and above (defect) the I.U.B. are labeled using white and red annotations.

4.4.2 A Novel Heat Input-related Approach for LWB Defect Optimization

To determine the correlation between the different defect classes and LWB variables, the process parameters laying along two lines within the AI-friendly process window in Fig. 4.8c were isolated to minimize changing parameters. These selections were made to analyze conditions with constant P and WFS (in line A), and conditions with constant TS and WFS (in line B). This comparison was based on the nominal heat input ($HI_{Nominal}$), as defined in Equation 4.2. $HI_{Nominal}$ is the conventional method for presenting heat input for laser brazing [65].

$$HI_{Nominal} = \frac{P}{TS} \quad (4.2)$$

Line A, including C_8 , C_6 , and C_4 , was graphed using $HI_{Nominal}$ plotted against TS , as shown in Fig. 4.10a. As $HI_{Nominal}$ increases by reducing TS , various defect classes emerge within this regime. At the lowest TS (50 mm/s) in C_8 , the braze seam exhibited large pores measuring 0.26 mm, exceeding the pore I.U.B. (0.2 mm). The increase in pore size in C_8 can be attributed to two primary causes. Firstly, the prolonged laser beam exposure time, due to its lower TS (compared to C_6 and C_4), leads to excessive vaporization of metals, Zn-coating, and moisture that forms large bubbles and produces gases which can be trapped during solidification. According to previous works [74,192], gas trapping in the braze during solidification results in pore formation. Secondly, although brazing at lower TS diminishes the solidification rate, the braze still solidifies faster than the escape rate of large bubbles, due to the increased volume of trapped gas in the molten braze. Increasing the TS to 65 mm/s in C_6 eliminated defect formation, including pores. Despite the higher solidification rate caused by the increased TS compared to C_8 , the vaporization rate of Zn and the associated fume generation were effectively reduced, preventing the formation of pores and other defects. However, further increases in the TS to 80 mm/s in C_4 led to pore formation again, but these pores were smaller, 0.113 mm, which is below the pore I.U.B. (0.2 mm). The appearance of pores in C_4 confirms that vaporization of metals, Zn-coating, and moisture can still occur even at high TS where $HI_{Nominal}$ is low. However, pore formation in C_4 can be attributed to degraded adhesion and wettability, primarily caused by the increased solidification rate resulting from the shorter laser

beam exposure time at high TS ; this condition represents a critical I.U.B where the initiation of LoA defects occurs. C_6 , with no defects or pores, can be identified as the optimal setting where vaporization and solidification rates are balanced. Thus, by correctly setting the TS or laser exposure time, pore development can be effectively controlled.

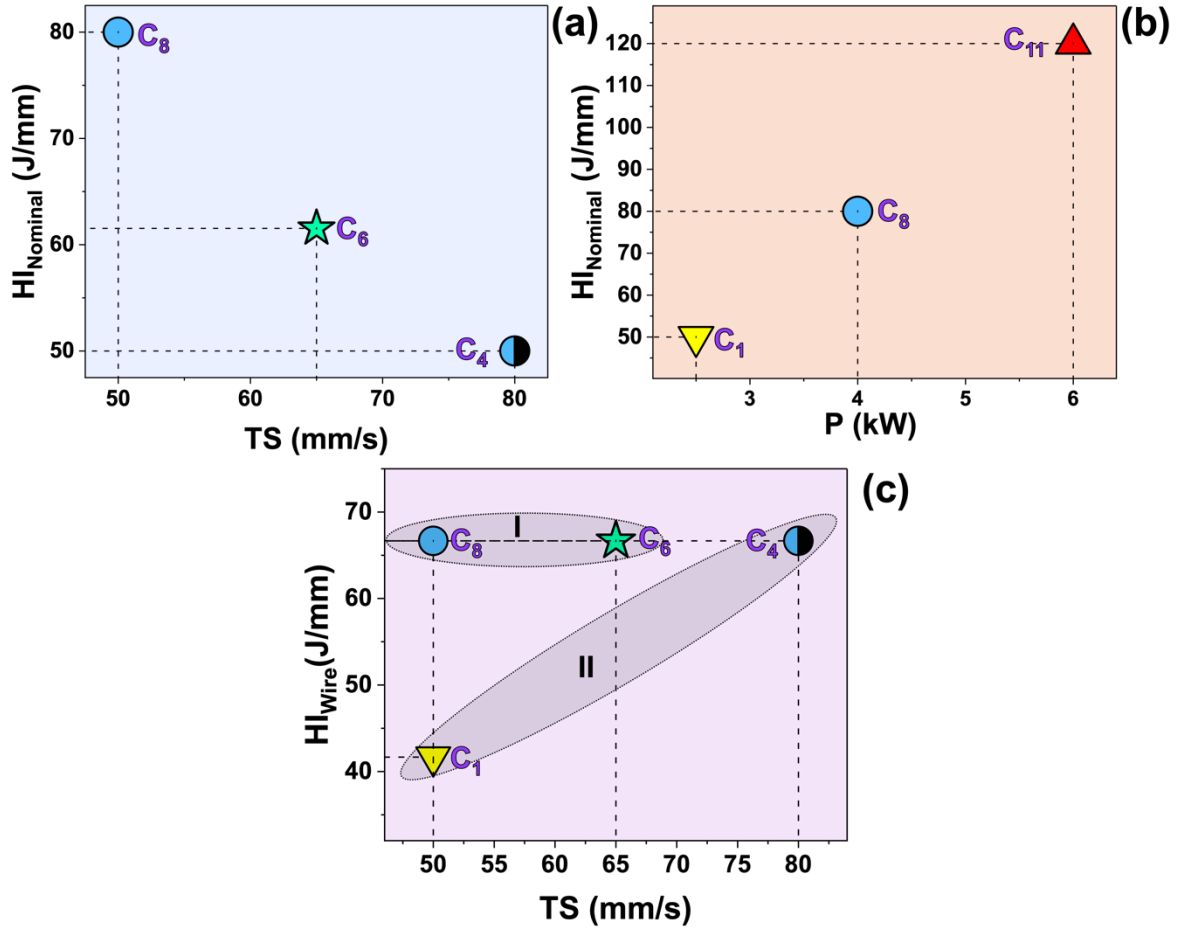


Fig. 4.10: Correlations between defects and $HI_{Nominal}$ change via (a) TS and (b) P . (c) Exploring the effect of HI_{Wire} under changing TS on defect development.

Line B, which includes conditions C_1 , C_8 , and C_{11} , reveals the effect of P on $HI_{Nominal}$ and defect development, as shown in Fig. 4.10b. In C_1 , with a power setting as low as 2.5 kW (the lowest $HI_{Nominal}$ in this regime), insufficient melting of the braze material at the bottom of the braze root interrupted bonding at the interface, causing a LoA defect. Increasing P to 4 kW

and 6 kW, in C_8 and C_{11} , respectively, prevented the LoA defect but promoted the formation of large pores in C_8 and BMM in C_{11} , respectively. As revealed from the comparison in Line A in Fig. 4.10a, the C_8 condition with the lowest TS caused the formation of large pores due to the prolonged exposure time of the laser beam. In Fig. 4.10b, the comparison in Line B further indicates that the increased linear energy delivery, resulting from the higher P in C_8 compared to C_1 , also contributed to the formation of large pores exceeding the upper bound. The occurrence of the BMM observed in C_{11} resulted from surpassing the process temperature beyond the T_m of the base material. This caused the incorporation of base material into the braze region, which facilitated the diffusion of the braze into the base metal, resulting in deep penetration that exceeded the upper limit for the BMM defect.

In the above discussed conditions, pores, LoA, and BMM are all found to be affected by $HI_{Nominal}$ under different P and TS levels. The lingering question is, how does WFS contribute to this result? The WFS is constant in these conditions but in relation to P and TS it is changing. To investigate this matter, a HI -related parameter called wire heat input (HI_{Wire}) is defined as the ratio of P to WFS in Equation 4.3 [195], which is irrespective of the TS of the laser torch; this HI -related component is a simplification for the heat required to melt the filler wire under the assumption of considering laser as a stationary heat source.

$$HI_{Wire} = \frac{P}{WFS} \quad (4.3)$$

$HI_{Nominal}$ has been extensively used in the literature to date, while HI_{Wire} has not been considered in the interpretation of the heat input in LWB, so it is worthy to investigate how it contributes to defect formation. The LWB conditions resulting in different defect classes were graphed in terms of HI_{Wire} vs. TS which revealed two types of correlations, separated in two groups, I and II, as shown in Fig. 4.10c. For the sake of elucidating each brazing condition comprehensively, a wire-adjusted HI term noted by $C_{i=1,2,3,...,n}$ is defined to collect all three HI components in 2D space as described in Equation 4.4.

$$C_{i=1,2,3,...,n} : (HI_{Nominal}, HI_{Wire})_X \quad (4.4)$$

In group I, C_8 :(80 J/mm, 66.6 J/mm)_{Pore>Thr.} and C_6 :(61.2 J/mm, 66.6 J/mm)_{defect-free} both use equal amount of HI_{Wire} but with different $HI_{Nominal}$. In contrast, group II compares C_1 :(50 J/mm, 41.6 J/mm)_{LoA} and C_4 :(50 J/mm, 66.6 J/mm)_{Pore<Thr.}, with equal $HI_{Nominal}$ and varied HI_{Wire} . In group I, the formation of pores with a size above the I.U.B has been promoted, due to the excessive $HI_{Nominal}$, while C_1 in group II experienced LoA due to the insufficient HI_{Wire} . In fact, it can be concluded that pore formation is sensitive to $HI_{Nominal}$, while LoA is sensitive to HI_{Wire} . However, this method of assigning two components for heat input to identify each condition's defect classes may not be effective. Analyzing the correlations between each braze defect condition and the HI components ($HI_{Nominal}$ and HI_{Wire}) suggests that there might be another component that can relate all these factors together, called relative heat input ($HI_{Relative}$). This component, described in Equation 4.5 [195], determines the heat delivery to the laser/wire/substrate interaction zone at a relative process speed, Equation 4.6 [195], as follows:

$$HI_{Relative} = \frac{P}{V_{Relative}} \quad (4.5)$$

$$V_{Relative} = TS + WFS \quad (4.6)$$

The units for all HI-related components in Equation 4.2-4.6 are expressed in Joules per unit of travel distance. The $V_{Relative}$ in Equations 4.5 and 4.6 is defined as the sum of TS and WFS, where an increase in either parameter reduces the amount of heat transferred to the system. It was observed that each brazing condition, in relation to the defects, occurs at a specific value of $HI_{Relative}$. Such a correlation could not be identified by relying solely on $HI_{Nominal}$ and HI_{Wire} values. The contour map in Fig. 4.11 provides a comprehensive overview of the correlations between HI -related components and various brazing conditions. The contoured axis in Fig. 4.15 represents the deviation of the $HI_{Relative}$ of each condition (X) with respect to the $HI_{Relative}$ of the defect-free condition, which is found to be 32 J/mm, as indicated in Equation 4.7 [195]. To elucidate each brazing condition comprehensively, the wire-adjusted HI term in Equation 4.4 was revised to conform to a 3D space, as described in Equation 4.8 [195].

$$\Delta HI_{Relative} = (HI_{Relative})_X - (HI_{Relative})_{Defect-free} \quad (4.7)$$

$$C_{i=1,2,3,...,n} : (HI_{Nominal}, HI_{Wire}, \Delta HI_{Relative})_X \quad (4.8)$$

Using $\Delta HI_{Relative}$ made navigating through each brazing condition straightforward and consistent. This method can predict which defect condition may arise under a specific LWB condition based on the corresponding $HI_{Relative}$, avoiding the confusion of the two other HI components. All conditions and corresponding HI-related parameters are summarized in Table 4.4. It should be noted that $HI_{Relative}$ depends on the wire diameter and, therefore, cannot serve as a universal measure. To achieve that, the heat of melting for the filler addition, as well as process efficiency, which is affected by laser energy loss due to reflection from the substrate surface, must be calculated, which is beyond the scope of this work.

The yellow lines in Fig. 4.11 delineate the boundaries where each brazing defect forms. The LoA defect is particularly prone to occur at a $\Delta HI_{Relative}$ of -5.33 J/mm, transitioning, as it increases to -3.42 J/mm, to pores with sizes below the I.U.B. Both conditions sit well below the defect-free condition where $\Delta HI_{Relative}$ is zero. By contrast, positive deviations from the defect-free condition led to the formation of larger pores and BMM defects. For instance, at 4.36 J/mm, there is a significant driving force available to form pores with sizes beyond the pore I.U.B. (0.2 mm diameter). With further deviation to 12.44 J/mm, the pore sizes reduce, but the BMM defects start to emerge with depths below the BMM I.U.B. (0.2 mm depth). At 22.54 J/mm, BMM defects exceed their I.U.B, defining the BMM defect formation region within this optimization scheme. Although the $HI_{Relative}$ and $\Delta HI_{Relative}$ values are not the exact amount of energy input being used or required for the aforementioned conditions, they can be considered as references to set explicit boundaries distinguishing between each defect formation condition to facilitate LWB optimization when looking for certain defect formation condition(s).

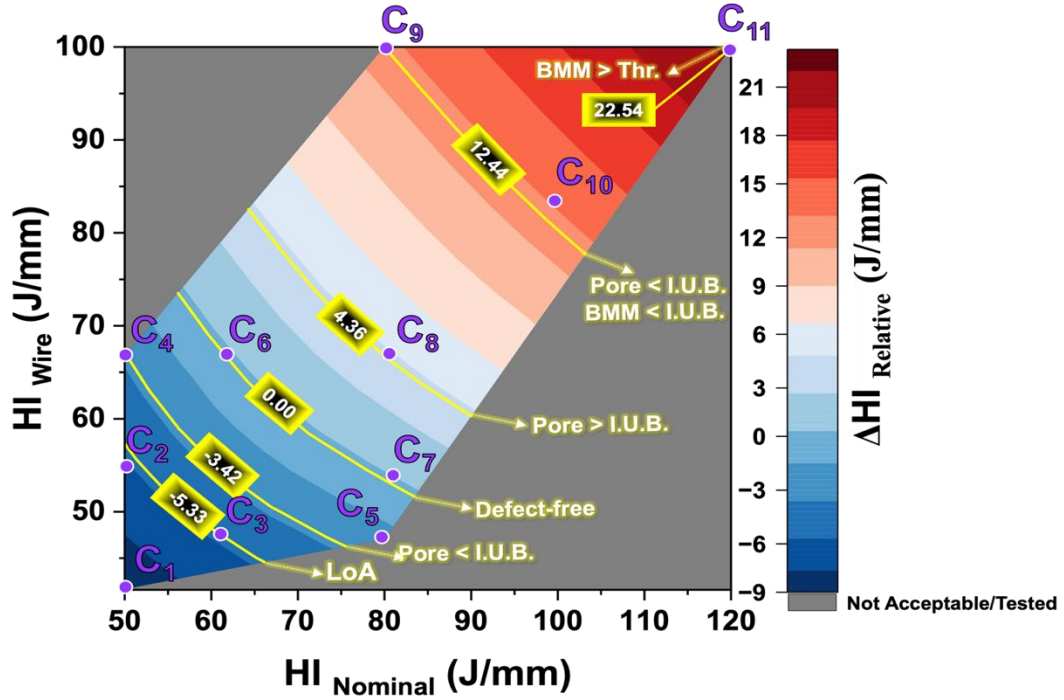


Fig. 4.11: LWB process road map showing the correlation of heat input-related parameters with various defects conditions.

Increased heat inputs and their impact on pore size have been extensively explored for laser welding; Atabaki et al. [202] reported a transition from small to larger pores with escalating energy levels. This aligns with the present findings that showed that an increase in $\Delta HI_{Relative}$ from -3.42 to 4.36 J/mm leads to an increase in pore size. However, in the present study, it was found that a further increase in the $\Delta HI_{Relative}$, i.e., an increase from 4.36 to 12.44 J/mm, reduces the pore size, where the BMM defect also starts to occur. This raises the question of why pores with a size below the pore I.U.B. (0.2 mm diameter) form at both negative and positive deviations in heat input.

Table 4.4: Summary of the LWB principle and HI-related parameters.

C_i=1,2,3 ,...,n	<i>P</i> (kW)	<i>TS</i> (mm/s)	<i>WFS</i> (mm/s)	<i>HI</i>_{Nominal} (J/mm)	<i>HI</i>_{Wire} (J/mm)	<i>HI</i>_{Relative} (J/mm)	ΔHI_{Relative} (J/mm)
C1	2.5	50	60	50	41.67	22.72	−9.27
C2	4	80	73	50	54.79	26.14	−5.85
C3	4	65	85	61.54	47.06	26.66	−5.33
C4	4	80	60	50	66.67	28.57	−3.42
C5	4	50	85	80	47.06	29.63	−2.37
C6	4	65	60	61.54	66.67	32	0
C7	4	50	73	80	54.79	32.52	0.52
C8	4	50	60	80	66.67	36.36	4.36
C9	4	50	40	80	100	44.44	12.44
C10	5	50	60	100	83.33	45.45	13.45
C11	6	50	60	120	100	54.54	22.54

At lower heat inputs, when $\Delta HI_{Relative}$ is -3.42 J/mm (C_4), the laser energy is insufficient to fully melt the base metal and filler material, resulting in incomplete wetting of the joint surfaces. This inadequate wetting leads to formation of voids that appear as pores within the braze/substrate interface, as the molten filler metal fails to flow and integrate smoothly across the joint interface. These small voids/pores are formed due to the high viscosity of the molten

material, which prevents it from uniform spreading, a phenomenon exacerbated by the limited energy available for fusion. As the heat input increases, so does the energy transferred to the joint. Initially, the increased heat aids in reducing pore size by facilitating better wetting and flow of the molten metals leading to the defect-free condition (C_6). Further increases in the heat input, up to $\Delta HI_{Relative} = 4.36 \text{ J/mm}$ (C_8), can prove counterproductive in the elimination of pores. Excessive energy input increases Zn vaporization, leading to the formation of large gas bubbles. The cooling rate at this level of heat input is high enough to prevent these bubbles from escaping the molten pool, resulting in pores with sizes exceeding the pore I.U.B. Interestingly, beyond a critical heat input I.U.B at $\Delta HI_{Relative} = 12.44 \text{ J/mm}$ (C_9), there is a paradoxical reduction in pore size. The pores at this level remain below the pore I.U.B, similar to those formed at $\Delta HI_{Relative} = -3.42 \text{ J/mm}$ (C_4), but they form through a different mechanism. At $\Delta HI_{Relative} = 12.44 \text{ J/mm}$ (C_9), the excessive heat input melts not only the filler material, but also the base metal, altering the braze molten pool dynamics. While Zn vaporization and gas entrapment still occur, part of the energy is consumed by partial BMM formation. This leads to a balanced condition, where both the pores and BMM stay below the I.U.B. of both. Notably, partial base metal mixing introduces Fe-rich islands within the braze, which enhances mechanical performance, as shown in a previous study [195].

Excessive heat input does not always result in progressive pore formation. At $\Delta HI_{Relative} = 22.54 \text{ J/mm}$ (C_{11}), pore formation is suppressed due to several factors. The increased energy significantly reduces the viscosity of the molten filler, promoting complete substrate wetting. While Zn vaporizes progressively, the prolonged molten pool lifetime and reduced cooling rate provide gas bubbles with enough time to escape, minimizing gas entrapment. Additionally, dilution of the Zn coating reduces gas generation at the interface, further contributing to pore suppression. However, despite the elimination of pores, this high heat input causes excessive BMM, surpassing its I.U.B. and creating a defective condition.

4.4.3 Full-scale Diagnosing of Defect Distribution within the Braze Seam using μ CT

Among the various defect types, including LoA, pore, and BMM that formed in the brazing seam (as identified through MT and discussed above), the pore defect demonstrated a wide range of size variations under the different brazing conditions within the developed AI-friendly window. LWB samples with pore defects are ideal candidates for μ CT analysis due to their high frequency and the significant contrast between the pores (or voids) and the surrounding braze matrix in the μ CT scans.

Four LWB conditions with varying $\Delta HI_{Relative}$, that included a defect-free along with three pore conditions (both below and above the pore I.U.B. (0.2 mm diameter)), were selected for the pore analysis using the μ CT results. Fig. 4.12a-d illustrates 2 cm sections of the μ CT scanned results of actual braze seams generated under the different $\Delta HI_{Relative}$ conditions selected. To reveal the pore distribution along the corresponding braze seams, reduced opacity braze seams with pore distribution are depicted in Fig. 4.12e-h. Fig. 4.13 shows the corresponding distributions of the pore volume along the braze seam, measured from the top surface plane of the braze seam. The color coding in Fig. 4.12 and 4.13 represents the variations in the volume of pores. Pores of varying volumes and morphologies are distributed along the braze seam in all the selected samples. Notably, even the condition designated as defect-free (C_6) based on the MT results contains a variety of pores, although their size lay far below the pore I.U.B. (0.2 mm diameter), as shown in Fig. 4.13c. In LWB conditions with a pore size below the I.U.B. based on the MT results (e.g., C_4 and C_9), the μ CT results shown in Fig. 4.13b reveal that the pores are more uniformly distributed compared to the condition with pores beyond the I.U.B. (C_8) (Fig. 4.13d). It should be noted that large pores in C_8 only appear every 2 cm along the braze seam, as revealed by μ CT.

To statistically analyze the variations seen in the volumes of the pores formed along the braze seam of each condition, the whisker plot of individual pore volume across four selected conditions (Fig. 4.14a) and their corresponding distribution histograms were examined (Fig. 4.14b-e). The boxes in Fig. 4.13a show the range of pore volumes encompassing the 5th to 95th

percentile pore volume. The whiskers in this plot indicates the 95% CI for the mean pore volume for each selected $\Delta HI_{Relative}$ condition. Outliers are included in the quantification, as the AI-friendly specifications account for even a single defect exceeding the pore I.U.B. (0.2 mm diameter). From Fig. 4.14a, the $\Delta HI_{Relative}=4.36$ J/mm condition, which consisted of pores above the I.U.B. based on the MT results, showed the highest mean pore volume ($1.57 \times 10^{-3} \text{ mm}^3$) compared to other conditions. The mean pore volume in the other conditions (including the defect-free condition), ranged between $0.39\text{-}0.60 \times 10^{-3} \text{ mm}^3$. From the histogram of pore volumes in Fig. 4.14b-e, it can be deduced that pores below the I.U.B. (0.2 mm) are distributed evenly with a similar range of pore volumes. In contrast, pores exceeding the I.U.B. are few and not distributed uniformly through the seam.

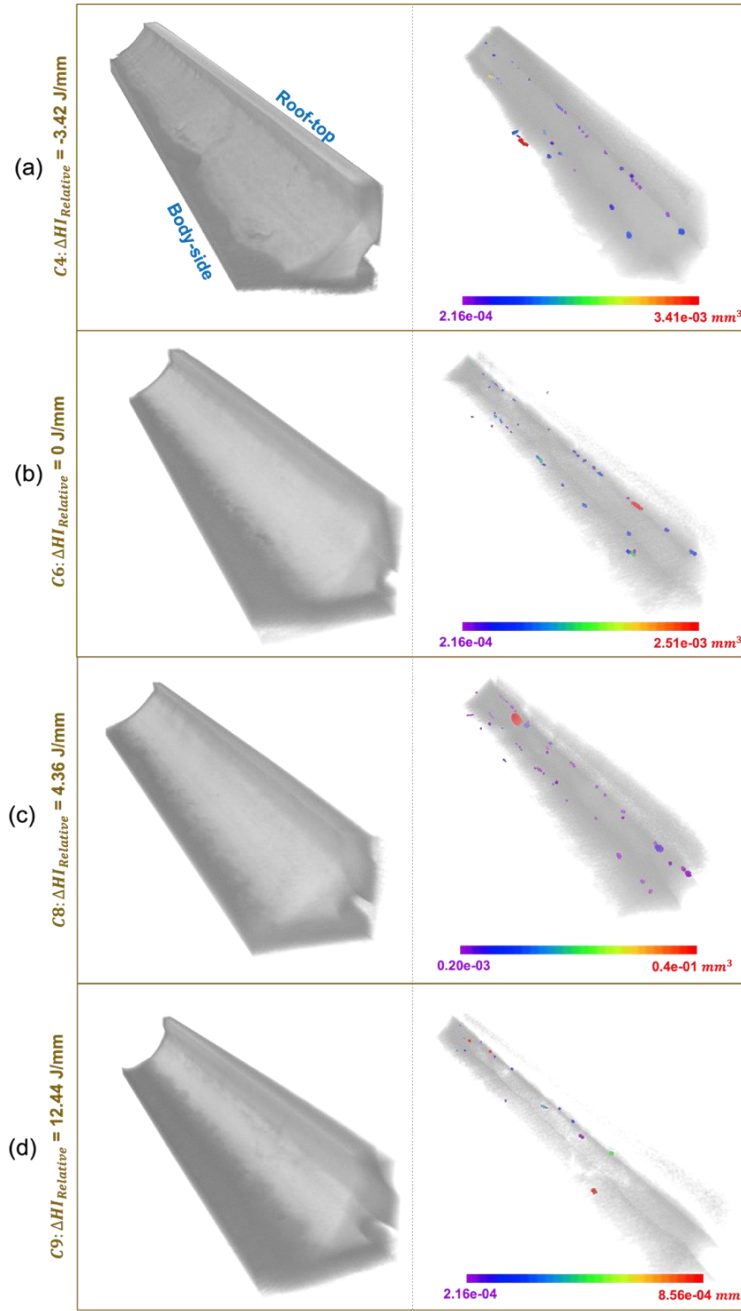


Fig. 4.12:(a-d) 3D visualization of representative 2 cm long braze seams of pore defect conditions with varying $\Delta HI_{Relative}$. The braze seam and substrates (roof-top and body-side parts) are distinguished by light grey and dark grey colors, respectively. (e-h) braze seams with reduced opacity, including pore distributions, in which the color-coding legend shows pore volume changes.

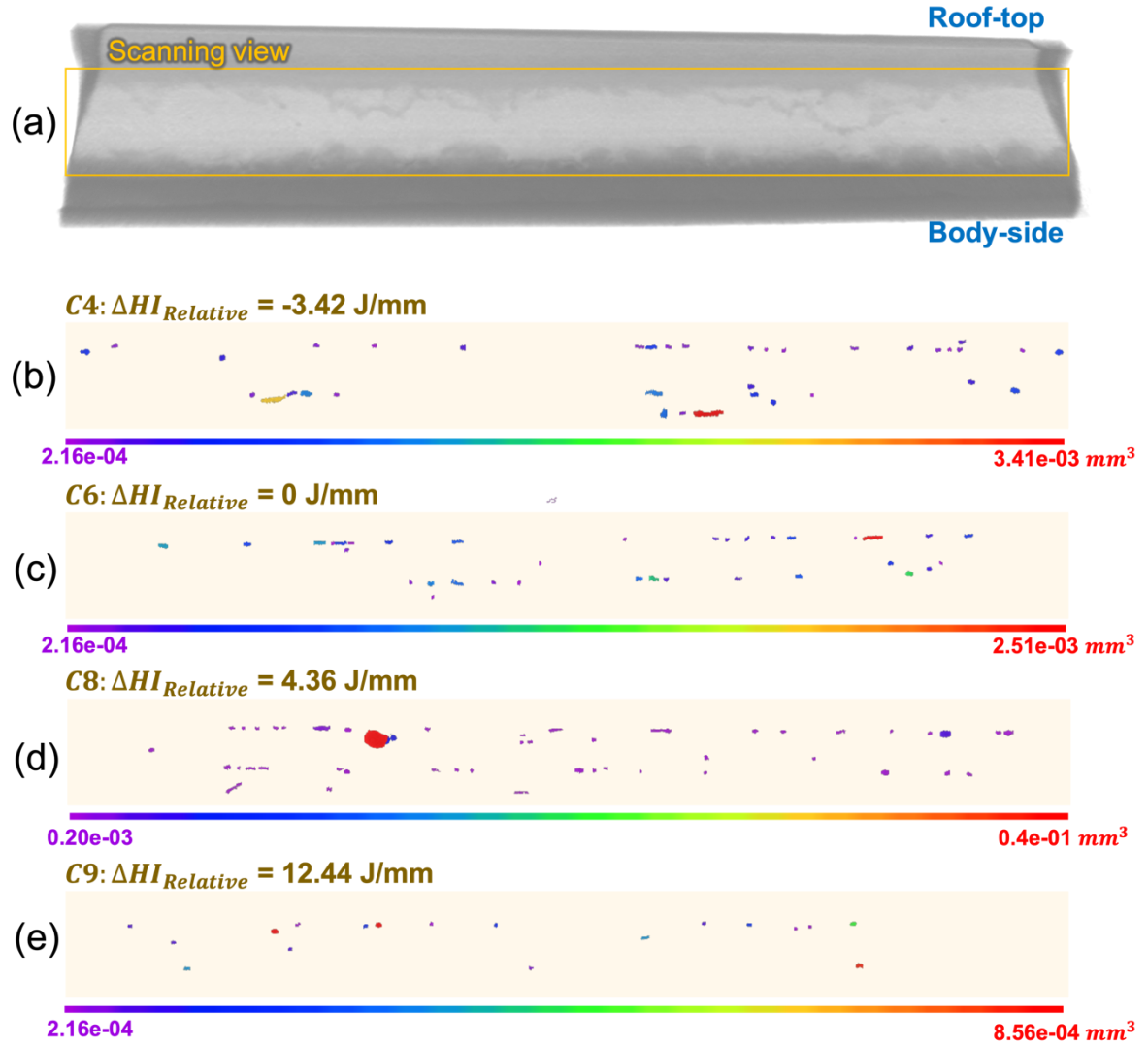


Fig. 4.13: (a) Top surface view of the μ CT scanned braze seam. The yellow box shows the view plane for the pore volume distribution in 2D along the braze seams with varying $\Delta HI_{Relative}$ as given in (b-e).

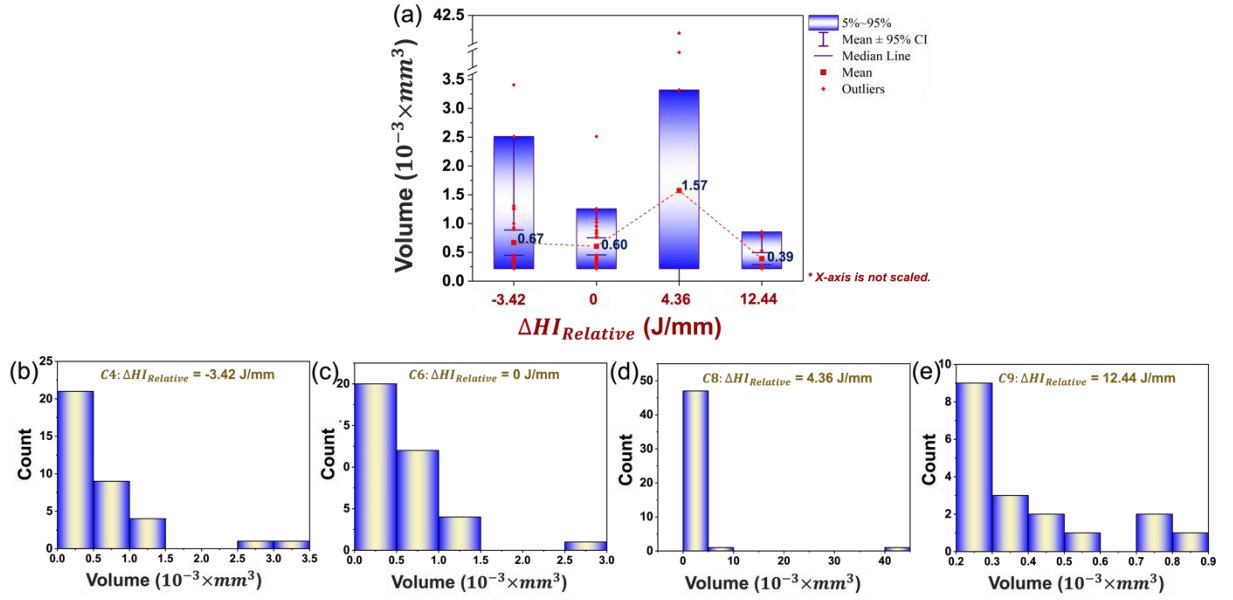


Fig. 4.14: (a) Whisker plot showing the quantification of the pore volume, and (b-e) their corresponding distribution histograms within the selected pore defect conditions with varying $\Delta HI_{Relative}$.

4.4.4 Generalization of Optimization

To generalize the optimization based on pore analysis, it is essential to directly compare primary defect characterization results from MT with complementary defect analysis using μ CT. Given that the pore diameter from the cross-sectional plane of the braze seams was included in the AI-friendly specifications, pore diameter measurements were also undertaken using μ CT to ensure compatibility with the optimization criteria and comparability with the MT results. For this purpose, the Feret length, a parameter used in the post-processing module in the DragonFly software, was applied to measure the distance between any two points along the mesh boundary at an arbitrary angle of the occupied space by the targeted object—here, each pore—during the segmentation and analysis of the μ CT data. For a single Feret, here pore, with irregular morphology, the lateral and longitudinal dimensions are defined as the minimum ($D_{min \text{ Feret}}$) and maximum ($D_{max \text{ Feret}}$) Feret diameters respectively. As the $D_{min \text{ Feret}}/D_{max \text{ Feret}}$ ratio approaches 1, the pore's morphology becomes more spherical; in this case, the pore diameter

is equal to both the $D_{\min. \text{ Feret}}$ and $D_{\max \text{ Feret}}$. According to Fig. 4.13b-e, most pores with non-spherical morphology are elongated along the length of the braze seam, rather than in the lateral direction. Therefore, $D_{\min \text{ Feret}}$ is characterized as the pore diameter of interest in the AI-friendly specifications in the context of this study, determined from the cross-sectional plane of the braze seams, making it comparable with the MT results. Fig. 4.15 illustrates the distribution of the pore diameter along the braze seam for each selected $\Delta HI_{\text{Relative}}$ condition, accompanied by the statistical analysis of the pore diameter presented in Fig. 4.16 (similar to the pore volume analysis in Fig. 4.13 and 4.14). From Fig. 4.16a, it is evident that pores in the condition with $\Delta HI_{\text{Relative}}=4.36 \text{ J/mm}$ have the highest mean diameter with $0.09 \pm 0.05 \text{ mm}$ compared to the other pore conditions. The mean pore diameter in the rest of the pore conditions, as well as the defect-free condition, are well below $0.08 \pm 0.01 \text{ mm}$. The evolution of the pore diameter between the selected LWB conditions in Fig. 4.16a follows a similar trend to that of pore volume in Fig. 4.14a, in which the condition with $\Delta HI_{\text{Relative}}=4.36 \text{ J/mm}$ is considered as the condition with largest pores in terms of both the pore diameter and pore volume. It should be noted that the difference between the pore size, in the condition with $\Delta HI_{\text{Relative}}=4.36 \text{ J/mm}$, is sharper in terms of the pore volume relative to the pore diameter. This is because the pore volume accounts for the pore size in all directions, while the pore diameter only compares its cross-sectional size. The MT results showed that the largest pore diameter detected within the selected conditions was 0.256 mm , while μCT revealed the presence of larger pores with a maximum diameter of 0.38 mm . Moreover, the pore diameter distribution for conditions with a $\Delta HI_{\text{Relative}}$ of -3.42 J/mm (Fig. 4.16b; braze seam with pores below the I.U.B. detected in MT) and 0 J/mm (Fig. 4.16c; defect-free condition from MT) indicate that the pore diameters range from 0.06 mm to 0.12 mm and with approximately symmetrical bell-shaped distributions, suggesting that they are closer to a normal distribution within the presented 2 cm scans. In contrast, the other two conditions, $\Delta HI_{\text{Relative}}$ of 4.36 J/mm and 12.44 J/mm (Fig. 4.16d and e), show skewed distributions, indicating an uneven spread of pore sizes, which suggests localized variations in the pore formation rather than a normal distribution. These results, thus,

recommend adding prudence to the optimization of any defect-free condition, such as C₆ with $\Delta HI_{Relative}=0$ J/mm, to account for the possibility of pores forming below the I.U.B.

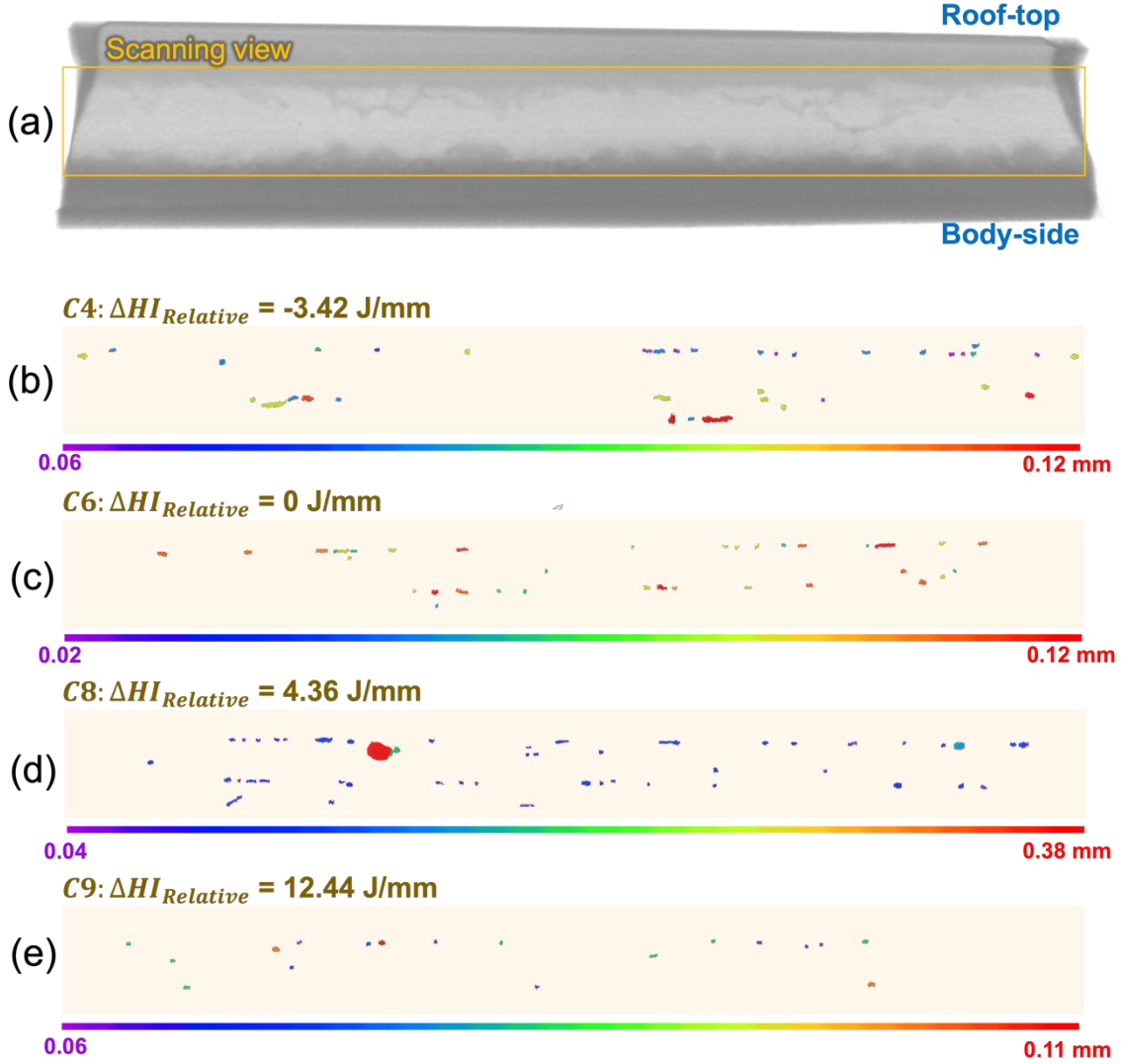


Fig. 4.15: (a) Top surface view of the μ CT scanned braze seam shows the reference plane for 2D pore visualization. (b-e) pore diameter distribution along the braze seam of the pore defect conditions with varying $\Delta HI_{Relative}$.

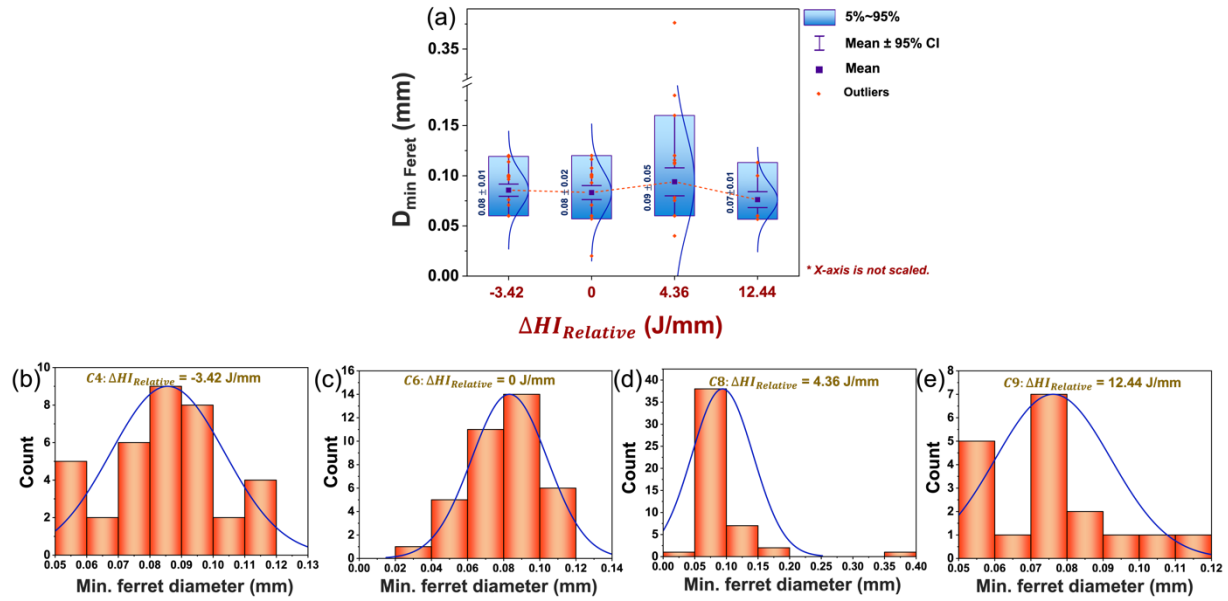


Fig. 4.16: (a) Whisker plot showing the quantification of the pore diameter ($D_{\min \text{ Feret}}$), and (b-e) their corresponding distribution histograms within the selected pore defect conditions with varying $\Delta HI_{\text{Relative}}$.

The μ CT characterization confirmed that the primary defect analysis undertaken using MT remains valid for LWB optimization; however, the extent of its deviation from the exact full-scale analysis by μ CT needs to be evaluated. For this purpose, a graph showing the correlation between the pore diameter and $\Delta HI_{\text{Relative}}$ values of the four selected conditions (pore and defect-free), for both the MT and μ CT results, is depicted in Fig. 4.17. The metallography results are gathered as maximum (max), mean, and minimum (min) discrete linear curves of four points. The μ CT results are shown by a purple polygonal domain that is defined based on considering the pore diameters within the lower 5% and upper 95% boundaries. Comparing the maximum pore diameter in the MT results and the upper 95% of pore diameters in the μ CT, there is a deviation of approximately +0.12 mm in the MT results for both the defect-free braze joints and those with pores exceeding the I.U.B. conditions, respectively with $\Delta HI_{\text{Relative}}$ of 0 and 4.36 J/mm. Despite this deviation, it did not influence the classification of the pores according to the defect specifications. However, even in the conditions labeled as defect-free,

it is important to consider that pores below the I.U.B., with diameters ranging from 0.02 to 0.12 mm, may still exist. Minimal deviation between the MT and μ CT results, which did not alter the $\Delta HI_{Relative}$ boundaries for defining the various pore defect braze conditions, ensures consistency in the classification of other defect conditions, such as LoA and BMM, for which quantification was not feasible using μ CT.

Comparatively μ CT scanning has so far been implemented to detect sub-surface defects in joints produced by laser welding [203], resistance spot welding [204], and friction stir welding [75,205], primarily for enhancing quality control applications; these weld samples consisted of a single material and featured symmetrical, rectangular joint geometries, allowing pore analysis to be performed using the existing segmentation modules during post-processing. However, due to the challenges in pore segmentation of braze samples with dissimilarities in the geometry and material of the braze region and substrates, a full-scale demonstration of pore formation evolution along the braze seam had previously been overlooked. For the first time, this study could demonstrate that achieving highly accurate pore detection for braze joints is feasible through the use of semi-automatic segmentation. While semi-automatic segmentation of μ CT data to identify pores within the braze seam might be perceived as time-consuming, it is still more efficient compared to MT, given the significantly longer sample preparation time required for the latter. Also, with MT, approximately 80% of the braze seam remains uncharacterized due to the bulk of the sample required for cutting and mounting.

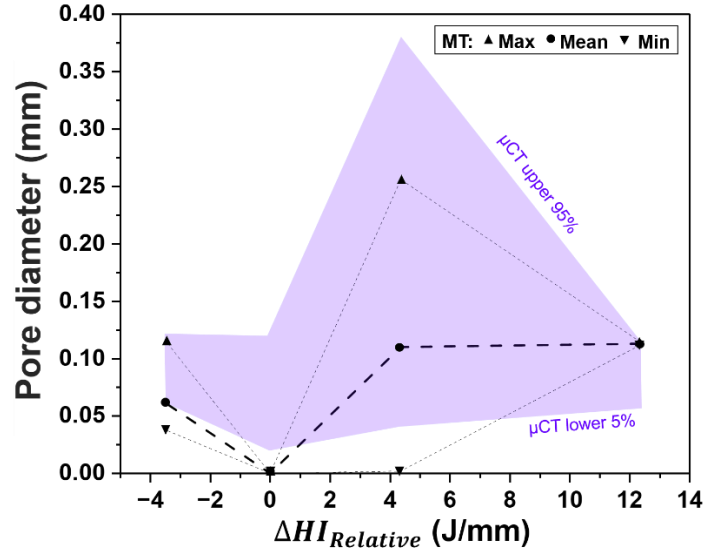


Fig. 4.17: Comparison of pore detection results characterized using MT and μCT methods.

This study is also the first to present the correlation between LWB process parameters, embedded in the $\Delta HI_{Relative}$ term, and pore evolution within the braze seam, validating process stability. The optimization strategy introduced—wire-adjusted heat input—demonstrates a scalable framework for other LWB applications involving similar substrates. The comprehensive μCT analysis conducted enabled 3D visualization and quantification of the pore evolution within the braze seam, providing valuable insights into the braze conditions and ensuring a thorough understanding of the process window's accuracy and reliability, which robustly validated the AI compatibility of braze joints produced using the proposed AI-friendly process window. The 2D slice outputs from the μCT scans of each braze seam were fed into the AI database required for real-time NDT training.

4.4.5 Performance of AI for Joint Geometry Prediction from UT Data

Even so, MT is a destructive measure that only shows defects on a few planes and μCT is time consuming and expensive to carry out. For this purpose, AI-enhanced UT inspection, is needed to fully characterize the weld in a time-efficient and cost-effective manner, especially for

industrial applications. To establish an automated sizing, alongside the detection of any indications of pores, LoA, BMM or other undesirable traits, a number of constraints need to be addressed. 1) Ultrasound imaging must be capable of seeing the defect; 2) The defect must be labeled in μ CT data; 3) The UT and μ CT data must be aligned properly. Among brazing defects, pores present a technical challenge in the creation of such a training set as a result of their limited size, spherical shape, and the possibility of clustering. Ultrasonically, the size and shape of a pore limit the reflection amplitude such that mostly pores above the I.U.B. are expected to be visible. The smaller pore size requires a higher resolution of scanning along the direction of brazing to ensure detection. The evaluation of braze seams produced using the process window established in this work showed reliable signal feedback for defect detection and the geometry of the braze. Fig. 4.18 presents the UT signal feedback from this window, compared with the corresponding μ CT 2D slice in a pore-containing region of the braze, an example of the most challenging case for UT detection. The UT image shows an excellent visualization of the pore by ultrasonic means, and it perfectly matches with μ CT data. It should be noted that the braze seams of all conditions, C_1 to C_{11} , have shown perfect fit for establishing AI-enhanced UT. Using rigid body image registration, these joints were then matched to within 0.1 mm of each other to allow for the creation of a baseline training set.

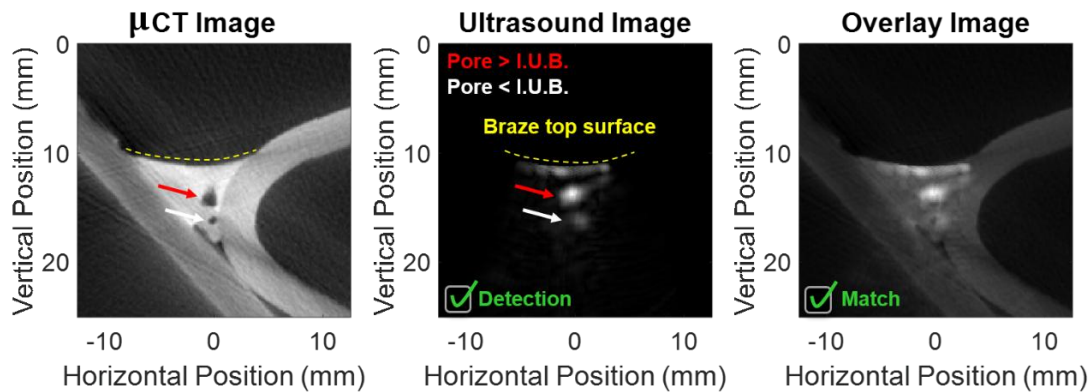


Fig. 4.18: An image of a pore condition (C_8) that is visible in both UT and μ CT imaging shows a perfect match. The visibility of the bottom surface is limited, as the pore obstructs the ultrasound path.

The semantic segmentation approach was conducted employing deep learning that was trained using the database derived from μ CT scanning of the LWB joints with various defect conditions within the resulting window. Fig. 4.19 illustrates an example from condition C_8 , showcasing the processing steps for both μ CT-based ground truth extraction and UT-based evaluation using semantic segmentation AI. The μ CT scans are processed to generate ground truth masks that delineate key features such as the joint, substrate, pores, and LoA. These masks provide detailed geometric data for the joint and substrate, which are aggregated and measured over a defined window of μ CT images. The UT data, including subsurface and surface scans, is then fed into a trained semantic segmentation model. This model predicts corresponding feature masks, and the resulting data is used to derive measurements for evaluation and comparison.

Given only the UT data, the semantic segmentation AI was able to identify joint geometry and substrate regions that strongly reflect those in the ground truth that was extracted from μ CT. The semantic segmentation approach, employing deep learning, trained using the database derived from μ CT scanning of the selected LWB joints, proved to be a reliable solution to extract actionable information about the joint geometry from the UT data. The semantic segmentation approach was used for this study over other possible options because it had the potential to produce high-resolution fine-grained mask outputs from which measurements could be straightforwardly derived. Conducting a more direct regression of the measurements given the UT data is possible, but the segmentation step provides an intermediate representation of the data that can be easily inspected by a human operator. This can be extremely beneficial, especially in cases where the extracted measurements exhibit variance along a joint or are outside of expected ranges. Additionally, the intermediate mask output can enhance the trustworthiness of the AI-based predictions, which is especially important in the context of NDT applications, as mask visualization can easily reveal biases in predictions, clear mismatch between input UT data and output masks, as well as catastrophic failures (e.g., due to ultrasonic hardware or electronics failures, vastly out-of-domain inputs, etc.). Although the

semantic segmentation approach was proven to reliably extract joint geometries, the identification of pore, LoA, and BMM defect indications is relegated to future work.

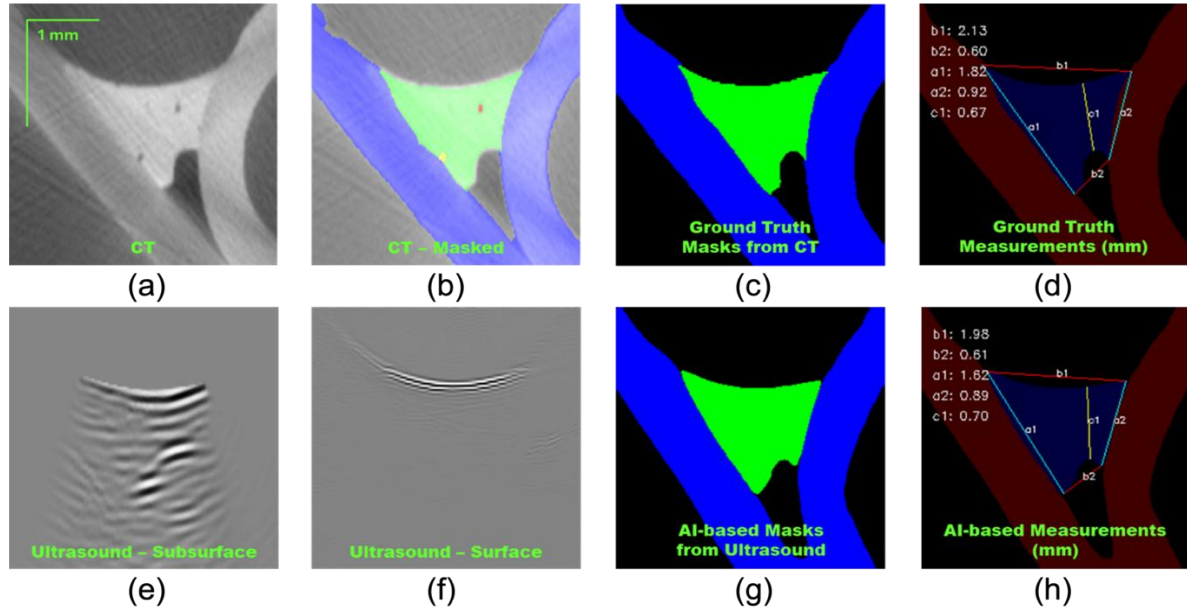


Fig. 4.19: (a-d) example from condition C_8 of processing of μ CT for ground truth extraction, and (e-h) UT using semantic segmentation AI for joint evaluation. Original μ CT scan (a) is processed (b) to extract ground truth masks for joint (green), substrate (blue), pores (red), and lack-of-adhesion (yellow). Joint (green) and substrate (blue) geometries from μ CT are then aggregated from a window of μ CTs (c) and measured (d). Ultrasonic subsurface (e) and surface (f) scans are given to the trained segmentation model as inputs, and model outputs masks (g). Measurements are then derived from output mask data (h).

The AI mask outputs from UT data aligned strongly with μ CT-based ground truth masks, which enabled reliable and straightforward measurement extraction from individual UT scans. Acquiring and subsequently processing the UT scans taken consecutively, either in real-time as they are acquired or offline, enables the automatic extraction of measurements at high sampling rates along LWB joints as exemplified for condition C_8 in Fig. 4.20. In this work, the UT sampling was conducted at 0.1 mm intervals along the joint, but in practice the movement of the UT probe can be modulated based on the needs of the user. Additionally, with this approach, extracted measurements can automatically be subjected to joint acceptance criteria downstream, defined by users according to their needs or standards. Integration of the

ultrasonic probe and electronics, as well as AI-driving software and the necessary computing hardware, into a robotic inspection system enables a fully automated ultrasonic NDT solution for LWB joints. In condition C8 (Fig. 4.20), there was a slight bias of the predictions for a1 (left leg in Fig. 4.19) and b1 (front face in Fig. 4.19) compared to ground truth, but this was not consistently observed in other samples. This is possibly due to the relatively small size of the dataset – 20 LWB joints each with 900 datasets – leading to misjudgment by the AI either on held-out samples during nested cross-validation or on regions of the joints demonstrating particular conditions, and may be rectified by scaling up the approach for data collection. Nonetheless, the extracted measurements provide a reliable judgement of joint geometry across the joint and demonstrate that it is possible to semi or fully automate a complete LWB joint analysis using ultrasound supported by AI. To further validate the approach, probability-of-detection and probability-of-false alarm analyses based on extracted measurements and joint acceptance criteria should be conducted in future work.

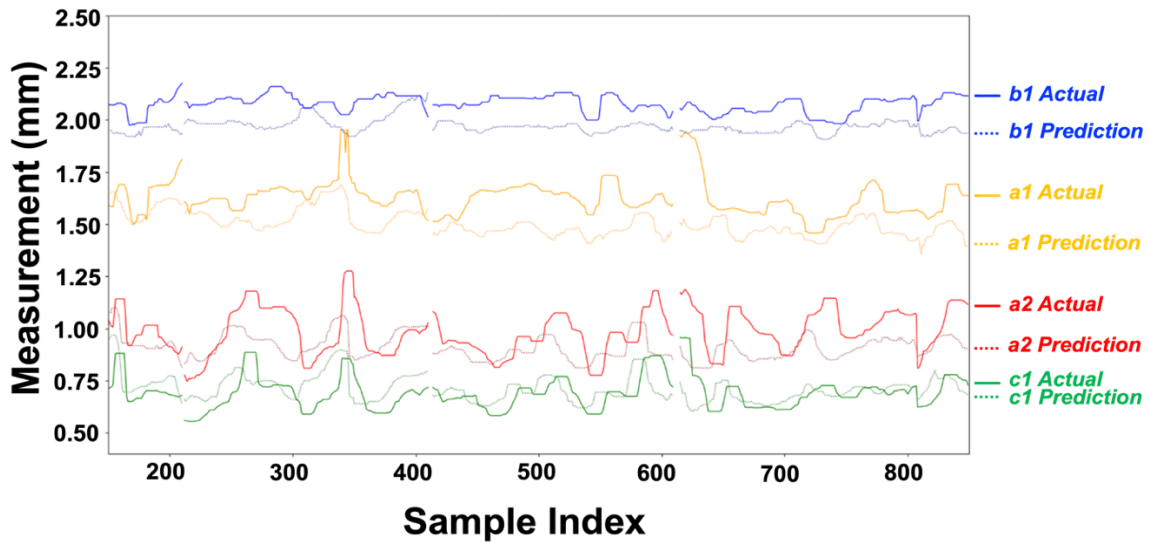


Fig. 4.20: Joint geometry measurements extracted from μ CT data ("Actual"), as well as UT data using segmentation AI ("Prediction"), along the length of sample C8. All joints yielded 1000 UT samples, but first and last 150 measurements were excluded, and gaps in data existed at sample indices 200, 400, 600, and 800, due to edge effects in μ CT scanning.

The process window developed in this work produced a set of samples with sufficient consistency within the joints and variance across the joints to support the development of AI for automated high-performance prediction of joint geometry from UT data. The MT measurements demonstrated greater variance as there was a relatively small sample size ($n=4$ per joint sample; Table 4.4). Across the four cross-sections from each joint sample, c1 (throat height) and a2 (right leg) measurements were most consistent between the MT and μ CT, while a1 (left leg) and b1 (front face) measurements showed greater variance in MT and greater discrepancy between MT and μ CT (Fig. 4.21a). There was very little discrepancy between the measurements derived from μ CT and UT using the AI approach (Fig. 4.21b), which was expected as the AI was trained to model a function from UT data inputs to mask outputs that were derived from μ CT scans. This demonstrates that the performance of the semantic segmentation AI, using a supervised deep learning approach, highly depends on the quality of its ground truth data (in this case, masks derived from μ CT). The developed process window produced samples with MT a1 (left leg) measurements ranging from 1.330 mm to 2.270 mm, a2 (right leg) measurements ranging from 0.770 mm to 1.170 mm, b1 (front face) measurements ranging from 2.090 mm to 2.650 mm, and c1 (throat height) measurements ranging from 0.590 mm to 0.830 mm, while standard error of the MT measurements ranged from 6.02% to 34.70% overall. The low variance within a LWB joint demonstrated the stability of the identified process parameters, and this contributed to the amenability of the joint to be inspected, including ultrasonically. Additionally, subtle variance along the joint, including minor changes in joint geometry as well as localized presence of pores and other purposely-induced defects, contributed to the robustness and coverage of the resultant ultrasonic dataset, and this helped to reduce epistemic uncertainty in the AI development process. The broad range of measurements and geometric variations across the manufactured joints established control over the process, which was afforded by the developed process window. This was crucial for maximizing ultrasonic dataset coverage that further reduced epistemic uncertainty in the AI development process. Ultimately, this work demonstrates a process window which is ideal for

creating a dataset, which can be used for the development of an end-to-end AI system to automate the ultrasonic inspection and evaluation of LWB joints.

Table 4.5: Extracted measurements from the MT, CT, and AI on the UT data, with standard error (SE) calculated using 95% CI.

Measurement	Condition Name	MT (mm)	MT SE (mm)	CT (mm)	CT SE (mm)	AI (mm)	AI SE (mm)
a1	C4	1.33	0.28	1.167	0.007	1.176	0.005
	C6	1.83	0.23	1.800	0.008	1.673	0.007
	C8	2.11	0.17	1.617	0.008	1.507	0.007
	C9	2.27	0.43	1.353	0.007	1.382	0.008
a2	C4	0.77	0.10	0.657	0.004	0.690	0.004
	C6	1.11	0.16	1.248	0.005	1.176	0.005
	C8	1.16	0.20	0.985	0.009	0.915	0.006
	C9	1.17	0.37	0.866	0.005	0.771	0.006
b1	C4	2.09	0.25	1.696	0.006	1.640	0.004
	C6	2.41	0.22	2.297	0.006	2.179	0.005
	C8	2.50	0.34	2.062	0.005	1.968	0.003
	C9	2.65	0.51	1.607	0.006	1.663	0.004
c1	C4	0.59	0.09	0.487	0.002	0.553	0.002
	C6	0.72	0.25	0.962	0.005	0.966	0.005
	C8	0.83	0.05	0.692	0.007	0.717	0.006
	C9	0.78	0.26	0.587	0.004	0.593	0.005

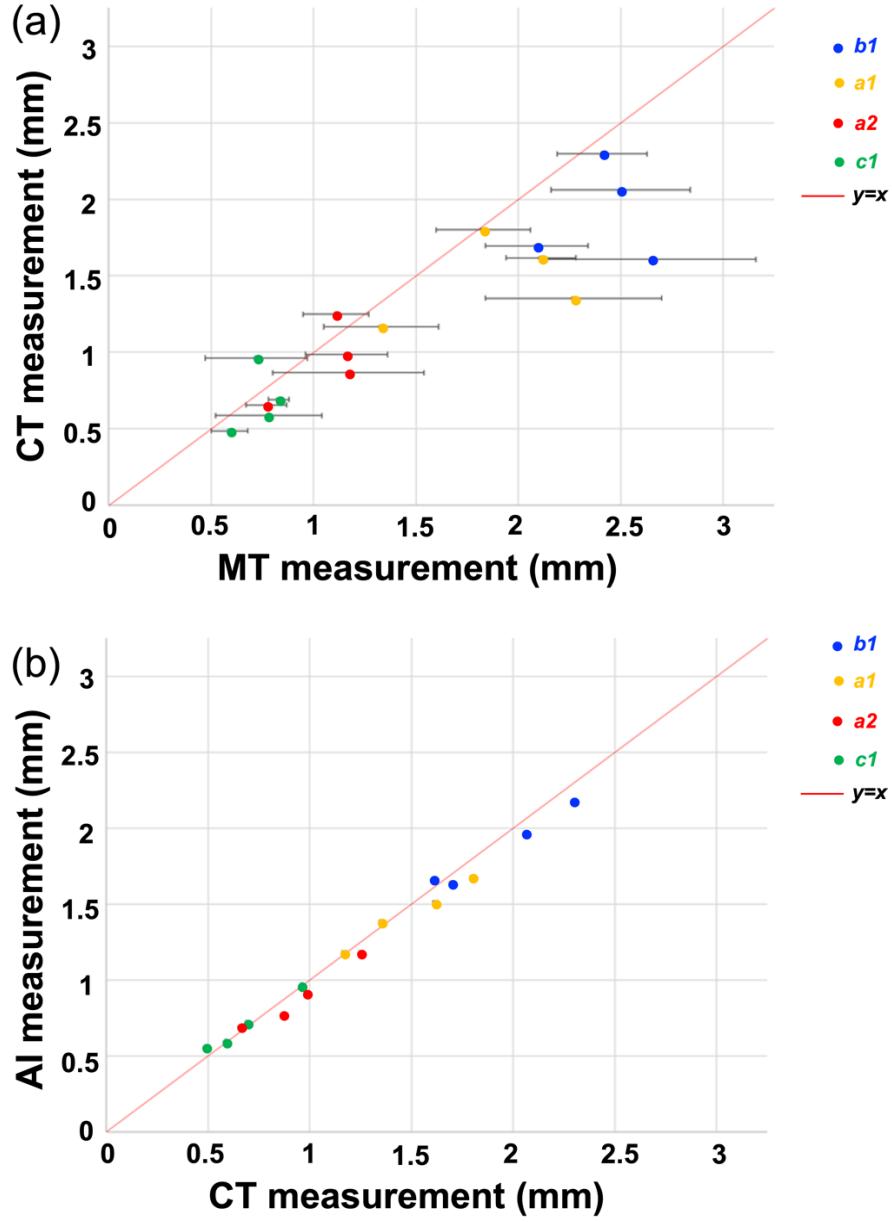


Fig. 4.21: Comparison of MT and μ CT measurements (a) as well as μ CT and AI-based measurements from UT (b). Both vertical and horizontal error bars are shown on both graphs, but errors are too small to visualize for μ CT and AI outputs due to both high sample size and measurements consistency.

4.5 Summary

This study highlighted advancements in laser-based manufacturing, focusing on process optimization and defect control in robotic LWB. The integration of AI played a key role in automating defect detection and joint geometry prediction from UT data at a real-time. Advanced characterization with μ CT scanning provided essential ground truth for training AI models and understanding defect evolution. These contributions paved the way for AI-driven quality assurance frameworks, improving the reliability and automation of laser processing in industrial applications. The main conclusions can be listed as follows:

- 1- A custom-designed process window was established which is capable of producing braze seams with various internal defect classes required for AI training of an industrial-level real-time ultrasonic technique for defect detection.
- 2- The conventional HI, $HI_{Nominal}$, was found to only show correlations with formed brazing defects at a constant WFS. Brazes made at a changing WFS were not comparable to conditions optimized through varying $HI_{Nominal}$ with constant WFS. A novel wire-adjusted heat input approach was proposed to find a direct correlation between HI and defect formation.
- 3- The HI-related boundaries determined from the novel wire-adjusted heat input approach facilitated LWB optimization by distinguishing the required HI for producing a defect-free braze seam, as well as braze seams with defects such as: LoA, pores and BMM below and above the specified I.U.B. for each defect type.
- 4- A full-scale defect analysis via μ CT scanning along the entire length of the braze seams (covering 2 cm per scan) of defect-free and pore conditions was conducted as a complementary defect characterization technique to the preliminary MT. It was confirmed that the presence of pores with the required diameter was highly repeatable along the braze seams of pore conditions; this proved the robustness of the boundary identified by the wire-adjusted heat input approach for each defect condition.
- 5- Comparing maximum pore diameter in MT and upper 95% of pore diameters in μ CT showed a small deviation of about +0.12 mm for the defect-free braze and braze conditions

containing pores above the pore I.U.B. (0.2 mm diameter). This deviation did not affect labeling the pore conditions according to the defect classes. However, even in a defect-free labeled condition, it should be taken into consideration that pores below the I.U.B., ranging in size from 0.02 to 0.12 mm in diameter, may still be present.

- 6- The process window effectively ensured stable and controlled LWB joint production, enabling reliable UT detection of the seam geometry and defect conditions.
- 7- The semantic segmentation approach, trained with deep learning on μ CT-derived datasets of LWB joints, accurately identified joint geometry and substrate regions using UT data. It showed strong alignment with μ CT-based ground truth, confirming its reliability. This approach proved to be a robust smart solution for extracting actionable information about joint geometry from UT data.

Chapter 5. Tailoring Intermetallic Compound Formation within Laser Weld Brazed Joints using a Novel Heat Input Approach⁴

5.1 Overview

The challenge of IMC embrittlement, resulting from thick IMC layers at the braze/substrate interface, and the lack of a clear strategy to manipulate IMC formation, has hindered the development of reliable LWB processes. This study addresses these challenges by introducing a novel approach to IMC manipulation during LWB of thin-gauge Zn-coated steel with Si-bronze filler on a double-flanged lap joint. By shifting IMC formation from the interface towards the interior region of the braze, the research mitigates embrittlement by developing a new IMC category, termed surrounded interface intermetallic compounds (SI-IMCs), distinct from traditional interface intermetallic compounds (I-IMCs). The study proposes a wire-adjusted heat input strategy to optimize brazing conditions, introducing a $HI_{Relative}$ that correlates with various brazing defects and IMC formation. The generic scientific contribution of this work lies in identifying a critical $HI_{Relative}$ value of 32 J/mm for defect-free brazing, with an additional I.U.B of 12.44 J/mm above this level to promote a high density of SI-IMCs, occupying up to $38.2 \pm 16.9\%$ of the braze cross-sectional area. These SI-IMCs, characterized by a shell-like Fe-Si layer and a bulky (Fe-rich)-Cu eutectic phase, enhance the mechanical performance of the brazed joints. Furthermore, this study reveals the novel role of Mn segregation in creating diffusion channels for Fe-Si IMC development, advancing the scientific understanding of IMC formation. Visualization through digital image correlation (DIC) during tensile testing showed that increasing the SI-IMC area fraction from $1.2 \pm 2.4\%$ to $38.2 \pm 16.9\%$ resulted in a 14% increase in tensile peak load and a 350% increase in ductility. This highlights the critical role of SI-IMCs in improving the strength and ductility of LWB joints,

⁴ This chapter is composed of a published research paper as outlined in the statement of contributions. **S. Akbarian**, A. Salandari-Rabori, S. Sarafan, P. Wanjara, J. Gholipour, A.R.H. Midawi, E. Biro, Tailoring Intermetallic Compound Formation within Laser Weld Brazed Joints using a Novel Heat Input Approach, J. Manuf. Process. Technol. 334 (2024) 118622.

offering a new pathway for enhancing the performance of brazed structures. A graphical abstract summarizing this chapter is presented in Fig. 5.1.

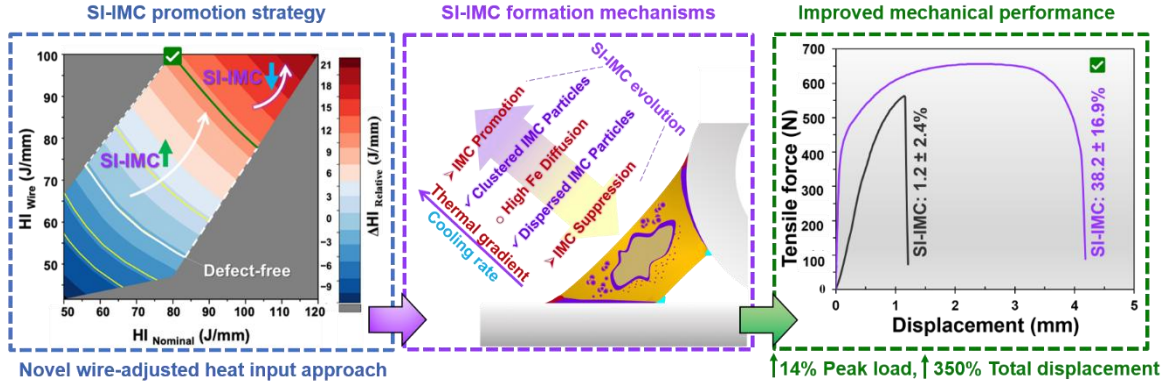


Fig. 5.1: Graphical abstract illustrating an overview of chapter 5.

5.2 Background

High formability-low strength IF steels are highly used steels for body-in-white and skin automotive applications [206]. However, IF steels suffer from poor corrosion resistance in-service. Therefore, they are typically coated with Zn, which offers a desirable degree of galvanic protection [5]. While Zn-coated steels offer enhanced corrosion resistance, the presence of Zn, and its associated IMCs, on the substrate surface adversely affects the post-weld properties of the final joints [9]. This occurs due to substantial Zn vaporization when subjected to conventional fusion welding processes. Zn vaporization results in brazing defects [42]; for instance, damage to the Zn protection around the joining area degrades the joint appearance and corrosion properties of the Zn-coating, as well as produces a lack of adhesion and porosity within the joint seam. These defects are of concerns as they can act as crack initiation sites that, in turn, can degrade the mechanical performance [41]. When assembling the exterior body-in-white structures with thin-gauge Zn-coated steel sheets, not only is the mechanical performance of the joints a major concern, but also achieving an attractive joint is crucial, especially for roof-top and decklid joints, which are visible to the end users [44,207]. To mitigate Zn-related damage, it is crucial to minimize Zn vaporization, which can be achieved by reducing the processing temperature for joining. LWB is a joining method with

lower temperatures than fusion welding processes, which can mitigate Zn vaporization damage, as well as limit substrate melting [10,11]. The reduced processing temperature of LWB comes from the lower heat input needed for joining due to the use of a filler wire with a T_m lower than the substrate(s).

Bonding in LWB joints is achieved by the melting of the filler and wetting of the substrates [208]. As a result, intermetallic compound phases form at the interface, which are the products of interfacial reactions between the molten filler and the substrate(s) during the wetting process [89]. However, IMC phases are brittle and can act as crack initiation sites, posing a considerable challenge to brazing applications [91]. Hence, unravelling IMC formation mechanisms and understanding their characteristics are critical, as controlling the IMC composition, thickness, and distribution is necessary to ensure joint reliability and life. IMC formation follows certain thermodynamic principles [175]. Only compounds with a lower ΔG will be stable as IMCs in the joint [94,95]. Therefore, compositions of the filler metal and substrate along with the processing temperature play a crucial role in IMC formation. In the joining of steels, a Cu-based brazing alloy with Si addition is commonly used for weld brazing applications [209,210]. Although IMCs are critical for joint performance, reported research on LWB development of Zn-coated-steel/Si-bronze joints is limited in the open literature. Li et al. [103] investigated the effect of laser energy distribution on the IMC layer thickness and distribution without any further identification of IMC phases and their influence on the mechanical performance of the joints. Kai et al. [211] showed that the dispersion of IMCs enhances the strength of LWB joints. However, the mechanisms of IMC formation were not thoroughly investigated. The studies concerning IMC in weld brazing have mainly investigated joints made using arc heat sources, such as the GMAB process [96,99].

It has been found that Fe-Si phases are the most common IMCs that form during the weld brazing process in general, where Fe diffuses from the substrate and reacts with Si in the filler alloy [175]. Both elements exhibit a tendency to form IMCs as per the Fe-Si phase diagram [176]. In GMAB [96] and LWB [211], two variants of Fe-Si IMCs, Fe_2Si and Fe_5Si_3 , with layer and granular morphologies, were found to be generated within the interface and adjacent

to the interface, respectively. In GMAB, it was found that the formation mechanism of granular IMC particles is IMC layer fragmentation [96]. Sarmast-Ghahfarokhi et al. [120] noted that a GMAB module with increased heat input can accelerate the density of certain precipitates with an irregular morphology within the braze matrix, leading to a composite-like microstructure. Such braze joints, compared to those with fewer precipitates, exhibited enhanced tensile peak load and elongation. However, it is still unclear whether the identified precipitates also exhibit particular properties/characteristics of the precipitates, and therefore may be formed by a different process. In comparative studies [98,99], it was reported that the density of such precipitates in GMAB is generally higher than in LWB, due to a larger heat input used in GMAB. Yet, it is not known if the high density of such precipitates or IMCs can be generated in LWB under increased heat input. However, it should be noted that Akbarian et al. [212] showed that using a laser heat source can alter the segregation behavior, causing Fe and Si to segregate towards the edge of the braze in accordance with high Zn accumulation in a GI coated steel during LWB, a phenomenon not observed with the arc heat source. Unlike GMAB, the mechanisms and kinetics of IMC formation in LWB have not been systematically investigated, despite their expected impact on the subsequent mechanical performance of LWB joints, as displayed in GMAB joints.

Previous studies highlighted the advantages of incorporating second phases within the braze to enhance mechanical performance in GMAB applications [31,120]. In comparison to GMAB, LWB offers nearly twice the joining speed along with a superior precision for intricate details, attributed to the higher energy density of the laser beam and its stability compared to the heat generated by an electric arc in arc welding processes [212]. Despite the evident benefits of using a laser over an arc in weld brazing processes, uncertainties persist regarding IMC formation mechanisms in LWB, underscoring an urgent need to address these gaps in the existing literature. It is important to note that IMCs are critical phases formed during laser weld brazing, which can significantly impact the mechanical properties of the resulting joint. However, there is a clear lack of knowledge regarding the formation mechanism, type, morphology and influence of these intermetallic phases, which are targeted in the current study

to be systematically assessed. The methodology used to address these gaps can pave the future path for other manufacturing methods, where the formation of intermetallic phases is inevitable and will greatly affect the resultant properties.

LWB of thin-gauge Zn-coated steel with Si-bronze filler material presents two significant academic challenges. The first challenge is the embrittlement caused by the formation of IMCs, which is often due to the thickening or brittle nature of IMCs, typically occurring at the braze/substrate interface. This embrittlement undermines the mechanical performance of the brazed joints. The second challenge is the lack of a clear strategy for manipulating IMC formation, necessary to reduce embrittlement, while maintaining joint integrity. To address these challenges, this study explores a novel strategy aimed at manipulating IMC formation by promoting IMCs in the interior of the braze rather than at the interface. A heat input-based approach is introduced to optimize defect levels and manage IMC distribution in the joints composed of GI-IF steels using Si-bronze filler wire for a double-flanged joint configuration used in real-world applications. Additionally, the study investigates the segregation behavior of constituent elements and examines how the resulting phases influence the overall mechanical properties of the joints. By addressing the challenges of IMC embrittlement and the need for an effective control strategy, this research provides critical insights into IMC manipulation and offers a framework for improving the performance of brazed structures through optimized LWB processes.

5.3 Materials and Methodology

5.3.1 Substrate and Brazing Filler Metal

The substrates used in this work were thin-gauge sheets of a GI-IF steel with a thickness of 0.65 ± 0.02 mm, a Zn coating weight of 50 g/m^2 at the top and bottom of the sheets and a chemical composition as presented in Table 5.1. A CuSi3Mn1 brazing filler wire, commercially known as Si-bronze, with a diameter size of 1.2 mm, was used for the LWB process. This is a conventional wire widely used in the automotive industry for brazing steels. The chemical composition of the filler wire is provided in Table 5.1. A double-flanged lap joint configuration

with the geometry shown in Fig. 5.2a and b was used for LWB of the roof joints in the present study using a custom-designed fixture, as shown in Fig. 5.2c.

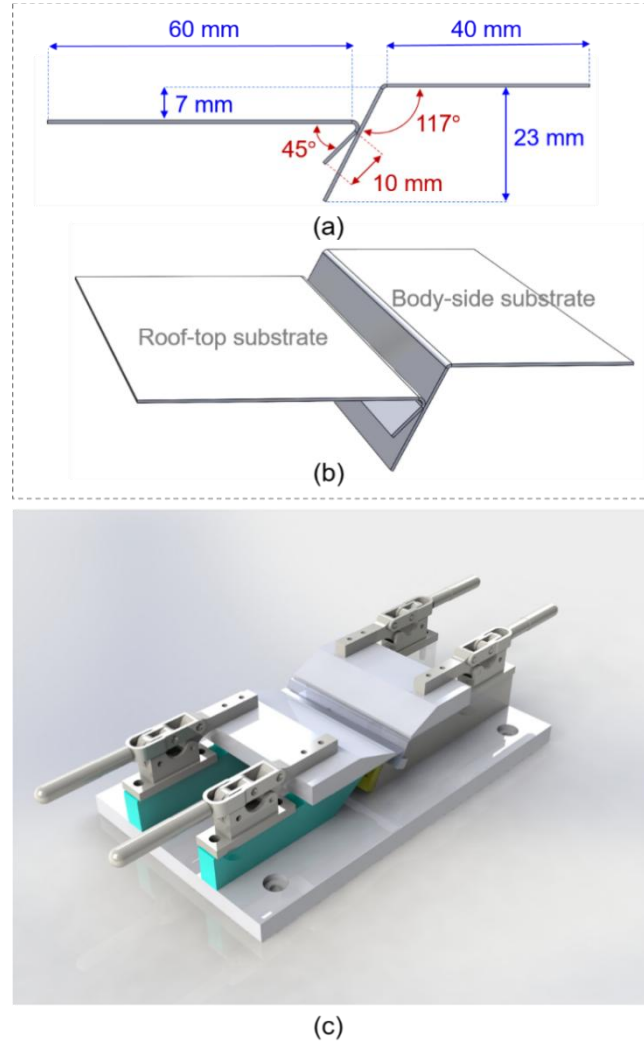


Fig. 5.2: Solid work models of LWB coupon for Roof joint configuration. (a) 2D side-view, (b) 3D, and (c) the custom-designed fixture.

Table 5.1: Chemical composition of the experimented materials. Weight percentages (wt.%) of GI-IF steel (substrate) and CuSi3Mn1 (filler) are included.

Material	Fe	Cu	C	Si	Mn	P	S	Al
GI-IF	Bal.	-	0.06	0.50	0.40	0.025	0.020	0.010
CuSi3Mn1	0.02	Bal.	-	2.90	0.80	-	-	-

5.3.2 LWB Process Setup

The LWB was conducted using an IPG Ytterbium-doped fiber laser with a maximum power capacity of 8 kW and a laser beam wavelength of 1070 ± 10 nm, delivered from a Robotic Hybrid Laser Welder (Lincoln Wayne-Trail/IPG) with multiple heads. The LWB parameters for the double-flanged lap joint configuration in Fig. 5.2 were optimized to produce brazes with a typical cross-sectional geometry for industrial practice, as specified in Fig. 5.3a, and flawless seam appearance by visual inspection of the top surface. In the present study, a laser head equipped with a cold wire feeder was used, which gave a laser beam diameter of 600 μm with a Gaussian beam profile. The beam was defocused 42 mm from the focal point to increase the laser beam spot size diameter to 3.2 mm on the workpiece. The lateral incidence angle of the laser beam was 90° , and the wire feed angle was kept at 5° , both relative to the surface of the substrate (x-y plane) as seen in Fig. 5.3b. This setup for positioning the filler wire in relation to the laser irradiation, as well as their interaction with the substrate, remained consistent across all brazing conditions. No shielding gas was used in this study so as to match common industrial practice. The LWB principal variables were P , TS , and WFS , that were selected within the ranges given in Fig. 5.3b, respectively 2-6 kW, 50-85 mm/s and 40-80 mm/s; these domains are typical of those used in manufacturing. An extensive design-of-experiments (36 runs) was conducted to optimize different sets of variable combinations for certain braze geometry and brazing defect conditions. Brazing defect conditions of interest are LoA, porosity, and BMM as proposed by the manufacturer. It is worth mentioning that the porosity and BMM imperfections were considered as brazing defects only when they were above I.U.B, which was set by manufacturer at 0.2 mm for both.

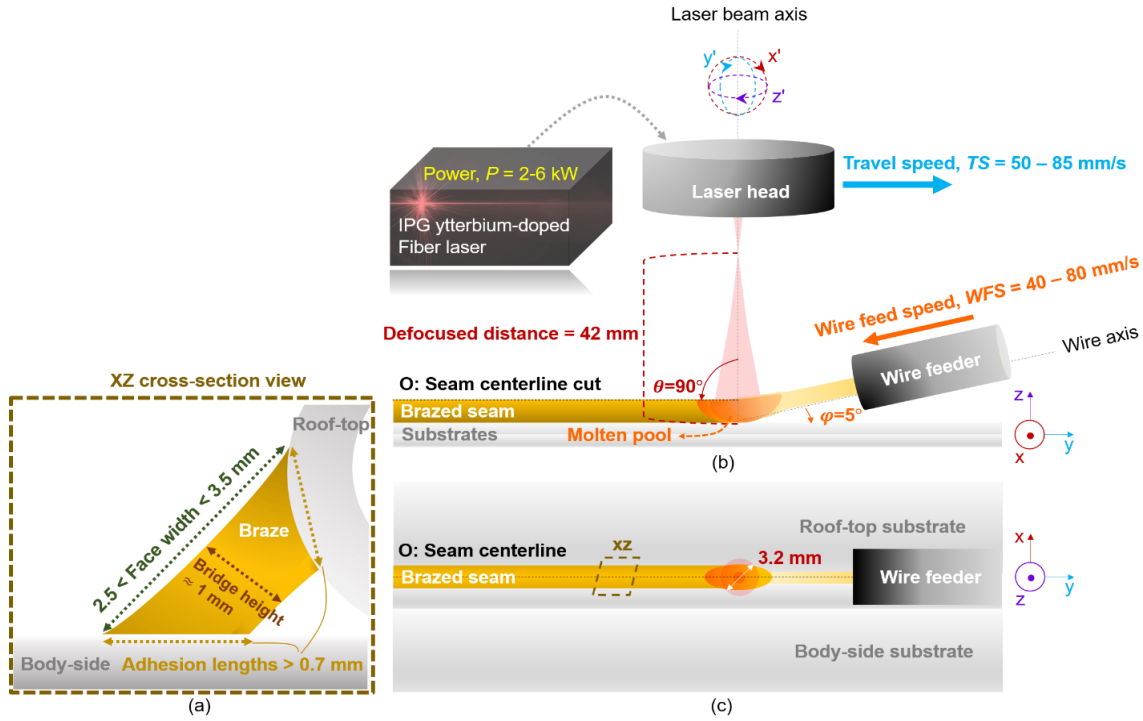


Fig. 5.3: Schematics of the LWB process. (a) typical braze geometry with industrially desired dimensional ranges; (b, c) LWB setup showing experimental constants for non-variable parameters including spot size diameter, defocused distance and laser/wire/substrate incidence angles. Experimental domains for principal variables such as laser P , TS , and WFS are included.

5.3.3 Characterization

The braze microstructure and geometry in each LWB condition were investigated by cross-sectioning the LWB seam perpendicular to the brazing direction. A highly-controlled polishing procedure was used on the four cross-sections extracted from each LWB seam, according to a metallographic method described in [212]. An optical Keyence VHX-7000 microscope was used to observe the polished brazed surfaces to provide a larger field of view for capturing the whole braze geometry of each cross-section.

The compositional distribution of IMC phases was examined using a JEOL JX-8230 electron probe micro-analyzer (EPMA) to generate high resolution compositional maps (down to step size of $0.3 \mu\text{m}$) for Fe, Cu, Si, Mn, and Zn, i.e., the substantial elements available in the

substrate and the filler metal. The composition of the substrate is predominately Fe and Zn (from the coating) with other elements added as alloying additions, and Cu, Si, and Mn are the majority constituents of the filler metal representing the brazing matrix.

5.3.4 Mechanical Testing

Brazing samples of interest were examined by tensile testing using a digital image correlation method to visualize and evaluate strain distribution on the sample during tensile deformation. Tensile samples were waterjet cut to the sub-size geometry according to ASTM E8 [213]. The tensile samples were prepared so that the tensile direction was perpendicular to the brazing direction, and the braze seam was located at the centre of the reduced width portion of the sample, as shown in the 3D computer-aided design (CAD) model given in Fig. 5.4. The geometry of tensile samples prepared from the roof configuration braze joint, as seen in Fig. 5.4b, can cause a complex loading situation due to the unlevelled surfaces of the roof-top and body-side with respect to each other. To correct the alignment of the sample and tensile grips, the samples were shimmed to the grip. A Tinius Olsen universal testing frame with a 10 kN maximum load cell capacity was used to pull the tensile samples in a low strain rate (LSR) loading condition at the cross-head speed of 1 mm/min. The strain distribution and concentrations were calculated using the Correlated Solutions Vic2D DIC software with a virtual strain gauge of 25 mm.

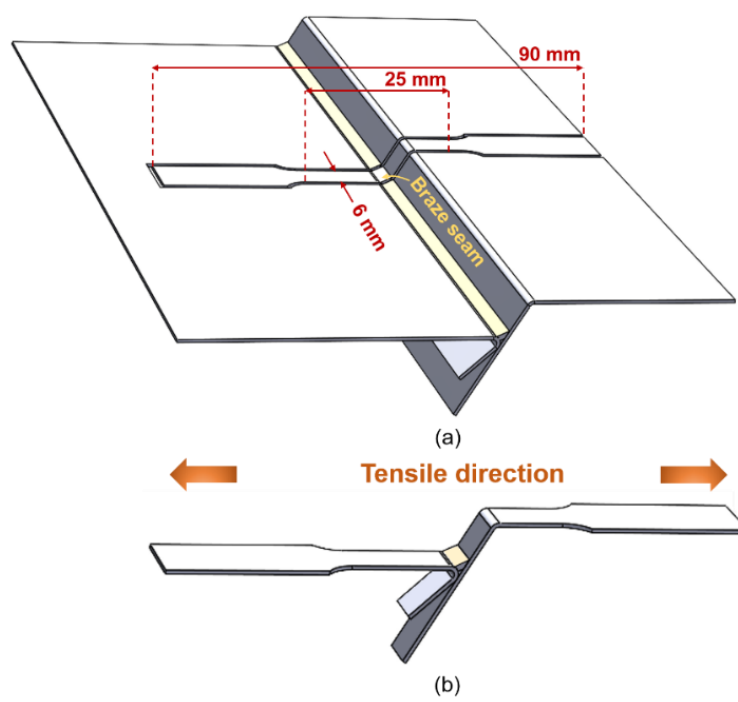


Fig. 5.4: CAD models of sample preparation for the tensile testing. (a) typical brazed coupon for the roof joint configuration used in this study, showing the cutting path of the waterjet to prepare one sub-size tensile sample; b) an extracted sample ready for tensile testing in the given direction.

5.4 Results and Discussion

5.4.1 IMC Manipulation Strategy

The three principal variables of LWB, namely, P , TS , and WFS , as varied within the given range in Fig. 5.3, play a decisive role in changing the heat input of the process. The formation of IMCs is highly dependent on the heat input of the LWB process due to the fact that IMC nucleation and growth are both diffusion-controlled phenomena affected by temperature changes [96]. In LWB, the $HI_{Nominal}$ is often defined as the P to TS ratio with a unit of Joule (J) per travel distance (as given in Equation 5.1 [65]). However, this conventional definition of heat input, i.e., $HI_{Nominal}$, cannot fully establish a pathway for manipulating the formation of IMCs during LWB, as the heat of the laser is not fully transferred to the base metal due to the addition of the filler wire. In the present study, to account for the necessary heat to melt the

filler wire during LWB, two newly developed heat input-related components, wire heat input (HI_{Wire}) and relative heat input ($HI_{Relative}$) are introduced, as noted in Equation 5.2 and Equation 5.3, respectively. The HI_{Wire} is defined as the P to WFS ratio, irrespective of the TS of the laser torch; this heat input is a simplification for the heat required to melt the filler wire under the assumption of considering laser as a stationary heat source. The $HI_{Relative}$, as given in Equation 5.3, determines the heat delivery to the laser/wire/substrate interaction zone at a relative process speed (Equation 5.4) as follows:

$$HI_{Nominal} = \frac{P}{TS} \quad (5.1)$$

$$HI_{Wire} = \frac{P}{WFS} \quad (5.2)$$

$$HI_{Relative} = \frac{P}{V_{Relative}} \quad (5.3)$$

$$V_{Relative} = TS + WFS \quad (5.4)$$

The units of all heat input-related components in Equations 5.1-5.3 are Joule per travel distance. It was found that each brazing condition associated with defects and IMC evolutions occurred at a certain value of $HI_{Relative}$. Such a correlation cannot be found by relying solely on $HI_{Nominal}$ and HI_{Wire} values. The contour map in Fig. 5.5 was graphed to provide a comprehensive overview on the correlations between the heat input-related components and the different brazing conditions. The contoured axis in Fig. 5.5a represents the deviation of the $HI_{Relative}$ value of each condition (X) with respect to the $HI_{Relative}$ value of the defect-free condition, as given in Equation 5.5, which was found to be equal to 32 J/mm in the present study. For the sake of elucidating each brazing condition comprehensively, a wire-adjusted heat input term, noted by $C_{i=1,2,3,...,n}$, is defined to collect all three heat input components in 3D space, as described in Equation 5.6.

$$\Delta HI_{Relative} = (HI_{Relative})_X - (HI_{Relative})_{Defect-free} \quad (5.5)$$

$$C_{i=1,2,3,...,n} : (HI_{Nominal}, HI_{Wire}, \Delta HI_{Relative})_X \quad (5.6)$$

According to Fig. 5.5a, $\Delta HI_{Relative}$ changes from $[\Delta HI_{Relative}]_{Min.} = -9$ J/mm to $[\Delta HI_{Relative}]_{Max.} = 23$ J/mm, in which each yellow line shows a specific brazing condition. As

$\Delta HI_{Relative}$ approaches its minimum values, LoA is a predominant brazing defect, while it shifts to the BMM defect once $\Delta HI_{Relative}$ reaches its maxima. In the lower $HI_{Relative}$ bounds, i.e., negative $HI_{Relative}$, the formation of pores with diameters below the critical I.U.B occurs. In the upper $HI_{Relative}$ area, i.e., positive $\Delta HI_{Relative}$, there is a variety in the pore and BMM defects, and IMC-like phases. At a $\Delta HI_{Relative}$ value of +4.36 J/mm, large pores with sizes greater than the critical I.U.B form in the braze, while with -3.42 J/mm, the pores are smaller in size (below the I.U.B). Comparing the wire-adjusted heat input terms of $C_1:(50,66.7,-3.42)_{Pore < I.U.B}$, $C_2:(61.5,66.7,0)_{Defect-free}$, $C_3:(80,66.7,4.36)_{Pore > I.U.B}$, and $C_4:(80,100,12.44)_{Pore, BMM < I.U.B}$, it can be well deduced that pore and BMM defects are controlled by $\Delta HI_{Relative}$.

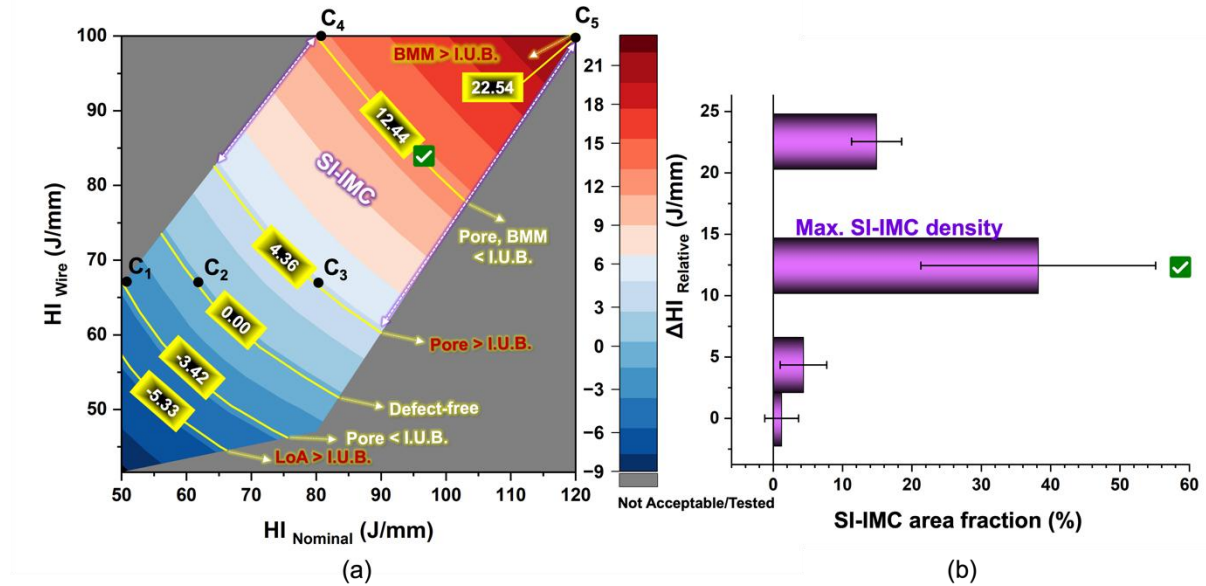


Fig. 5.5: LWB window for IMC manipulation. (a) Contoured LWB process map employing wire-adjusted heat input methodology, featuring $HI_{Nominal}$ and HI_{Wire} on the X and Y axes, respectively, with $\Delta HI_{Relative}$, which is the deviation from required $HI_{Relative}$ to achieve defect-free condition, represented by the Z contour lines. The defect conditions including LoA, pores and BMM, with above and below the I.U.B corresponding to each wire-adjusted heat input, are each annotated along the $\Delta HI_{Relative}$ boundaries. (b) Bar graph showing the SI-IMC density versus $\Delta HI_{Relative}$ values starting from zero (the defect-free condition). Error bars are calculated based on 95% CI.

Second phase regions dispersed within the braze region, as shown in Fig. 5.6a-d, can potentially be SI-IMC phases, rather than the I-IMC layers, which can typically form within

the braze/substrate interface. Formation of these potential SI-IMC phases is aligned with the purpose of shifting the IMC formation from the interface towards the interior region of the braze. In the context of the present study, these two terminologies, I-IMC and SI-IMC, have been introduced to differentiate various IMC types generated within various region of the brazed joint. The density of SI-IMCs was quantified by the area fraction of the braze, calculated from the average of four cross sections, with 95% confidence intervals applied. According to Fig. 5.5b, the increase in the SI-IMC fraction, which started to appear from C₂, is proportional to $\Delta HI_{Relative}$ and/or $HI_{Relative}$. However, the SI-IMC fraction does not reach its peak at the highest HI values at C₅, as the SI-IMCs tend to merge with the BMM defects. The maximum SI-IMC fraction with $38.2 \pm 16.9\%$ of area is achieved at C₄, where the $\Delta HI_{Relative}$ is not high enough to cause BMM above the I.U.B of being labelled as a defect. It is evident that achieving a significant fraction of SI-IMCs necessitates striking a balance between conditions where pore and BMM defects are occurring (with a size above the I.U.B), namely the C₃ and C₅ conditions, respectively. The meaningful correlation found between the principal variables and the different defect conditions, as well as IMC fractions within the braze matrix implies the accuracy of the proposed wire-adjusted heat input approach in this study.

Achieving a high density of SI-IMCs, in C₄, where the defects are eliminated in the braze, as evidenced in Fig. 5.6c, makes this condition worthy of further investigation and analysis. Promotion of IMCs within the interior region of braze rather than the interface aligns with the objective of this study to mitigate IMC embrittlement stemming from the braze/substrate interface; braze of C₄ condition allowing the analysis of the most severely promoted IMCs within the braze matrix. Such a high density of the second phase within the braze matrix has never been investigated in the previous studies on LWB, except for the formation of precipitates with smaller fraction in GMAB joints on Zn-coated steels at elevated heat input [99]. Thus, the present investigation will focus on comprehensively understanding the characteristics and properties of these phases, SI-IMCs, as elaborated in the subsequent section.

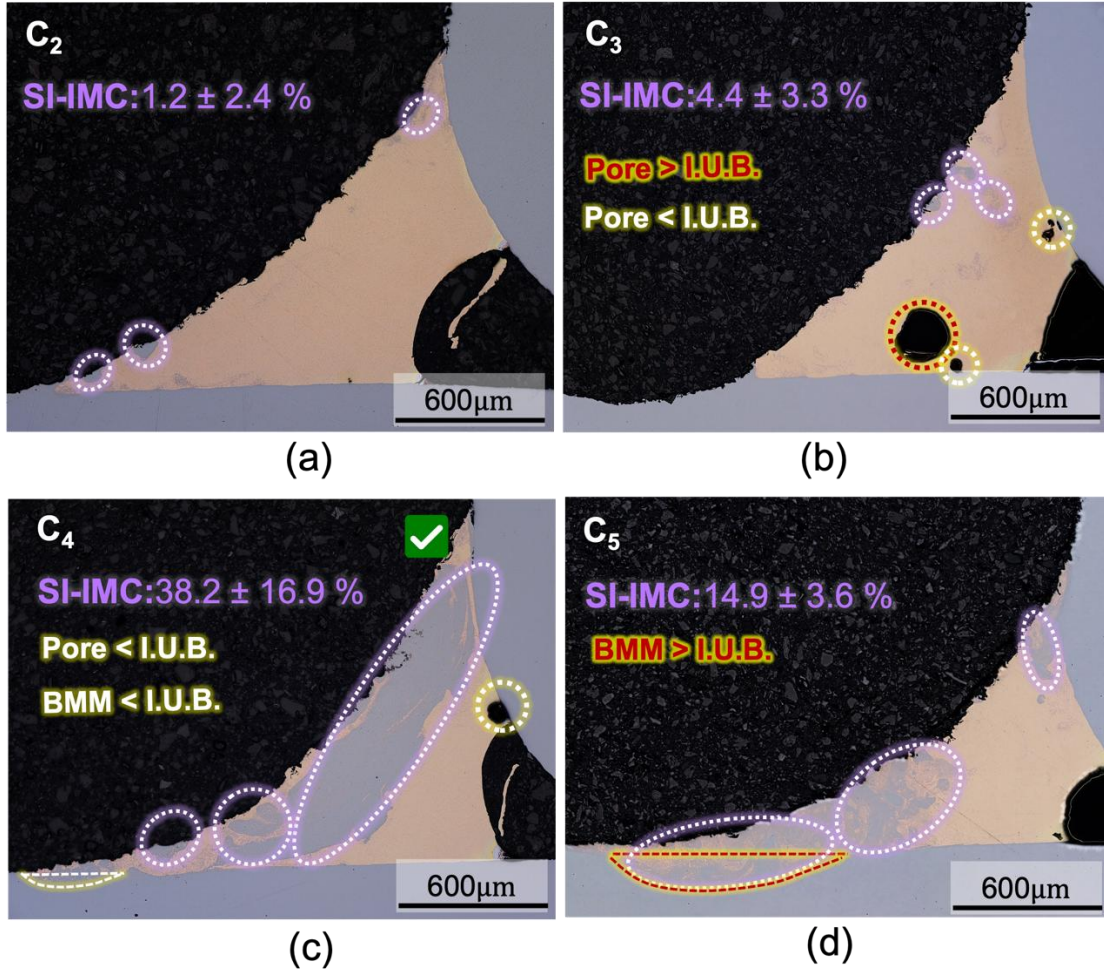


Fig. 5.6: Braze microstructure of the LWB conditions susceptible of containing SI-IMC phases. (a) C_2 , (b) C_3 , (c) C_4 , and (d) C_5 are labeled in correspondence to Fig. 5.5a. SI-IMC are shown by purple dotted circles. SI-IMCs were quantified by measuring their area fraction from four cross-sections of each condition which are presented as mean $\pm$ errors based on 95% CI. The presence of defects such as pores and BMM, with above and below the I.U.B annotated by red and white colors, respectively.

5.4.2 Mechanisms of SI-IMC Formation

5.4.2.1 SI-IMC Microstructure

In Fig. 5.7, the braze microstructure of the LWB condition with the highest SI-IMC density is annotated with four regions of interest that elucidate key phenomena in SI-IMC

identification. The selected scanning area at the braze interior “Region A” provides a broad overview on the phenomena happening within the phases formed inside the brazing matrix, as depicted in Fig. 5.8. As can be seen, three distinct phenomena are happening, which are annotated with color codes of blue, green, and yellow, as follows:

I_A “Blue”: The development of an Fe-rich phase solidified with columnar and equiaxed eutectic structures that appears as a bulky island. The Fe-rich bulky island is composed of (Fe-rich)-Cu-Mn-Zn in which Cu-Mn and Zn are segregated from the filler and the coating, respectively, forming a cellular net-shape Cu-Mn-Zn segregation surrounding each columnar and equiaxed regions;

II_A “Green”: Si segregation at the interface of the Fe-rich bulky island with the Cu-rich braze matrix, resulting in the formation of an interfacial layer with both flat and dendritic morphologies;

III_A “Yellow”: Dispersion of granular particles containing Fe and Si elements within the Cu-rich region of the braze. These particles are less than 1.5 μm in size and their dispersions begin to cluster when approaching the top surface of the braze.

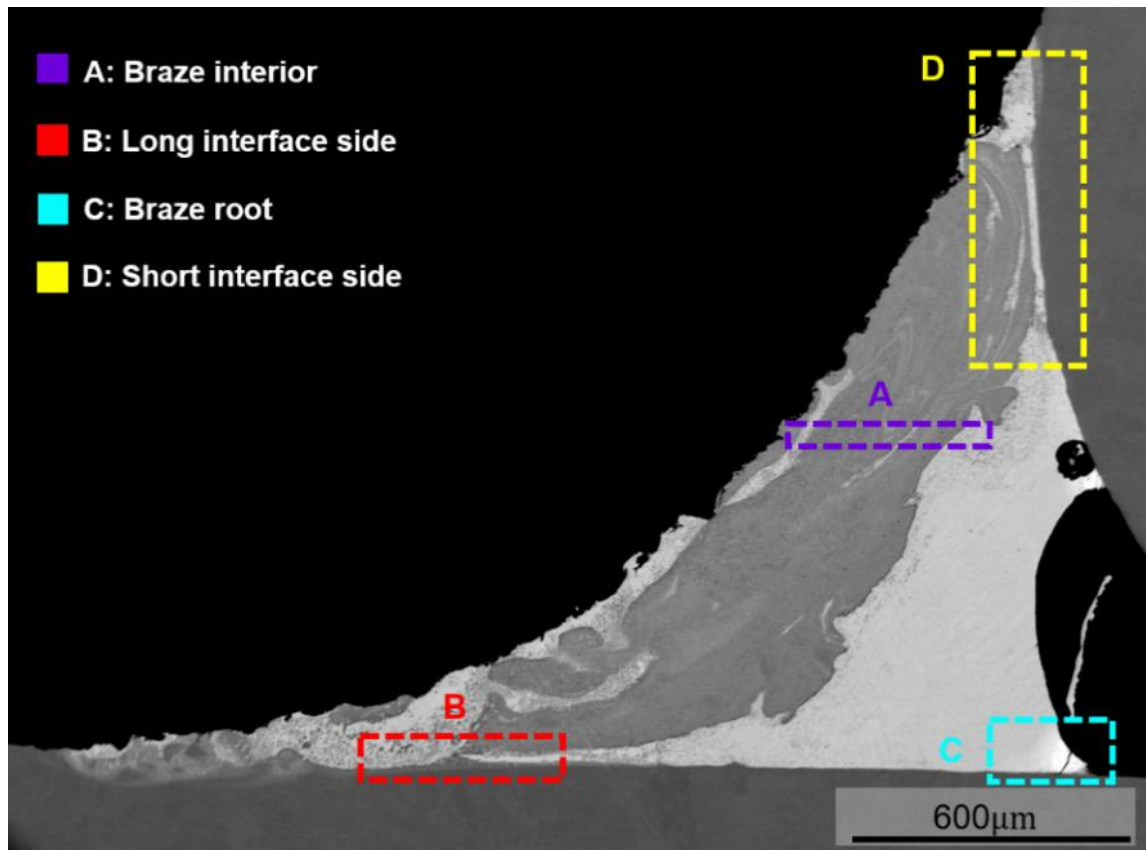


Fig. 5.7: Back scattered SEM micrograph of the braze with high density SI-IMC. In this micrograph that distinguishes between the Fe-rich (dark gray) versus Cu-rich (bright gray) phases. Four regions were selected for high resolution EPMA analysis as shown by the colored boxes that correspond to “Region A” braze interior (Fig. 5.8), “Region B” long interface side (Fig. 5.9), “Region C” braze root (Fig. 5.10), and “Region D” short interface side (Fig 5.11).

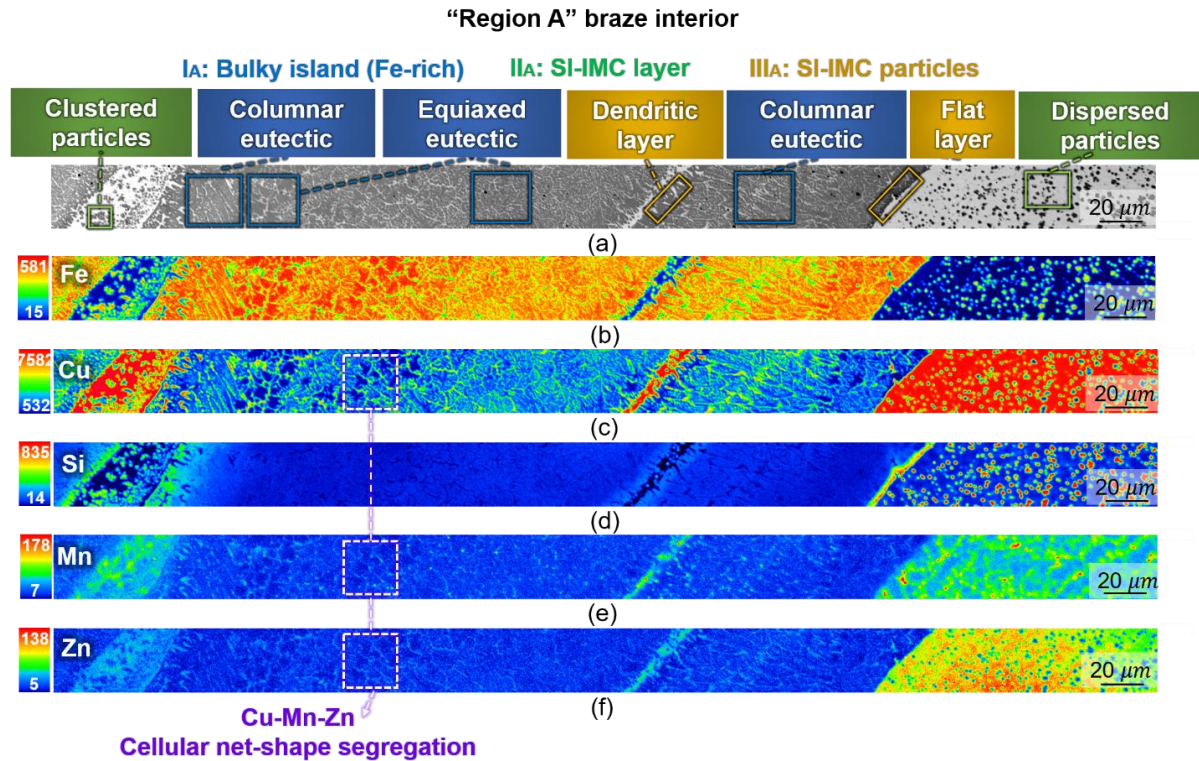


Fig. 5.8: High resolution EPMA elemental maps indicating SI-IMCs found within the braze matrix. These maps were taken from the centre of the braze with high density SI-IMC (braze interior region “A” in Fig. 5.7). The distribution and morphologies of elemental segregation are comprehended from these maps. Four distinct phenomena including formation of a Fe-rich bulky island (IA), SI-IMC layer (IIA), SI-IMC particles, and Cu-Mn-Zn segregation are evident.

As noted by Yu et al. [96], Fe and Si can segregate due to the high tendency of these elements to form IMCs, such as FeSi, FeSi₂, Fe₂Si₃, and Fe₅Si₃, which commonly occur at the brazing interface of steel/Si-bronze joints. The interfacial layers and dispersed particles in II_A and III_A, respectively, are both Fe-Si IMCs. Fe-Si IMCs can appear as dispersed granular phases, as was found during GMAB of steel/Si-bronze in [103,211], and in the interfacial region of Al/steel brazed joints [193,214]. Formation of the IMC particles and IMC layers within the braze interior region substantiates that IMCs are not only limited to the interface of the solid substrate where the molten filler first contacts. As discerned in the present study, IMCs indeed develop in regions where the elements of the substrates intermix with the braze material substances under an elevated heat input resulting from the optimization scheme in the wire-adjusted heat

input approach introduced in Section 5.4.1. This heat input level was sufficient to diffuse a high Fe content into the braze interior without causing substrate melting that exceeds the I.U.B for the BMM defect.

Regarding the Fe-rich bulky island (phenomenon I_A), it can be suggested that a Fe-rich core has been surrounded by a shell composed of Fe-Si IMCs layers with flat and dendritic morphologies. To form the Fe-rich core, a high Fe content must diffuse from the substrate into the braze. Given the absence of excessive BMM (beyond the critical I.U.B) and spattering during the brazing process under this condition, the diffusion of high Fe content may potentially be attributed to the geometry of the brazing coupon. This unique geometry could facilitate the accumulation of diffused Fe atoms, possibly originating from the marginal BMM occurrence ($BMM < I.U.B$). What adds to the significance of this brazing condition with high SI-IMC density is the development of homogenous bonding with the brazing matrix, as evidenced from the formation of the Fe-Si shell-like IMC layer and the absence of any microcracks within this IMC type, located between the Fe-rich bulky island and brazing matrix. In fact, the SI-IMCs, which are surrounded by the Cu-rich braze matrix, form a composite of Cu-rich/SI-IMC.

To gain deeper insight into the correlation between the interfacial reactions at the braze/substrate interface (I-IMC) and the Fe-rich bulky island with Fe-Si IMCs (SI-IMCs), the I-IMC formation along the long interface side was further explored. As shown in Fig. 5.9, interfacial evolutions resulting in segregation of the Fe and Si elements to form Fe-Si IMCs, align well with the expected outcome of the coexisting Fe-rich substrate and Cu-rich braze with Si elements as a diffusion system. In this region, the occurrence of IMC growth can also be observed as follows: I_B) thickening of the interface IMC layer; the interface IMC layer morphology alters from the nearly flat to a dendritic shape with an increased thickness underneath the Fe-rich giant island as annotated by the “growing dendritic layer”. Considering that the progression of I-IMC and SI-IMC formation in each cross-section of the braze is not similar, it can be suggested that Fe diffusion for I-IMC growth is provided from both instantaneous braze/substrate interface and the prior formed Fe-rich bulky island with respect

to the braze length. II_B) Clustering of the dispersed IMC particles from enrichment of the Fe and Si within the larger particles compared to the smaller particles, in which the smaller particles are clustering together to build larger ones, called “clustered particles”. The dendritic morphology of the IMC clusters suggests a growth mechanism for the dendritic IMC layer at the interface. IMC particles, free-floating in the Cu-rich braze matrix, have contributed to IMC growth through clustering. In fact, it can be hypothesized that the IMC particles of the smallest size are the primary IMC nuclei that can form within the Cu-rich matrix. Comparing the IMC growth that occurred in “Region A” and “Region B”, where clustering of the IMC particles is evident, it can be concluded that IMC growth advances towards the braze top surface. Accordingly, it can be speculated that the IMC formation mechanisms in “Region A”, for SI-IMCs, are similar to the phenomena occurring at the interface for I-IMCs, in which it has been shifted towards the interior braze through Fe-rich phase accumulation. It is also interesting to mention that both I-IMC and SI-IMC containing the Fe-Si composition are characterized by a Si gradient with high concentration; thus, Si segregation can be considered as an IMC identifier in steel/Si-bronze joint systems.

To delve deeper into the intricacies of IMC morphologies and their distribution, as well as to explore how these phases extend into the braze interior, the subsequent section will examine the impact of thermal gradients and cooling rates on these phenomena. This will involve considering Zn vaporization as a temperature indicator; for instance, a higher Zn content within a specific region of the braze interior will correspond to a lower temperature at that location.

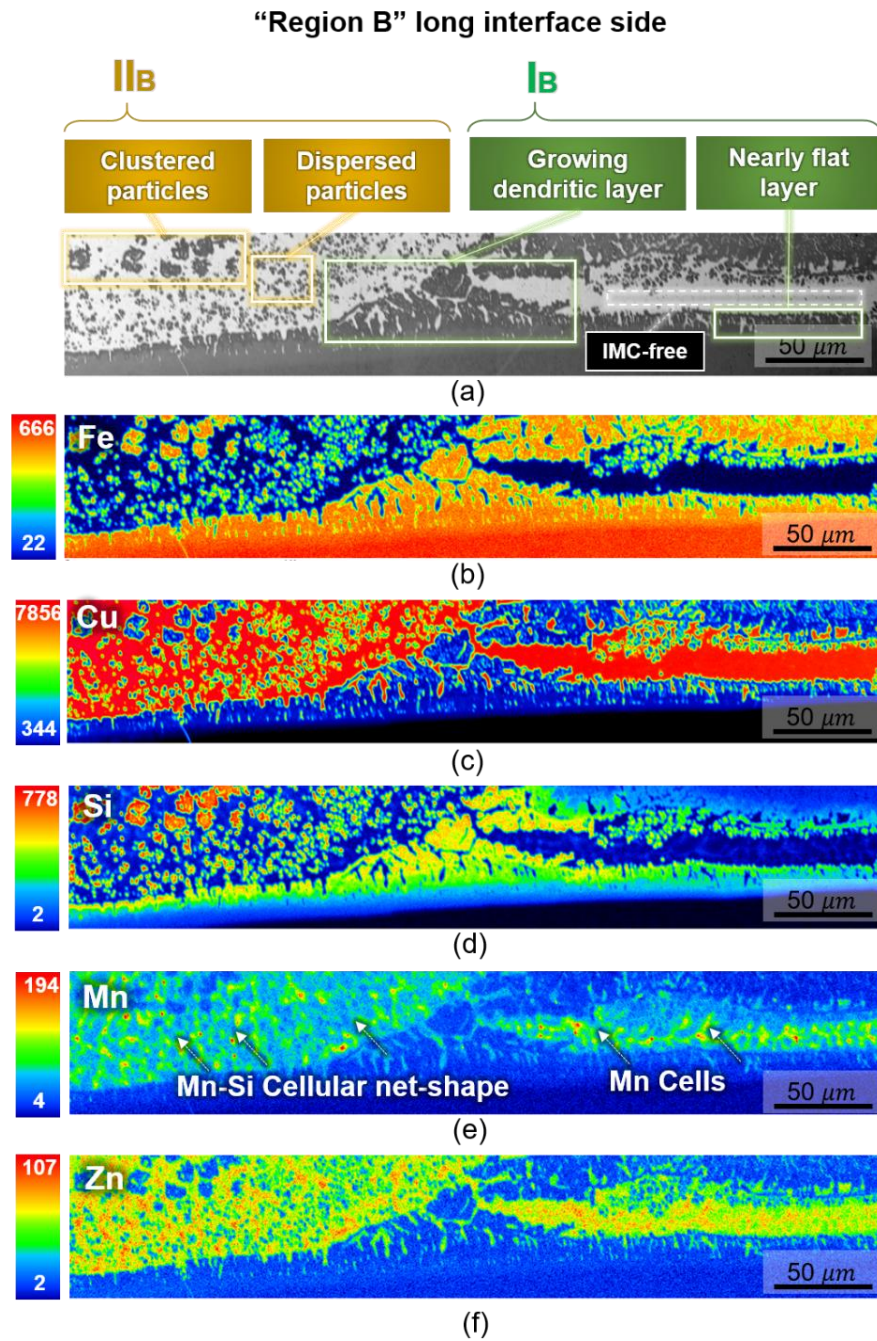


Fig. 5.9: High resolution EPMA elemental maps indicating IMC formation along the interface. This micrograph was taken from region “B” in Fig. 5.7 corresponding to the Long interface side of the braze with high density SI-IMC. Two IMC growth mechanisms including IMC dendritic growth and IMC clustering are comprehended.

5.4.2.2 Effect of Thermal Gradient (TG) and Cooling Rate (CR)

“Region C” in Fig. 5.10 shows the development of a Zn-rich tail at the braze root and a reduction in the thickness of the IMC layer when moving along the interface to the bottom of the braze. The reason for differences between the interfacial morphologies of “Region B” and “Region C” is tied with the TG when the molten filler contacts the solid substrates, as well as the CR during the solidification stage. From the Zn gradient along the longer interface side in Fig. 5.10, it can be interpreted that the temperature decreased with increasing distance from the laser/metal incidence zone. Higher TG at the top surface of the braze compared to the root region of the braze has been expected from previous studies using thermo-fluid simulations according to Mohammadpur et al. [10]. As a result, the diffusion of Fe from the substrate into the braze was reduced; thereby, a thinner Fe-Si I-IMC layer could be formed, as shown in Fig. 5.10b and d. A similar phenomenon is also evident along the shorter interface side, designated as “Region D” in Fig. 5.11, wherein thinning of the interaction layer is succeeded by a reduction in Fe-Si IMC formation as it approaches the root region of the braze. This is well aligned with gradient in IMC density along the braze from the bottom to the location adjacent to the braze top surface in “Region A” and “Region B” owing to a high dependency of IMC nucleation and growth on the temperature and time factors. In fact, higher temperatures can accelerate the formation of IMCs. This is due to the fact that at elevated temperatures atoms across the interface between the two metals can diffuse more readily into each other's lattice structures, which results in furthering the formation of IMCs [215,216].

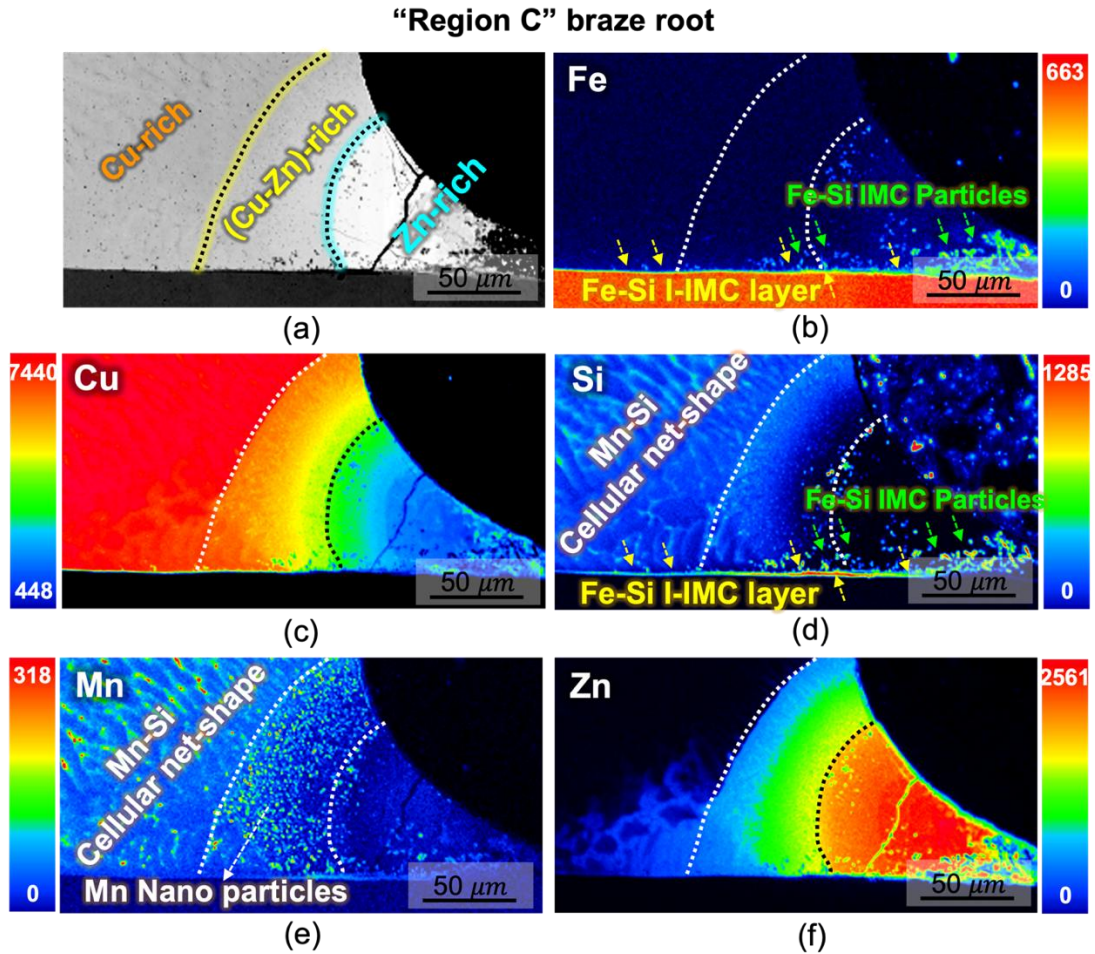


Fig. 5.10: High resolution EPMA elemental maps from the braze root. These maps were taken from Region C in Fig. 5.7, corresponding to the long interface side of the braze with high density SI-IMC. This region is Zn-rich where the TG and CR are both at minimum level leading to slowed IMC formation condition; role of Mn segregation in IMC formation is evident from absence of Si in (Cu-Zn)-rich region, a transition zone between Mn-Si cellular net-shape segregation in Cu-rich matrix and presence of Si in Fe-Si IMCs found within and adjacent to the interface).

The Zn-rich root in “Region C” also contains dispersed particles that are smaller than the Fe-Si IMC particles in “Region A” and “Region B”. These particles are very small, less than 1 μm in diameter, with two distinct compositions and morphologies. The first particle type is the Fe-Si IMC particles that form adjacent to the Fe-Si I-IMC layer of the interfacial region (Fig. 5.10b and d). The other type of particle forms by segregation of Mn atoms as nano particles

within the (Cu-Zn)-rich matrix (Fig. 5.10e). Further towards the braze interior (Cu-rich matrix), the Mn accumulated with Si to form Mn-Si cellular net-shape segregation (Fig. 5.10d and e). However, from the braze matrix in “Region B”, where it is close to the interface, Mn segregated with a cellular net-shape morphology, which is surrounded by Zn instead of segregating with Si (Fig. 5.9e and f). In “Region C”, where Zn is in high abundance, there is a transition from cellular net-shape Mn-Si segregation within Cu-rich matrix to dispersed Mn within Cu-Zn-rich matrix, in such a way that Mn atoms are locked up by the accumulated Zn atoms (where the Zn concentration is high). As the Zn concentration decreases adjacent to the interface (that can be followed in Fig. 5.10b and d), the Si segregates with Fe atoms to form Fe-Si IMCs in the braze, rather than remaining with Mn atoms. Comparing affinity of Mn and Fe to form compounds with Si, Mn’s affinity for Si is not as strong as Fe’s affinity for Si, since Fe-Si compounds are more thermodynamically stable compared to Mn-Si [217]. Mn plays a crucial role by altering the local lattice structure, creating diffusion pathways that allow Si atoms to move more easily. This leads to an enhanced formation of Fe-Si IMCs, as Si prefers to bond with Fe over Mn. Mn facilitates this process rather than directly bonding with Si. The absence of Si in areas where Mn remains as nanoparticles, and its reappearance in Fe-Si IMCs, suggests that Mn plays a substantial role in initiating IMC development by providing diffusion channels or pathways that facilitate the movement of Si atoms toward Fe during Fe-Si IMC formation. These diffusion channels allow atoms to move more freely compared to regions without Mn segregation. Due to the reduced TG and CR in “Region C” in Fig. 5.10, the pace of IMC formation slowed, which made this phenomenon more apparent during Fe-Si IMC formation. Mn channeling can also be seen in “Region B” in Fig. 5.9, where both dispersed and clustered particles of Fe-Si IMCs were nucleated and growth within the Mn-Si cellular net shape segregation. Due to the increased TG and CR in “Region B” compared “Region C” the IMC formation occurred faster and progression of Si diffusion towards Fe atoms through Mn channels is well evident. It should be noted that there is an IMC-free region in “Region B” as seen in Fig. 5.9a, in which Mn cells are still present, while there is no Fe-Si IMCs. This could be attributed to the reduced availability of Si content in that area. One possible explanation is

that as the Cu-rich matrix shrinks due to the floating of Fe into the braze, forming a bulky Fe-rich region, there is less Cu-rich matrix available despite of the presence of Mn cells, which contains the Si needed for Fe-Si IMC formation. This reduced availability of the Cu-rich matrix may limit the granular IMC formation or further growth of the I-IMC layer. Furthermore, the available Si atoms in this region have been consumed during the growth of IMCs in neighboring regions with higher Fe content. Mn channeling phenomenon has been studied for the first time in the current work, since the contribution of any other braze/substrate constituent elements in Fe-Si IMC formation has never been disclosed so far in the open literature based on the best knowledge of the authors.

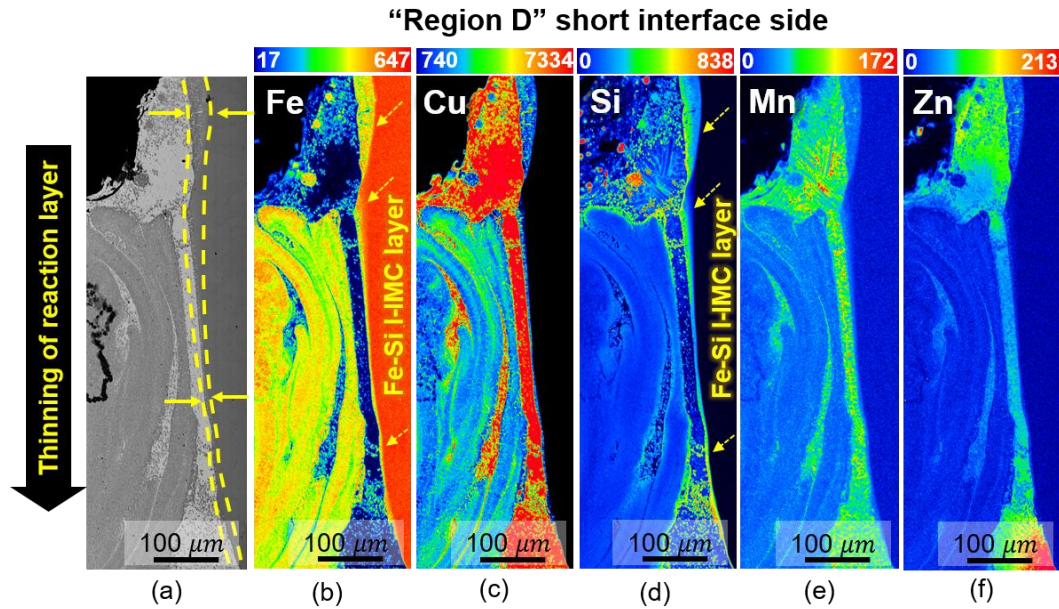


Fig. 5.11: High resolution EPMA elemental maps showing reduction in IMC thickness by increase in Zn concentration. These maps were taken from of the short interface side of the braze with high density SI-IMC (Region D in Fig. 5.7).

Using a specific braze geometry, like the double-flanged configuration used in the roof joint in this study, which represents a real-world braze joint application, allows for the examination of a unique braze microstructure that has not been previously explored, unlike simplified configurations, such as lap or overlap configurations. This coupon geometry, along with using an elevated heat input controlled by wire-adjusted heat input approach, explicitly unveil key

IMC formation mechanisms due to the wide range of TG and CR provided by this complex geometry. As demonstrated in the schematic in Fig. 5.12, during the heating cycle, shown in Fig. 5.12a, the molten wire contacts the surface of the substrate during the wetting stage. In this stage, although the laser/material interaction mostly concentrated on the filler wire, the substrate is also affected by the laser at the interaction volume. Hence, during formation of the braze molten pool, in Fig. 5.12b, the areas that are closer to the laser beam interaction area (top surface of the braze molten pool) would experience a higher temperature due to a higher heating rate according to Mohammadpour et al. [10]. Considering Zn vaporization as a temperature indicator, a locally higher Zn content corresponds to a lower temperature at that point. In fact, the Zn concentration gradient along the brazing interface aligns with the temperature gradient during the heating cycle. Therefore, when considering the molten braze metal from the surface to the bottom, its peak temperature during the LWB process decreases and therefore IMC growth is suppressed at the braze/substrate interface. Then, during the solidification stage, Fig. 5.12c, which begins as the laser heat source passes, the areas that are closer to the braze surface will experience a higher CR. In general, a higher CR during LWB is expected to suppress IMC diffusion-controlled-growth resulting in refinement in the IMC morphology [104]. However, in the case of the present roof joint with the double flanged configuration, a higher solidification rate caused by the higher CR further assisted stabilization of the IMCs that formed within the top surface of the braze. This suggests that the IMC formation from the high Fe accumulation within the braze molten pool that occurred during the wetting stage progressed faster due to the higher CR. Therefore, instead of further mixing of the Fe with the Cu-matrix, SI-IMCs layers formed at the interface of the Fe-rich bulky island/Cu-rich matrix and IMC clusters were formed rapidly at the top surface of the braze.

Fig. 5.12d gathers the correlations observed between the IMC evolution and changes in the TG and CR within the braze geometry. The enrichment and morphology of the SI-IMCs varied with their distance from the braze top surface due to the difference in the braze seam's TG. A simultaneous increase in the TG and CR on reaching near to the braze top surface promotes IMC growth mechanisms – including IMC layer thickening (I-IMC) and particle clustering

(SI-IMC). Furthermore, in regions where the bulky Fe-rich islands formed, TG and CR are slightly under their maximum levels, which appeared to facilitate the diffusion of high Fe content. This phenomenon is primarily attributed to the Marangoni flow, which is induced by surface tension driven flow; this provides the activating driving force for the formation of pore and BMM defects. According to Deng et. al. [104], the high Marangoni flow accelerates the migration of molten material within the braze joint. Pores act as sites for localized concentration gradients, leading to differential surface tensions along the liquid-solid interface [218]. Similarly, the presence of BMM introduces compositional variations that further enhance surface tension differentials. These variations in surface tension generate Marangoni flows, wherein material moves from regions of low surface tension to regions of high surface tension. Consequently, this high Marangoni flow drives the redistribution of elements, including the migration of molten material towards regions of higher surface tension [219,220]. Consequently, this facilitated the diffusion of high Fe content into the braze interior region, leading to the formation of a high density of SI-IMCs within the brazed joint. The insights from IMC formation mechanisms (e.g. SI-IMC nucleation, SI-IMC growth, and SI-IMC distribution correlation with TG and CR affected by the roof joint geometry) has been achieved for the first time through the IMC manipulation strategy found using our novel wire-adjusted heat input approach which allowed the application of elevated heat input whilst eliminating braze defects.

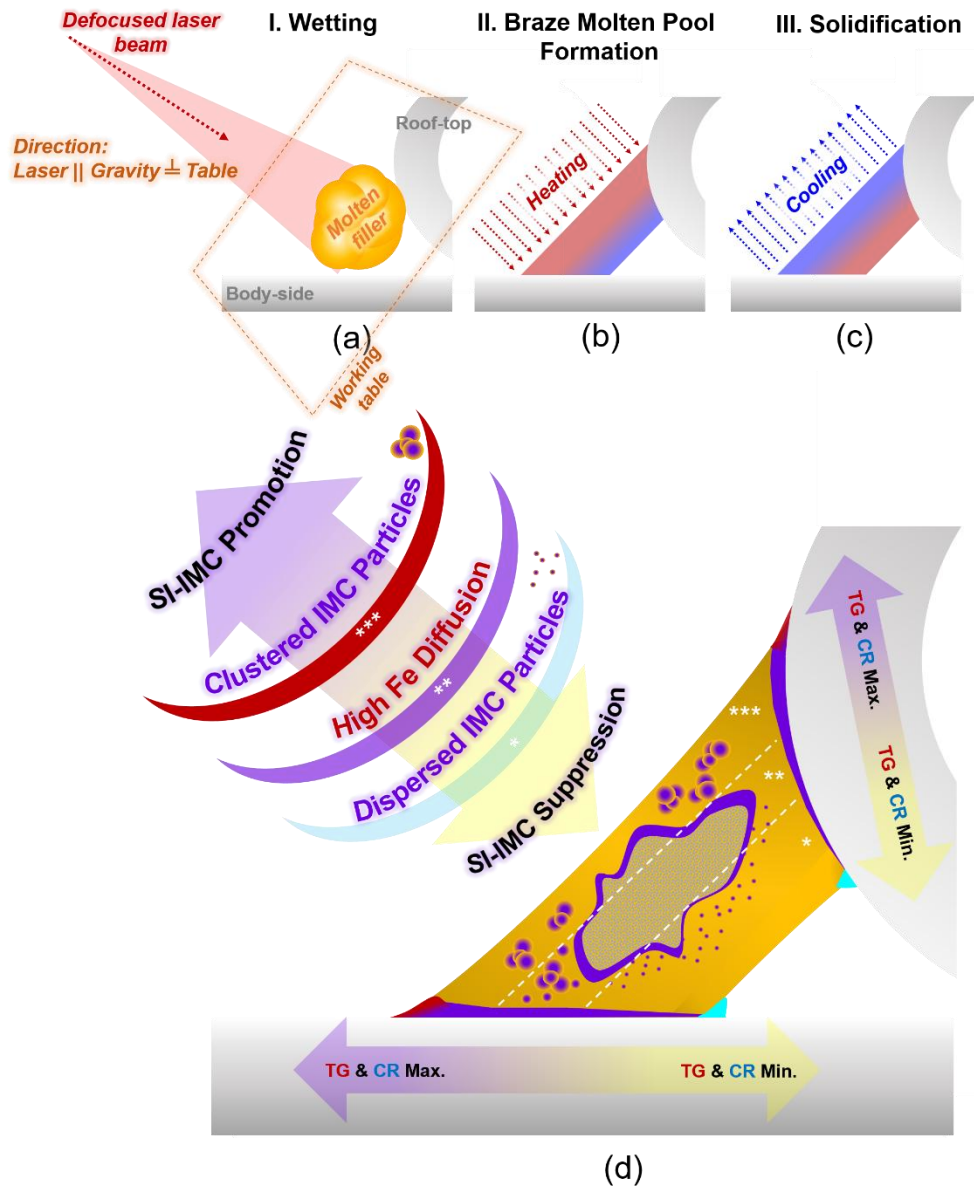


Fig. 5.12: Schematic diagrams summarizing the effect of TG and CR as geometry-related factors on SI-IMC formation. (a) Laser beam and filler metal interaction (wetting stage). (b) TG during the heating stage. (c) TG during cooling stage. (d) Sketch of the correlation between SI-IMC development and change in TG and CR within in the braze joint in laser weld brazing on the roof joint configuration.

5.4.3 Mechanical Performance of LWB Joints

In evaluating the mechanical performance of the resulting braze joints, Cu and Fe, the main constituent elements from the filler material and substrates respectively, play a crucial role. The overall impact, however, is more closely tied to the products of these elements in forms of Cu-rich braze matrix, as well as IMC phases resulting from Fe diffusion into the braze matrix. To investigate the effect of SI-IMCs formation on the joint mechanical performance, the tensile behavior of the braze samples with low density ($1.2 \pm 2.4\%$) and high density ($38.2 \pm 16.9\%$) SI-IMCs were examined. According to their strain distribution maps, as shown in Fig. 5.13, the three stages from deformation to failure during the tensile testing of the sample are characterized as follows: I) unbending of the body-side component, II) transient strain localization, and III) failure. The highest value at the gradient scalebar in each strain map demonstrates the highest local strain in each stage. Comparing their unbending stage, the bending is completed at a maximum localized strain of 0.08 and 0.09, for samples with low and high density of SI-IMC, respectively, which are nearly equivalent. In both samples, the occurrence of unbending at the same region (base metal) is responsible for the equal strain values. Then, in stage II, the braze with low SI-IMC density showed four localized strain zones (maximum strain of 0.26), as annotated in Fig. 5.13c, which consisted of locations at the braze tails of both sides, braze top, and the roof-top base metal far from the braze. By increasing the tensile loads, in stage II, the strain localization zones shifted towards the braze tail at the roof-top side leading to failure (stage III) at a localized strain of 0.41. In the braze with high SI-IMC density, the strain initially localized at the roof-top side and the braze tail at this side with a maximum strain of 0.41 (stage II in Fig. 5.13g). In the failure stage (III), strain localization shifted towards the base metal at a maximum strain of 1.8. Shifting the failure from near the braze, in the sample with low SI-IMC density, to far from the braze, in the sample with high SI-IMC density, can be attributed to the strengthening behavior of SI-IMCs within the braze which improved the joint strength and delayed failure. This fracture behavior was consistent in the three tensile repetitions. It should be noted that the exact mechanism linking the

strengthening of the braze to the shifting failure location and improving the joint strength must be further studied.

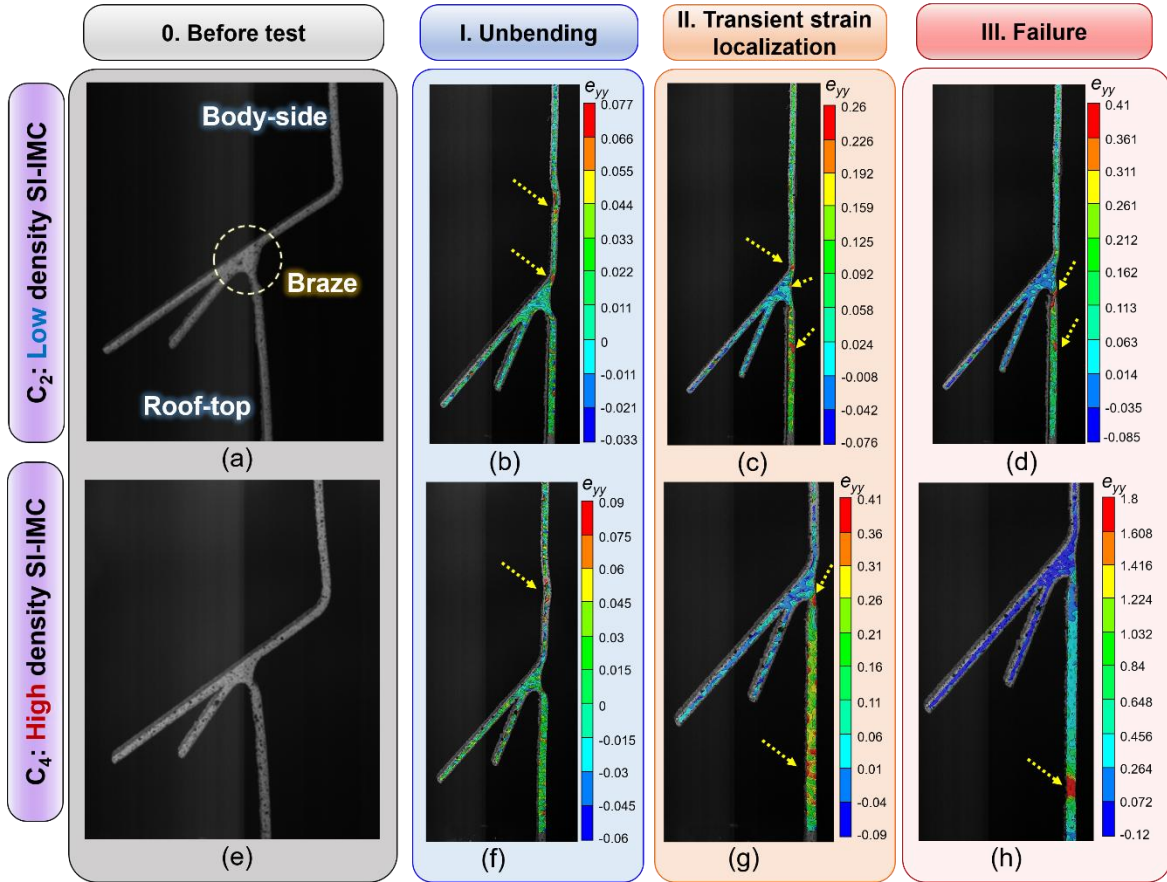


Fig. 5.13: Strain distribution maps during tensile testing calculated by DIC technique. Different stages of tensile deformation, before and during the test, are depicted for two braze conditions with low and high SI-IMC fractions. (a-d) Low density SI-IMC ($1.2 \pm 2.4\%$) or C_2 condition. (e-h) High density SI-IMC ($38.2 \pm 16.9\%$) or C_4 condition. The field of view is the brazing seam cross-section. The strain localization zones in each stage is indicated by yellow arrows.

A comparison of the tensile force versus displacement curves of these brazed samples with the initial GI-IF steel substrate, as depicted in Fig. 5.14, shows that the tensile properties of the brazed joint with low density ($1.2 \pm 2.4\%$) and high density ($38.2 \pm 16.9\%$) of SI-IMC differ considerably from the initial substrate. Specifically, the sample with low SI-IMC density failed rapidly at a diminished peak load of 560 N with a significant reduction in the total displacement

at a value of 1.2 mm compared to the initial substrates that exhibited a higher peak load of 591.0 N and displacement of 11 mm at failure, as tabulated in Table 5.2. By contrast, the braze with high SI-IMC density exhibited an enhanced peak load (650 N) and displacement when compared to the low SI-IMC density sample. Also, the high SI-IMC density sample exhibited a higher peak load than the initial substrate, and the reduction in total displacement was lower relative to the low SI-IMC density sample. It is evident that the tensile behavior of the braze with high SI-IMC density is similar to the initial GI-IF steel substrate, which can be recognized from the similar trend in the shape of their force-displacement curves.

Enhancement of the mechanical performance was due to dispersion strengthening by second phases, this phenomenon was also reported from previous studies on braze joints made using LWB [211] and GMAB [120] that described the brazed areas as composites. The presence of Fe-Si IMCs within the braze matrix, namely the aforementioned SI-IMC phases, supports the idea that SI-IMCs can act as a strengthening agent during tensile deformation. Hence, the observed reduction in ductility that accompanies this strengthening is similar to that seen, for instance, in composite materials; however, here the reduction observed was considerably lower for the sample with high SI-IMC density. Specifically, the tensile peak load and ductility of the brazed joint increased by 14% and 350%, respectively, in high density SI-IMC sample relative to the low density sample. Furthermore, to assess their toughness specifically, the area under the tensile force versus displacement curves for each condition was computed with tolerance representing 95% confidence interval, as listed in Table 5.2. The toughness indicator values for the brazed joints with 0.44 J in the low SI-IMC density and 2.52 J in the high SI-IMC density samples are smaller than the GI-IF substrate (5.69 J). Interestingly, the toughness reduction was significantly less pronounced ($\sim 55\%$) in the high SI-IMC density sample compared to the low density (more than an order of magnitude). Thus, the reduction in the toughness of the brazed joint can be mitigated by increasing the SI-IMC fraction owing to their strengthening effect. Noteworthy, while the IMC composition and thickness at the braze/substrates interface are known to have significant impact on tensile strength of the brazed joints in general according to Bunaziv et al. [214], in the case of this study, the major difference

between the braze joints of C2 (with low SI-IMC density) and C4 (with high SI-IMC density) lies in the SI-IMC density rather than in the I-IMC (or interfacial IMC) thickness. Therefore, the strengthening effect seen in high SI-IMC density braze was attributed to the high fraction of SI-IMCs. This is a reasonable approach, as no failures were observed at the braze/substrate interface. The strengthening mechanism is, therefore, expected to be related to the SI-IMCs rather than the I-IMCs.

The mechanical performance of the sample with high SI-IMC density was enhanced; however, the presence of SI-IMCs may not be the sole factor improving the joint mechanical performance. It should be noted that the braze joints were all approximately the same size, but the complex geometry of the brazing coupons (replicating the roof joint) and asymmetrical members makes the stress-state during the test, and subsequent failure, complex. Hypothetically, the results suggest that the high density of SI-IMC could improve joint mechanical performance under such similar complex deformation; however, this needs to be further substantiated to be conclusive. Nonetheless, there is precedent data for such mechanical performance improvement where failure is shifted from the joint to the base metal region due to the existence of a high density of precipitates along the interfacial region of the braze, which has been reported by Sarmast-Ghahfarokhi et al. [120]. Further investigation should be conducted on producing braze joints with simple configurations (i.e. lap shear) that contain a high density of SI-IMC. To understand the direct effect of the promoted SI-IMCs, the braze region should be a weak part of the tensile coupon, which is proposed to be possible by using a higher strength substrate compared to the IF steel, which is the focus of a follow-on study. While further investigation is required to eliminate potential geometrical effects that may have contributed to the observed improvement in mechanical performance, this study successfully introduced a clear strategy for IMC manipulation using the novel wire adjusted heat input approach which can be extrapolated to other applications. It also addresses critical research gaps, offering a deeper understanding of IMC formation mechanisms and strategies to promote IMC growth within the braze matrix—rather than at the braze/substrate interface—to reduce

embrittlement from thickened IMCs. These provide a solid foundation for mitigating IMC embrittlement and improving joint integrity, with adaptable applications across various fields.

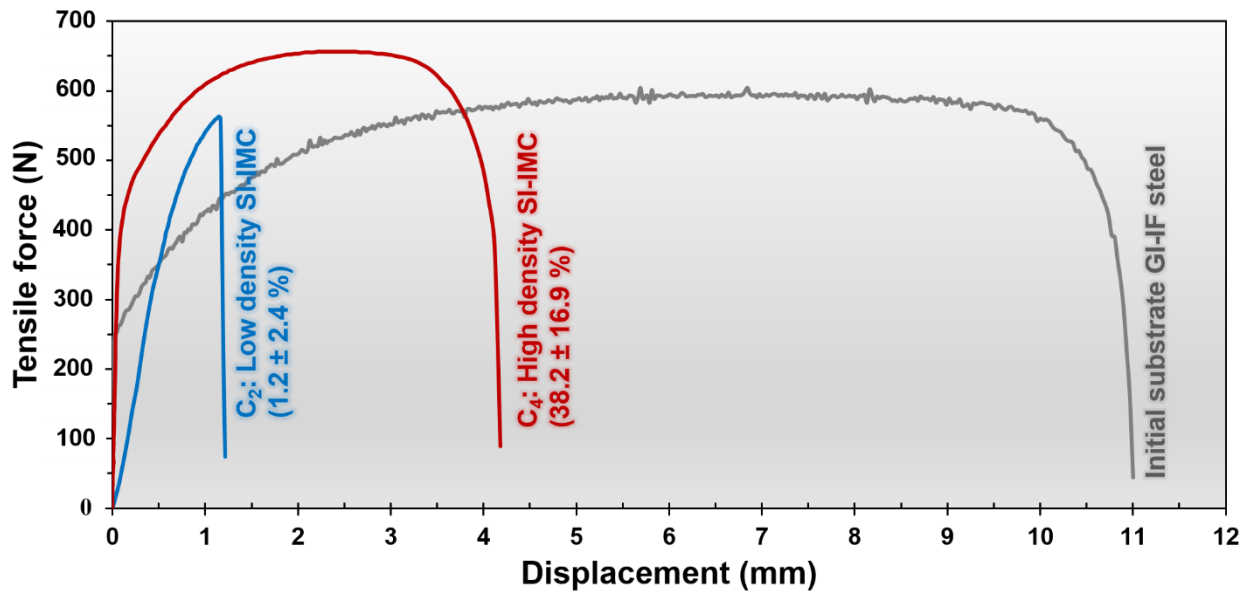


Fig. 5.14: Representative tensile force versus displacement curves. This graph includes tensile curves of the GI-IF steel substrate and brazed samples with low and high SI-IMC density.

Table 5.2: Corresponding peak loads and total displacements averaged from three repetitions of tensile tests. The deviations from the mean value of the measurements are based on 95 % confidence intervals.

Sample	Peak load (N)	Total displacement (mm)	Toughness (J)
Initial substrate GI-IF steel	591.0 ± 8.9	11.1 ± 1.7	5.7 ± 0.4
Low SI-IMC ($1.2 \pm 2.4\%$)	559.7 ± 8.7	1.2 ± 0.3	0.4 ± 0.1
High SI-IMC ($38.2 \pm 16.9\%$)	648.3 ± 14.1	4.3 ± 0.5	2.5 ± 0.1

5.5 Summary

In the present study, a novel heat input related approach was proposed to tackle the challenges of IMC embrittlement and the absence of a clear strategy for manipulating IMC formation during LWB of thin-gauge Zn-coated automotive steel sheets with Si-bronze filler material. This approach, applied to a double-flanged lap joint configuration, manipulated IMC distribution by promoting IMC formation to the interior of the braze rather than braze/substrates interface. The comprehensive analysis of elemental segregation within the revolutionized braze microstructure and, examination of the subsequent joint mechanical performance provided critical insights into improving LWB processes. These findings not only address specific issues related to automotive brazing but also offer a broader framework for advancing LWB and similar processes in various industrial contexts, demonstrating significant academic and practical relevance. The main conclusions could be drawn as follows:

1. A novel IMC manipulation method was introduced using a wire-adjusted heat input approach with 3D representation of $C_{i=1,2,3,...,n}$: ($HI_{Nominal}$, H_{Wire} , $\Delta HI_{Relative}$). For the proposed condition under an elevated heat input, a value of 12.44 J/mm above the $HI_{Relative}$ for a defect-free braze condition, was found to be highly favorable for shifting the severity of IMC formation at the braze/substrate interface, i.e., I-IMC, to the braze interior region, i.e., SI-IMC formation. The principal growth mechanisms for both I-IMC and SI-IMC types were IMC particles clustering and IMC layer thickening, respectively.
2. A high density SI-IMC braze was achieved with $38.2 \pm 16.9\%$ of its area fraction composed of a bulky (Fe-rich)-Cu eutectic phase covered by shell-like Fe-Si layer with flat/dendritic morphologies and dispersed/clustered Fe-Si particles.
3. TG, CR, (relating to braze configuration) and wire-adjusted heat input were identified as pivotal factors in shaping IMC morphology and distribution during LWB. TG and CR showed a linear relationship with IMC growth promotion. SI-IMCs were found to be mostly promoted where TG and CR reached their peak levels within the braze under a controlled elevated heat input; this enabled the necessary Marangoni flow for an

increased Fe diffusion into the braze to form high density of SI-IMC with the Fe-Si composition, without causing substantial melting of the base metal.

4. The significant contribution of Mn in initiating IMC development by facilitating Si diffusion towards Fe atoms during Fe-Si IMC formation, called Mn channeling, was clarified for the first time. As the Zn concentration decreases, TG and CR increases. Si then segregates towards the Fe atoms in the braze rather than remaining with Mn atoms, which are segregated as a Mn-Si cellular net-shape structure within the braze.
5. The strengthening effect of the SI-IMCs is evident from the enhanced mechanical performance observed under tensile testing. As the density of SI-IMCs increased within the braze matrix, the tensile peak load and ductility showed notable improvement by 14% and 350%, respectively, when comparing the high and low SI-IMC density samples.

Chapter 6. Disentangling the Failure Response of Various Intermetallic Compounds with Surrounded Interface Formed during Laser Weld Brazing Process ⁵

6.1 Overview

This study elucidates the effect of SI-IMCs, i.e., IMCs formed within the braze seam, on the mechanical performance and fracture behavior of the joints. LWB was performed on a Zn-coated Gen 3 Q&P980 steel to produce braze joints containing various SI-IMCs that were dependent on the heat input of the LWB process. SI-IMCs formed at a lower heat input ($\Delta HI_{Relative}$ range of 2.3–7.5 J/mm) significantly improved the mechanical performance, with a 91.7% increase in the peak load, a 319.84% increase in the displacement, and a 1200% improvement in toughness compared to the degraded properties observed in SI-IMCs formed at higher heat inputs ($\Delta HI_{Relative}$ range of 13–17.5 J/mm). EBSD was used to examine the fracture microstructure and texture of the SI-IMCs and the Cu-rich matrix of the braze. Five types of SI-IMC structures with different morphology and composition were identified, each affecting crack behavior. Four distinct SI-IMC structures were found in the braze, noted as α , β , γ , and δ in this study. Fine α -, β -, and γ -SI-IMCs delayed failure, while coarse α - and δ -SI-IMCs accelerated failure. It was found that a strong $\{100\}$ texture alignment between SI-IMCs and the braze matrix can enhance strain accommodation That improve crack resistance. A unitless $GND.Ratio_{(Cu/Fe)}$ was introduced to measure dislocation density and control joint toughness in the Cu-rich braze joints. At a $GND.Ratio_{(Cu/Fe)} > 0.67$ with high braze/SI-IMC texture alignment, a “SI-IMC-induced twinning” mechanism was activated, locally rotating the Cu-rich matrix orientation to $\langle 112 \rangle$ and forming twins near the crack propagation path, enhancing strain accommodation, and shifting the fracture mode from brittle to ductile, which

⁵ This chapter is composed of a research paper as outlined in the statement of contributions. **S. Akbarian**, Salandari-Rabori, D. Ramachandaran, M. Shojaei, S. Sarafan, P. Wanjara, J. Gholipour, E. Biro, Disentangling the failure response of various intermetallic compounds with surrounded interface formed during laser-weld-brazing process, to be submitted to Metall. Mater. Trans. A.

improved the mechanical performance. A graphical abstract summarizing this chapter is presented in Fig. 6.1.

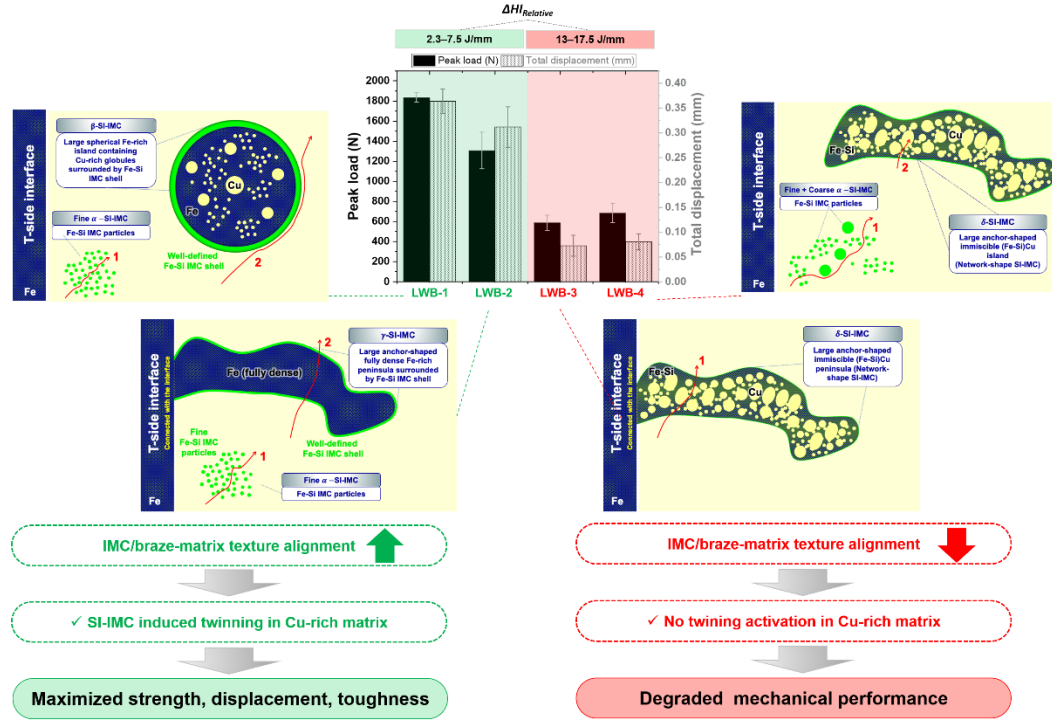


Fig. 6.1: Graphical abstract illustrating an overview of chapter 6.

6.2 Background

The urgent need to mitigate global warming has driven a major shift toward sustainable transportation. EVs have emerged as a key solution to reduce greenhouse gas emissions [3]. One of the key challenges in EV design is the need for lightweight structures to improve energy efficiency [4]. As a result, multi-material integration has become a necessity in the automotive industry, promoting the combination of dissimilar materials with tailored mechanical properties [16]. However, multi-material joining inherently introduces metallurgical complexities, particularly the formation of IMCs at the interfaces, which can significantly influence mechanical performance [17,18]. While IMCs are often associated with embrittlement [25], their exact role in deformation and fracture behavior remains unclear. Given the increasing use of LWB in automotive structural applications [28,29], understanding

the microstructural evolution of IMCs and their impact on the mechanical response is essential for optimizing joint reliability and ensuring passenger safety.

LWB has gained widespread industrial adoption due to its ability to produce high-quality joints with minimal heat input, reducing thermal distortion while maintaining sufficient mechanical strength [115,179]. Compared to fusion welding methods like laser welding, LWB reduces substrate degradation by operating at a lower process temperature, achieved through the use of filler metals with lower melting points. Additionally, the high energy concentration of LWB facilitates a refined joint morphology, leading to improved joint appearance and reduced damage to the base metal. However, despite these advantages, IMC formation is an unavoidable by-product of the brazing process, occurring both at the braze/substrate interface and within the solidified filler region [110,221,222].

A major concern with IMCs in brazed joints is their inherent brittleness. As IMCs are stoichiometric phases with fixed atomic ratios, the atoms in these phases are ionically and covalently bonded, limiting their ability to accommodate lattice strains induced by deformation [26,27]. This rigidity makes them susceptible to crack initiation at the IMC/braze-matrix interfaces, increasing the risk of premature failure compared to joints without IMCs. In weld brazing of steels, Cu-based alloys containing Si, commonly known as Si-bronze, are widely used as filler material [19]. Fe-Si IMCs with various stoichiometries, including Fe_2Si , Fe_5Si_3 , FeSi_2 , Fe_2Si , and FeSi , can form either within the braze/substrate interfacial layer or as dispersed particles within the joint [20–24].

Despite extensive research on the chemistry of Fe-Si-containing IMCs, their influence on the mechanical performance of joints remains an open topic of discussion, with conflicting perspectives and findings in the existing literature. Fe-Si IMC, often referred to as precipitates, have been identified as primary fracture sites in LWB joints due to their presence along the crack path [21,115]. While in previous studies the degraded mechanical properties of joints were related to the presence of IMCs within the braze, the potential contribution of a smaller bead geometry (as possible with LWB) to this behavior requires further exploration. Studies

on the mechanical response of arc brazed material under indentation loading have reported that the precipitates aid in shear band formation, allowing the material to maintain strength, while enhancing ductility [121]. Although a high density of precipitates in arc-brazed joints was reported to enhance tensile properties, it is unclear whether this consistently increases strength or displacement [124]. However, the fracture behavior of the joints with a high precipitate density remains unexplored, as failure occurs in the softer IF steel substrate, regardless of the precipitate density.

Recent studies on multi-phase FCC-based alloys [24,223–227], particularly high-entropy alloys (HEAs) that have undergone thermomechanical processing [226], has suggested that IMCs may play a more complex role in mechanical behavior rather than simply acting as embrittling phases. IMCs can raise local stress levels, leading to twinning activation, which enhances strain hardening and restricts micro-crack formation. Additionally, localized reductions in SFE have been shown to improve strain accommodation, contributing to increased ductility and delayed fracture. While the strengthening effect of IMCs has been observed in IMC-containing HEAs under tensile loading, the extent to which they apply to a multi-material coupon, such as brazed joints, has yet to be fully determined. To date, studies on LWB have primarily focused on IMC formation at the braze/substrate interface and their presence as precipitates. The effect of IMCs on mechanical performance, whether beneficial or detrimental, remains poorly understood. In most studies, including dissimilar joints like aluminum to steel using Al-based filler materials with extensive literature, the focus remains on IMCs at the braze/substrate interface, with limited attention to IMCs formed within the braze seam [109,228]. The broader impact of these IMCs on the fracture behavior, particularly their morphology-dependent influence on the surrounding matrix, mechanical response, strain redistribution, and crack propagation, remains a critical research gap.

In previous work, a process optimization strategy was introduced to promote IMC formation within the braze, leading to the formation of SI-IMC phases, which positively impact mechanical performance [195]. These SI-IMCs differ from conventional I-IMC layers, which

typically form at the braze/substrate interface. SI-IMCs appear as Fe-rich islands with Fe-Si IMC layers and particles, whereas I-IMCs mainly consist of Fe-Si IMC layers at the braze/substrate interface in the CuSi3Mn/Fe system. This was studied during LWB of GI-IF steels using Si-bronze filler wire in a double-flanged joint configuration used in automotive construction for roof-ditch applications. SI-IMCs were observed to improve strength and displacement during tensile loading, enhancing mechanical performance. However, the joint's complex geometry hindered a direct assessment of the SI-IMC contributions in the fracture behavior. To address these gaps, this study aims to clarify the role of SI-IMCs in the brazed joints by examining their impact on the mechanical response of LWB joints using a simplified lap joint configuration to minimize geometrical errors during the tensile tests. A Gen 3 Q&P980 steel, offering significantly higher strength than GI-IF steel, is used as substrates to promote fracture within the braze region and enable a more accurate evaluation of braze performance.

This study offers a pioneering investigation into the development of a new generation of IMCs, referred to as SI-IMC structures, which influence braze mechanical performance and fracture behavior. By disentangling the morphology-dependent interactions between the SI-IMCs and the surrounding Cu-rich matrix, it challenges the traditional perception of IMCs as mere failure initiators. Instead, it highlights their potential to modify strain distribution, strain-hardening behavior, and crack propagation pathways. These findings provide a foundation for optimizing IMC/braze-matrix interactions in brazed joints, advancing the design of stronger, more reliable multi-material automotive structures.

6.3 Materials and Methods

6.3.1 Material Selection

The substrates used in this work were 1.2 mm thick sheets of an industrially produced Zn-coated 1.2 mm thick Gen 3 Q&P980 steel with a composition as described in Table 1. Its microstructure consisted of a combination of 38% ferrite, 47% tempered martensite, 9% fresh martensite, and 6% retained austenite as examined in previous work [229]. The Gen 3 Q&P980

substrate has an UTS of 1013 MPa, yield strength (YS) of 640 MPa, and 20% elongation, ensuring a high resistance to displace any base metal failure into the “softer” braze joints. This Gen 3 Q&P980 steel was also chosen for the present study for its structure that contains 92% body centered cubic (BCC) phases, which will distinguish the Fe-rich phases in the braze from the Cu-rich matrix during texture analysis due to the steel’s and braze’s distinct crystal structures, as the braze alloy has FCC structure. The steel sheets were waterjet cut in 10 × 20 mm² rectangular samples to be used as coupons for the lap joint configuration. Coupons were cut so that the brazing direction was perpendicular to the rolling direction. A Si-bronze filler wire with CuSi3Mn1 composition as detailed in Table 6.1, and 1.2 mm diameter in size was used for the LWB process.

Table 6.1: Chemical composition of Gen 3 Q&P980 substrates and filler wire for LWB.

Material	Fe	Cu	C	Si	Mn	Cr+Mo+Ti+Nb
Gen 3 Q&P980	Bal.		0.2	1.1	2.0	0.22
CuSi3Mn1	0.02	Bal.	-	2.90	0.80	-

6.3.2 LWB Process Setup

The LWB was conducted using a Robotic Hybrid Laser Welder (Lincoln Wayne-Trail/IPG) with multiple heads irradiating various ranges of IPG Ytterbium-doped fiber laser with a maximum power capacity of 8 kW. The laser had a wavelength of 1070 ± 10 nm, and a beam waist diameter of 600 μ m, with a Gaussian beam profile, was used. The welding head was mounted on a robotic arm delivery. Fig. 6.2 shows a schematic of the LWB process for a lap joint configuration. To maintain consistency with the LWB process setup used in the roof joint configuration from our previous work [195], the same laser beam spot size (3.2 mm) was used, and no shielding gas was incorporated into the process, similar to standard industrial procedures. The beam was defocused 42 mm from the focal point to achieve a 3.2 mm spot size diameter centered on the filler wire centerline at the contact point of the filler wire with the substrate (Fig. 6.2a). A lap joint was used with an overlap of 4.5 mm for all conditions.

The filler wire was directed to the surface of the bottom sheet substrate adjacent to the edge of the top sheet substrate with 0.6 mm distance from its center as shown in Fig. 6.2c. The lateral incidence angle of the laser beam was 90° , and the wire feed angle was kept at 5° , both relative to the surface of the top sheet substrate (XZ plane), as displayed in Fig. 6.2b. The LWB principal variables including P, TS, and WFS were selected within the ranges given in Fig. 6.2b, respectively 2–6 kW, 50–85 mm/s and 40–80 mm/s; these parameter ranges are typical of LWB in industrial settings.

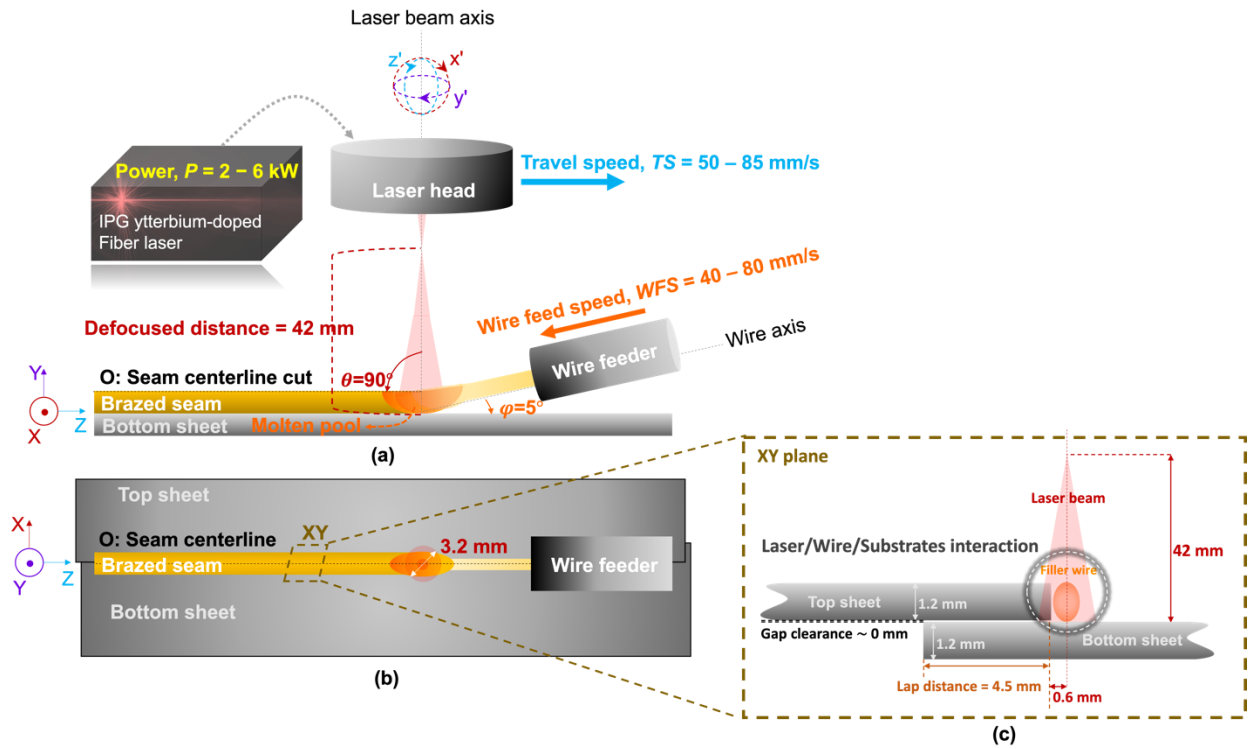


Fig. 6.2: Schematics of the LWB process in the lap joint configuration used in the present study. (a, b) LWB setup showing non-variable parameters including the spot size diameter, defocused distance, and laser/wire/substrate incidence angles. Experimental domains for the variable parameters such P, TS, and WFS are included. The defined XYZ coordinate system is aligned with the operational framework for subsequent microstructural and mechanical analyses. (c) The XY plane view field of the laser/wire/substrates interaction zone with the specified experimental dimensions.

The wire-adjusted heat input approach, originally introduced in our previous work [195], was employed to optimize the different variable combinations to achieve the desired braze

conditions. The process heat input, referred to as the $HI_{Relative}$, is calculated using Equation 6.1 [195], which accounts for all key LWB parameters simultaneously. The SI-IMC fraction is proportional to the amount of $HI_{Relative}$ up to a critical amount of heat, beyond which severe base metal melting occurs, shifting the predominate location of the SI-IMC formation from the braze region towards the base metal side of the braze/substrate interface. The primary objective of optimization in the present study is twofold: (1) Establishing the threshold for brazeability, ensuring complete wetting and the absence of surface defects. (2) Investigating SI-IMC morphology and distribution under varying conditions. The fraction of SI-IMCs correlates with the $\Delta HI_{Relative}$ thresholds with respect to a baseline [195]. The baseline for this study was defined as the lowest heat input that resulted in successful adhesion without defects. Equation 6.2 presents a modified form of $\Delta HI_{Relative}$ for a given condition “ i ”, where the baseline corresponds to the braze condition achieving defect-free adhesion, with the lowest $HI_{Relative}$. The LWB parametric set of P , T , and WFS were trialed for 13 conditions with varying $\Delta HI_{Relative}$.

$$HI_{Relative} = \frac{P}{TS+WFS} \quad (6.1)$$

$$(\Delta HI_{Relative})_i = (HI_{Relative})_i - (HI_{Relative})_{Baseline} \quad (6.2)$$

6.3.3 Microstructure Characterization

The braze microstructure and geometry for each LWB condition were examined by cross sectioning the LWB seam perpendicular to the brazing direction using waterjet cutting. An automated polishing procedure, featuring a reduced force setup, was applied to four cross-sections extracted from each LWB seam, following the method outlined in our previous work [179]. A Keyence VHX-7000 OM was used to observe the top surface of the braze seam and the polished cross-sections. The OM enables the preliminary microstructural analysis to explore the braze conditions with potential SI-IMCs. Any degree of substrate mixing can lead to SI-IMC formation. Therefore, in the OM analysis, regions where a high Fe content diffused into the braze were assumed as potential SI-IMCs during preliminary analysis.

6.3.4 Mechanical Testing

LWB joints of interest were evaluated through tensile testing, utilizing a DIC method to visualize and analyze strain distribution during deformation. Tensile samples were waterjet-cut to sub-size geometry following ASTM E8 [230], with the tensile direction set perpendicular to the brazing direction. The braze seam was positioned at the center of the reduced-width section of each sample.

To accommodate the 1.2 mm height difference between the top and bottom sheet substrates, shims were used to ensure proper gripping. Tensile tests were conducted using a Tinius Olsen universal testing frame equipped with a 10 kN load cell, applying a LSR loading condition at a crosshead speed of 1 mm/min. Strain distribution and localizations were analyzed using Correlated Solutions Vic-2D DIC software, incorporating a 25 mm virtual strain gauge for measurement. All strength measurements from the tensile tests were reported in force (N) rather than stress (MPa) due to the irregular cross-sectional area of the braze region and the varying mechanical properties of the base metals and braze, making accurate stress calculations impractical. Peak load and total displacement for each condition, reported as the mean of four repetitions, with error bars representing the 95% CI.

6.3.5 Fracture Analysis

The EBSD technique was used to comprehensively investigate microstructure and crystallographic orientation of the braze region and steel substrate from the cross section (XY plane) of the fractured samples. In order to prepare the samples for EBSD, a vibratory polishing step using colloidal silica suspension was added after the metallographic preparation procedure described in section 6.3.3. A high-speed Hikari charge-coupled device (CCD) camera was used for pattern acquisition at two step sizes with 1 μm and 0.5 μm for low and high magnifications, respectively on a Apreo 2 S LoVac field-emission scanning electron microscope (FE-SEM). The EBSD was operated at an accelerating voltage of 20 kV and a beam current of 13 nA. The EBSD results were post-processed using AZtecCrystal software. The Fe-BCC phase input database in EBSD aligned with any region with high Fe content including substrates and IMCs,

and the Cu-FCC phase correlated with the Cu-rich region of the braze. The EBSD system enabled simultaneous detection of elements through the energy dispersive X-ray spectroscopy (EDS) detector for elemental mapping. The fracture surface morphology of LWB joints were characterized using Zeiss LEO 1530 FE-SEM at an accelerating voltage of 20 kV and an operating distance of approximately 9.8 to 10.8 mm, and a beam current of 3 nA.

6.4 Results and Discussion

6.4.1 Producing Braze Joints with Various SI-IMC Morphologies

The optimized LWB window for producing braze seams, containing various SI-IMCs correlated with the $\Delta HI_{Relative}$ values, are shown in Fig. 6.3a, with detailed process parameters listed in Table 6.2. According to the wire-adjusted heat input approach, the $\Delta HI_{Relative}$ values are the function of the $HI_{Nominal}$ and HI_{Wire} of each condition, outputting as $\Delta HI_{Relative}$ levels based on the detailed formulas that can be found in our previous work [195]. Within this window, a zero $\Delta HI_{Relative}$ corresponds to the baseline condition (LWB-0) with $HI_{Relative}$ of 32.52 J/mm; this is the required HI to melt the filler wire sufficiently to complete the wetting process without any notable surface defects, such as lack of adhesion and blow holes, as seen from the braze seam top view in Fig. 6.3b. It is worth noting that this required $HI_{Relative}$ (32.52 J/mm) value determined in the present study for the baseline condition in the lap joint configuration is approximately equal to the reported $HI_{Relative}$ (32 J/mm) in our previous work [195] that was needed to eliminate both surface and sub-surface defects in a roof joint configuration using GI-IF steel substrates. However, the baseline criteria in the present work are different, as sub-surface defects were not taken into consideration. Specifically, although 32.52 J/mm is the lowest allowable brazing HI, pores and substrate mixture are observed from the cross-sectional view of the braze seam in the LWB-0 condition, as given in the XY plane view in Fig. 6.3b. The substrate mixture in the baseline condition is attributed to the laser/wire/substrate positioning in the experimental lap joint configuration, which led to the intentional melting of the top sheet in the present study, so as to increase the chances of SI-IMC formation.

Further increases in the $\Delta HI_{Relative}$ beyond the baseline resulted in notable changes to the distribution and morphology of the Fe-rich phases within the braze seam. LWB seams were selected from four $\Delta HI_{Relative}$ levels ranging from 2.3 J/mm to 17.5 J/mm. The top view images and corresponding cross-sectional microstructure of the braze seams of selected $\Delta HI_{Relative}$ confirm that these joints exhibited no surface defects, as seen in Fig. 6.3c-f, while various morphology and distribution of the SI-IMCs are expectable, as evident from the substrate mixture and Fe-rich phases within the braze region. The purple-colored contour lines and arrows in Fig. 6.3b-f indicate where the melting from the top sheet advanced into the braze region to form Fe-rich phases with different morphologies; some formed as an island and some as peninsula, while others dispersed as small particles, as summarized schematically in Fig. 6.3g. The formation of Fe-rich peninsulas made it challenging to quantify SI-IMC density for each LWB condition. Therefore, the braze joints were categorized based on their corresponding $\Delta HI_{Relative}$ levels.

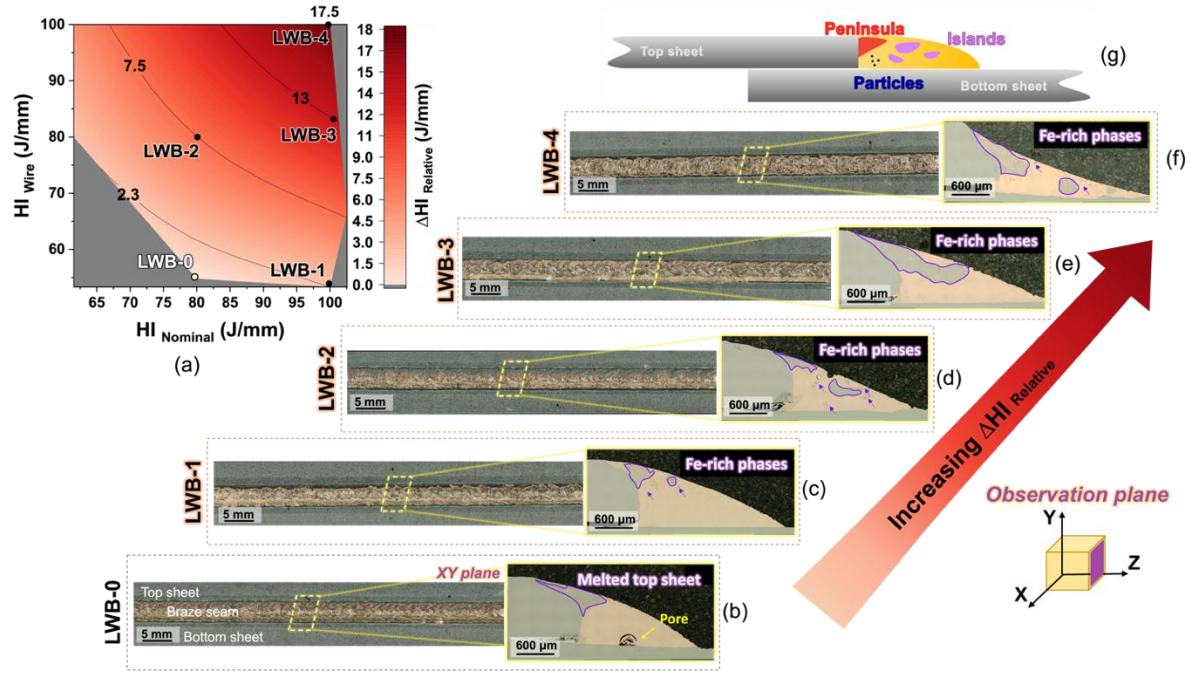


Fig. 6.3: (a) The optimized LWB window to achieve SI-IMCs with various morphologies in the lap joint configuration for Gen 3 Q&P980 substrates. The top view and cross-sectional microstructure of the LWB seams of the (b) baseline condition (LWB-0), as well as the conditions with four $\Delta HI_{Relative}$ levels (c) 2.3 J/mm, (d) 7.5 J/mm, (e) 13 J/mm, and (f) 17.5 J/mm. Purple annotations in (b-f) indicate Fe-rich phases within the braze matrix. (g) Schematic diagram showing the variety of expected SI-IMCs.

Table 6.2: Detailed LWB process parameters of conditions with various $\Delta HI_{Relative}$.

Condition , LWB-i	P (kW)	TS (mm/s)	WFS (mm/s)	$HI_{Nominal}$ (J/mm)	HI_{Wire} (J/mm)	$HI_{Relative}$ (J/mm)	$\Delta HI_{Relative}$ (J/mm)
LWB-0	4	50	73	80	54.48	32.52	0
LWB-1	4	40	75	100	53.33	34.82	2.3
LWB-2	4	50	50	80	80.00	40.02	7.5
LWB-3	5	50	60	100	83.33	45.52	13.0
LWB-4	4	40	40	100	100.00	50.02	17.5

Geometrical measurements of the braze seams of the selected LWB conditions across the four cross-sections of the seam are depicted in Fig. 6.4, following the assumptions for geometrical parameters in Fig. 6.4a. Each braze cross-section consists of two braze/substrate interfaces: the top sheet (T-side) and the bottom sheet (B-side). The braze/substrate interface on the B-side was quantified as the braze adhesion length (L_1). The measurement of the adhesion length from the T-side was found to be challenging due to the melting of the top sheet, which resulted in the appearance of Fe-rich phases with peninsula morphology, for instance. In particular, peninsula formation prevented the clear definition of a boundary between the braze (Cu matrix) and the substrate. To address this, a geometrical parameter, $L_{2,Net}$, was defined to accurately quantify the vertical length of the braze, as the T-side interface may be either smaller or larger than the top sheet thickness depending on the amount of the braze material deposition. This parameter accounts for both the braze area and L_1 by approximating the braze region as a 90-degree (right angle) triangle, enabling $L_{2,Net}$ to be calculated using Equation 6.3.

$$L_{2,Net} = \frac{2 \times A_{Braze}}{L_1} \quad (6.3)$$

In this assumption, Equation 6.3, A_{Braze} represents the actual braze area, encompassing any penetration of the Cu-rich matrix into the substrates, as well as the presence of the Fe-rich matrix within the braze region. As shown in Fig. 6.4d, the calculated $L_{2,Net}$ consistently aligns with the trends observed in A_{Braze} (Fig. 6.4c) rather than L_1 (Fig. 6.4b). This geometrical

parameter serves as a reliable metric for evaluating the influence of the joint geometry on the subsequent mechanical performance of the selected braze joints.

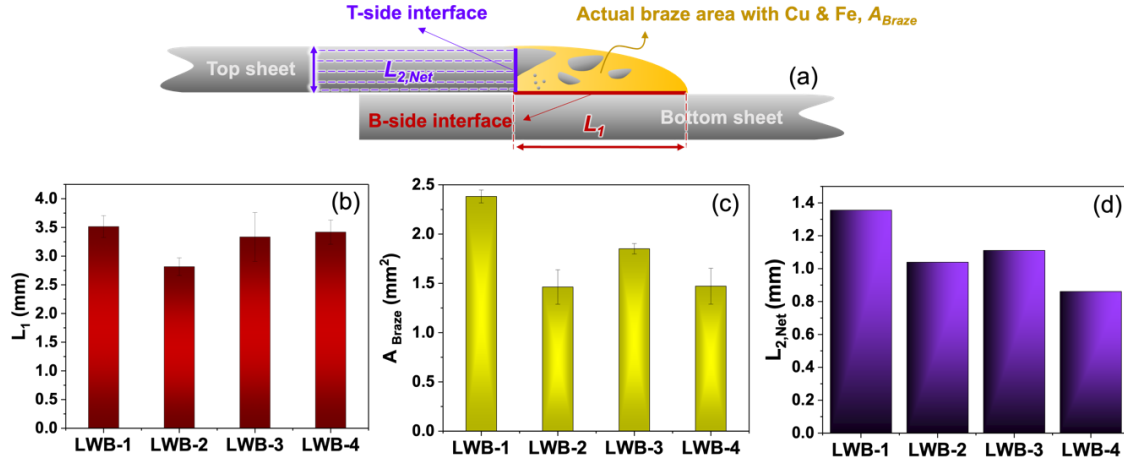


Fig. 6.4: (a) schematic diagram of the defined geometrical parameters for the lap joint configuration with Fe and Cu mixed within the braze region and interfaces. Geometrical measurements of LWB-1 to LWB-4 for: (a) L_1 , B-side interface length, (b) A_{Braz} , the actual braze area including Fe-rich phases and any progression of Cu into the substrates, and (d) $L_{2,Net}$, geometrical parameter enabling measurement of the interface length from the T-side with melted top sheet calculated from Equation 6.3.

6.4.2 Mechanical Performance of SI-IMCs-included LWB Joints

Tensile testing of the LWB joints, containing various SI-IMCs morphologies and distributions, showed a range of strengths. The representative force-displacement curves in Fig. 6.5a showed that the joint strengths (averaged over four repetitions) ranged from 587 N to 1836 N, as displayed in Fig. 6.5b and Table 6.3. Condition LWB-1, with the lowest $\Delta HI_{Relative}$, resulted in the highest mechanical performance with the highest peak load, total displacement, and toughness of 1836 ± 49 N, 0.36 ± 0.02 mm, and 0.41 ± 0.03 J, respectively. However, it was seen that an increase in $\Delta HI_{Relative}$ to 13 J/mm drastically degraded the mechanical performance of the joint performance; LWB-3 had a strength of 587 ± 76 N, a displacement of 0.07 ± 0.02 mm, and toughness of 0.02 ± 0.01 J. There is a direct correlation between mechanical performance and $\Delta HI_{Relative}$ up to 13 J/mm. However, with a further increase in $\Delta HI_{Relative}$ to

17.5 J/mm, the strength and total displacement slightly improved; the LWB-4 condition had a strength of 685 ± 97 N, displacement of 0.08 ± 0.02 mm, and toughness of 0.03 ± 0.01 J. The trend in strength variation across different LWB conditions was analyzed and compared with the variations in their geometrical factor, $L_{2,Net}$, allowing the classification of braze conditions into two distinct groups: LWB-1 and LWB-2, which exhibit improved mechanical properties and are classified as “high performance”, and LWB-3 and LWB-4, which demonstrate degraded mechanical properties and are “low performance”. These distinctions are visually represented by green and red colors in Fig. 6.5 and Table 6.3. Comparing the minimum values of the “high performance” and “low performance” groups, the LWB joints made with $\Delta HI_{Relative}$ of 2.3-7.5 J/mm had strength, displacement, and toughness values that were 91.7%, 319.8%, and 1200% greater, respectively, than the joints made with $\Delta HI_{Relative}$ of 13-17.5 J/mm. In our previous work [195], it was found that increasing the SI-IMC area fraction from $1.2 \pm 2.4\%$ to $38.2 \pm 16.9\%$ improved the mechanical performance of the LWB joints in a double-flanged joint configuration in GI-IF steels, resulting in a 14% increase in the tensile peak load and a 350% increase in the total displacement, and 525% increase in toughness; however, as no cracks propagated within the braze region, the role of the SI-IMCs in fracture behavior was unclear. In the present study, failure occurred within the braze containing a high density of SI-IMCs, due to the highly crack-susceptible braze region combined with the high failure resistance of the Gen 3 Q&P 980 substrate. Compared to our previous study on LWB in GI-IF materials [195], the change in the base material (to Gen 3 Q&P 980), that led to movement of failure into the braze, both improved and degraded the mechanical properties, depending on the heat input. These findings clarify how LWB parameters shape braze microstructure and mechanical performance, advancing the understanding of process-structure-performance relationships in LWB joints.

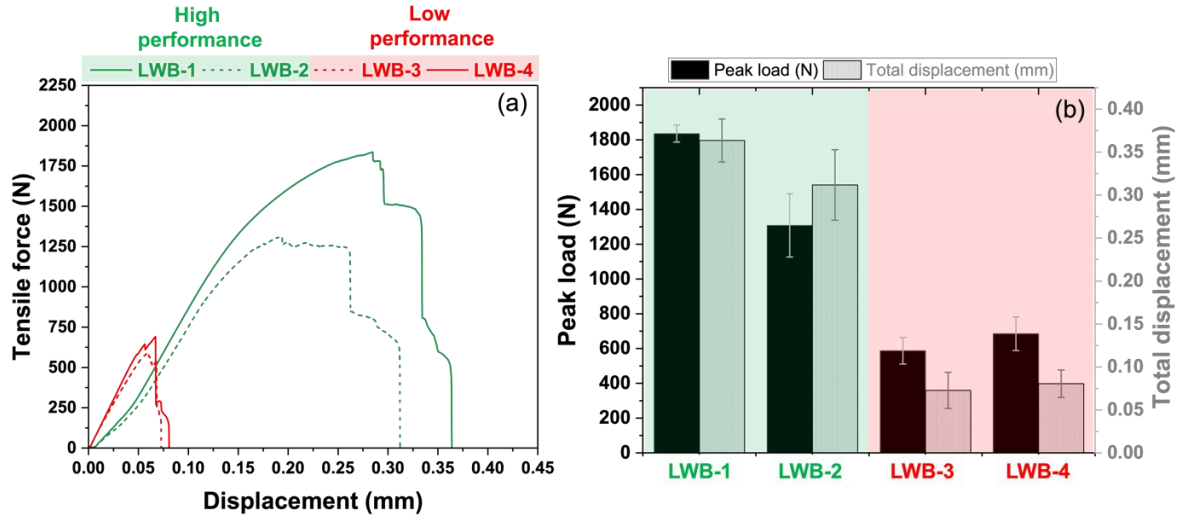


Fig. 6.5: (a) Representative tensile force versus displacement curves of the LWB joints containing SI-IMCs formed under various $\Delta HI_{Relative}$ levels. (b) Peak load and total displacement for each condition, averaged over four repetitions, with error bars representing 95% CI.

Table 6.3: Mechanical properties of the LWB joints. Peak load and total displacement values represent the averages from four tensile test repetitions. Deviations from the mean are reported with 95% confidence intervals. The max Hencky strains were measured from the DIC strain distribution maps of the stage before failure.

Condition, LWB-i	Peak Load (N)	Total displacement (mm)	Toughness (J)	Max Hencky Strain (%)
LWB-1	1836 ± 49	0.36 ± 0.02	0.41 ± 0.03	24.85 ± 1.07
LWB-2	1308 ± 182	0.31 ± 0.04	0.26 ± 0.06	12.88 ± 1.56
LWB-3	587 ± 76	0.07 ± 0.02	0.02 ± 0.01	1.87 ± 0.98
LWB-4	685 ± 97	0.08 ± 0.02	0.03 ± 0.01	5.23 ± 0.92

To examine the strain distribution during tensile testing, the strain distribution maps were obtained from the DIC analysis. The maps correspond to three stages of deformation, where (1) zero strain, (2) onset of strain localization, and (3) maximum strain localization occurred; these stages are associated with the conditions before tensile loading, crack initiation, and at the final stage before failure, respectively. Fig. 6.6 illustrates the strain distribution maps, and the cross-sectional microstructures, clearly highlighting the crack propagation path in the failed

samples. A key similarity across all of the DIC maps is the initial strain localization, which identifies the crack initiation sites. In all conditions, strain first accumulated in the braze root region, as indicated by the yellow arrow. According to Khan et al. [113], this location is critical and aligns with expectations for lap joint configurations, where there is nearly no clearance gap between the top and bottom sheets, combined with the accumulation of brittle Cu-Zn alloy, creating a high-stress zone.

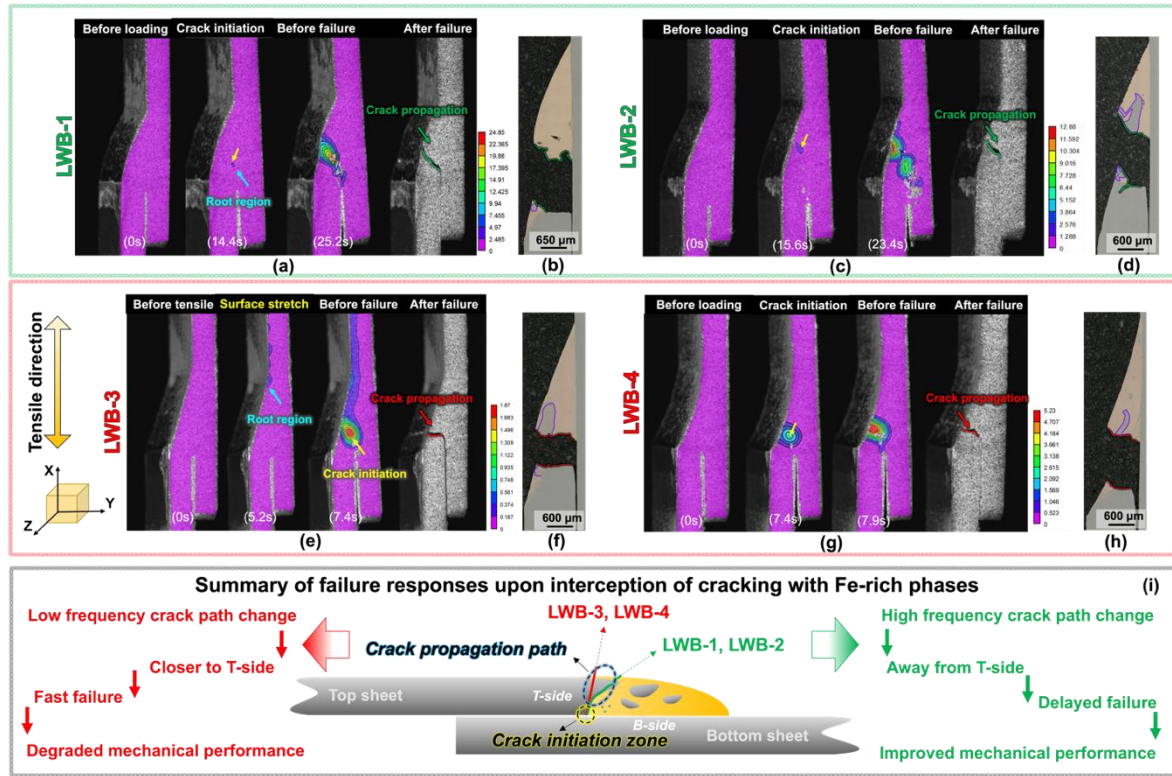


Fig. 6.6: (a, c, e, g) DIC strain distribution maps during tensile testing, quantified as Henky strain (% ϵ_{xx}). The “after failure” stage only indicates the crack propagation path with no superimposed strain map. (b, d, f, h) Post-failure microstructural analysis of the fracture surfaces, illustrating interactions between the crack propagation path (green lines in LWB-1 and LWB-2, and red lines in LWB-3 and LWB-4) and the Fe-rich phases (purple lines). (i) Schematic diagram showing correlation between the crack propagation path and mechanical performance affected by the LWB conditions.

Crack initiation occurred at 14.4 s and 15.6 s for LWB-1 and LWB-2, respectively, (Fig. 6.6a and c), while it occurred faster in LWB-3 and LWB-4, only taking 7.4 s in both cases (Fig.

6.6e and g). Notably, in LWB-3 with the lowest mechanical performance, crack initiation and propagation occurred so rapidly that distinct frames were not capture of both the crack initiation and failure stages; prior to the crack initiation, base metal surface stretching occurred adjacent to the braze toe at the B-side, as seen from Fig. 6.6e. The underlying microstructure of the fracture surfaces confirm the presence of Fe-rich phases or SI-IMCs within the cracking path (Fig. 6.6b, d, f, and h). Despite failure occurring closer to the T-side interface in all cases, the crack propagation paths varied, where failure progressed through the braze bulk with frequent propagation path changes more in LWB-1 and LWB-2 conditions compared to LWB-3 and LWB-4, where failure proceeded with fewer changes in propagation path, directly proceeding to the braze top surface, as illustrated in the schematic in Fig. 6.6i. The changes in the crack propagation path also align with the fluctuations observed during the load drops after the peak load. As observed from Fig. 6.5a, these fluctuations span a broader displacement range in LWB-1 and LWB-2, compared to LWB-3 and LWB-4. According to the literature, even the sole presence of IMCs and precipitates within the crack path is sufficient to identify them as crack initiators and propagation accelerators in some studies [21,115]. However, the variations in the crack propagation paths observed in the presence of SI-IMCs with different morphologies suggest that the sharp differences in the mechanical response may be influenced by these phases. Rather than acting solely as crack initiators and propagation accelerators, SI-IMCs appear to modulate the crack propagation path, emphasizing their role as crack-path-altering agents.

6.4.3 Identification and Classification of IMCs near/along the Crack Propagation Path

The nature and morphology of the Fe-rich phases intersecting the crack propagation path were characterized using phase, elemental distribution, and KAM analyses on the T-side fracture of the LWB samples, as they experienced the highest strain localization. In LWB-1, the crack initially propagated from region “R1”, where a high fraction of Fe-rich particles is dispersed adjacent to the braze/substrate interfacial layer, as seen from the band slope image and phase map in Fig. 6.7a. The phase maps clearly distinguish between the Cu-rich and the Fe-rich matrices. The expectable phases such as the well-known Fe-Si IMCs (e.g. Fe_2Si , Fe_5Si_3 , FeSi_2 ,

Fe₂Si, FeSi as reported in [20,21,24]) could not be distinguished separately within the Fe-rich matrix due to overlapping patterns of their signals in EBSD analysis. However, the elemental maps in Fig. 6.7a could reveal significant segregation of Fe and Si within the interfacial layer and the Fe-rich particles, enabling detection of Fe-Si phases. According to Akbarian et al. [195], as discussed in Chapter 5, the segregation of Si in regions with high Fe concentrations indicates the formation of Fe-Si IMCs, classified as SI-IMCs and I-IMCs, depending on their location within the joint.

This study further revealed that Fe-rich phases with high Si segregation, identified as SI-IMCs or I-IMCs, exhibit the lowest dislocation activity, as observed from the KAM map in Fig. 6.7a. The KAM values in the Fe-Si IMC regions are approximately zero due to minimal strain accommodation within these phases, indicating that they are stoichiometric phases that will be brittle in nature. Due to the brittle behavior of IMCs, these phases are not deformable and dislocations cannot form within them, as seen in the nearly zero KAM regions of the maps. This work provides the first direct evidence that strain accommodation is limited within IMCs. In region “R2”, the crack partially propagated along the interface of a bulky spherical Fe-rich island containing Cu-rich globules, as shown in Fig. 6.7b. Elemental and KAM analyses confirm that significant Si segregation occurred at the periphery of this globular island, forming an Fe-Si IMC shell surrounded by the Cu-rich matrix of the braze. The crack did not penetrate the spherical Fe-rich island and its surrounding Fe-Si IMC shell, instead, it deviated from the interface of this shell and propagated within the Cu-rich matrix of the braze.

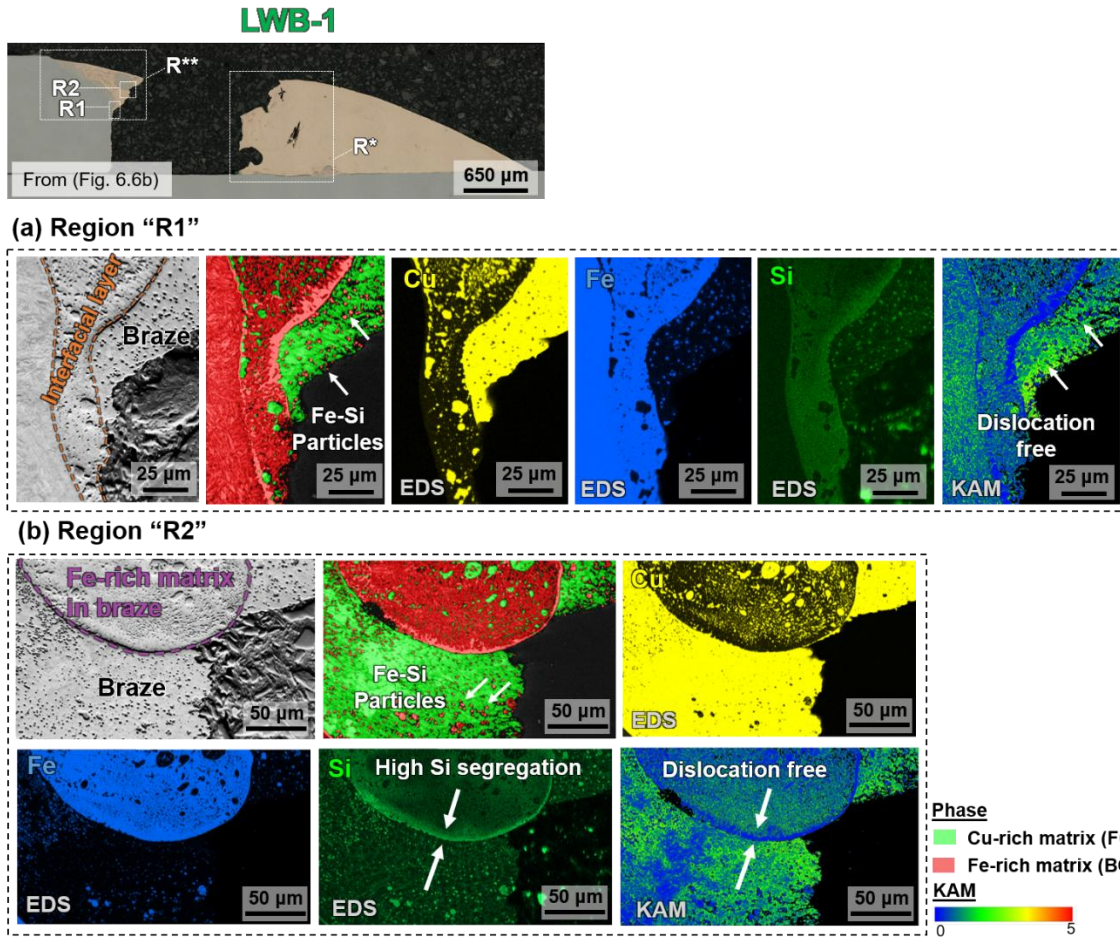


Fig. 6.7: EBSD mapping of the near crack propagation path of the LWB-1 joint after tensile fracture from regions “R1” and “R2” in (a,b) that show band slope micrographs, phase maps indicating Cu-rich and Fe-rich matrices, EDS elemental maps indicating Fe and Si segregation within Cu matrix and KAM maps. The regions “R*” and “R**” were analyzed in Fig. 6.12 and 6.13.

Similar to LWB-1, IMCs identified with high Fe-Si segregation and zero KAM are observed in LWB-2, as seen from the “R1” and “R2” regions in Fig. 6.8. In region “R1”, SI-IMC particles are present within the initial stage of crack propagation, Fig. 6.8a. In region “R2”, Fig. 6.8c, the crack intersected an anchor-shaped SI-IMC peninsula shelled by an Fe-Si IMC layer at its periphery and within the Fe-rich phase in the braze. The SI-IMC peninsula in LWB-2 formed as a fully dense Fe-rich structure, whereas the spherical SI-IMC structure in LWB-1 contained Cu-rich globules within the Fe-rich matrix.

Instead, Si was evenly distributed within the Fe-rich matrix, and the Cu-rich globules solidified within the Fe-rich matrix. KAM analysis showed no defined regions with zero KAM, except in the Cu-rich matrix, where deformation was limited due to the constraint from its surrounding Fe-rich matrix. The type of SI-IMC formed in LWB-3 can be identified as a (Fe-Si)Cu phase with a network-shape morphology, indicative of the immiscibility of Fe and Cu at elevated temperatures (between 1361 K-1700 K according to Wang et al. [231]).

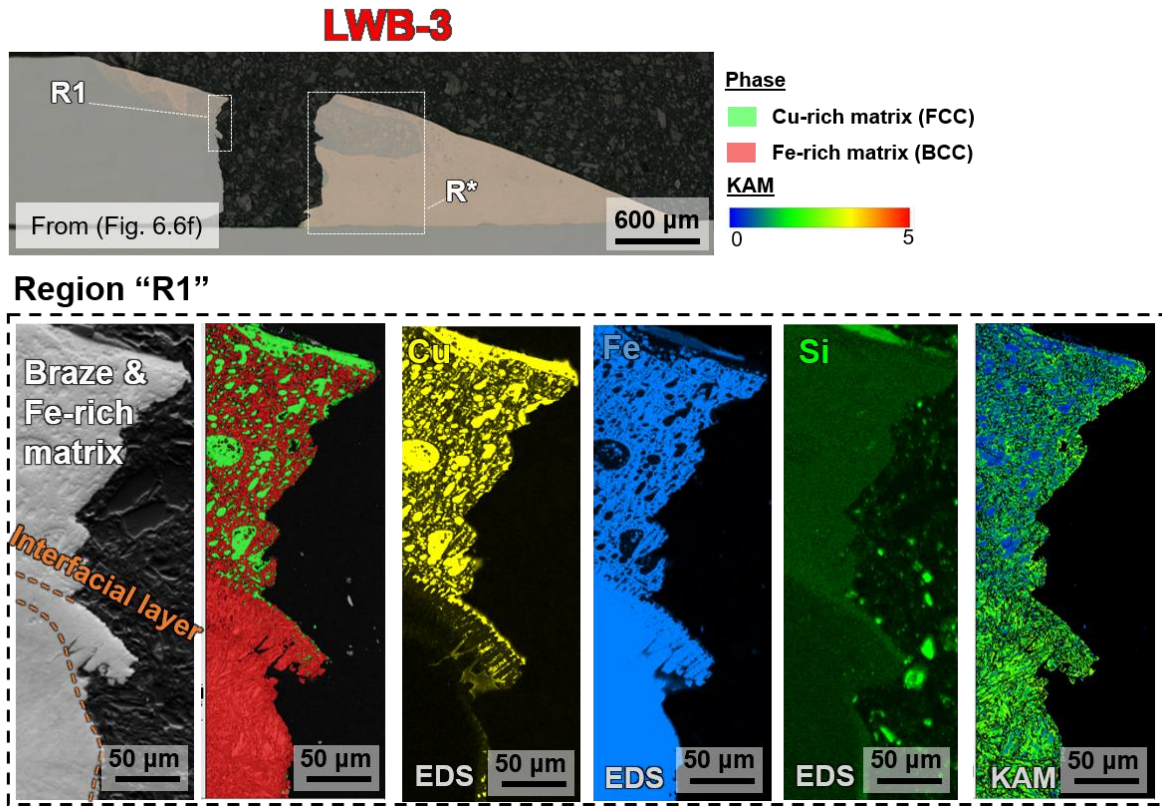


Fig. 6.9: EBSD mapping of the near crack propagation path of the LWB-3 joint after tensile fracture, consisting of the band slope micrograph and the phase, elemental, and KAM maps.

In LWB-4, fine Fe-Si SI-IMC particles are present within the crack propagation path, closer to the crack initiation site in region "R1", but the crack subsequently propagated towards the interface of larger Fe-Si particles, as shown in Fig. 6.10a. These SI-IMCs with coarsened spherical morphology form/grow during LWB; specifically, LWB-4 involved the highest HI amongst all the conditions studied, which allowed sufficient Si segregation for IMC growth.

In addition to the coarsened SI-IMC particles in LWB-4, the (Fe-Si)Cu phase (network-shape SI-IMC similar to LWB-3) formed within the braze in the B-side (Fig. 6.10b), where it was observed to cause internal cracking.

Fig. 6.11 illustrates a summary of the identified SI-IMC structures with their crack propagation response. In the present study, the identified SI-IMC structures are designated as α , β , γ , and δ , distinct from the nomenclature used for common alloys in the literature. The Fe-Si particles with a small size below 3 μm in diameter are called fine α -SI-IMC (observed in LWB-1, LWB-2, and LWB-4), while those above 3 μm in diameter are coarse α -SI-IMC (LWB-4). The large spherical Fe-rich island containing Cu-rich globules surrounded by the Fe-Si IMC shell in LWB-1 is different from the coarse α -SI-IMC, classifying as a new type named β -SI-IMC. The SI-IMC with anchor-shaped peninsula are formed with two types: anchor-shaped Fe-rich peninsula surrounded by an Fe-Si IMC shell in LWB-2 (named γ -SI-IMC), and large immiscible (Fe-Si)Cu peninsula with network-shape in LWB-3 (named δ -SI-IMC). The δ type SI-IMC formed as an island in LWB-4.

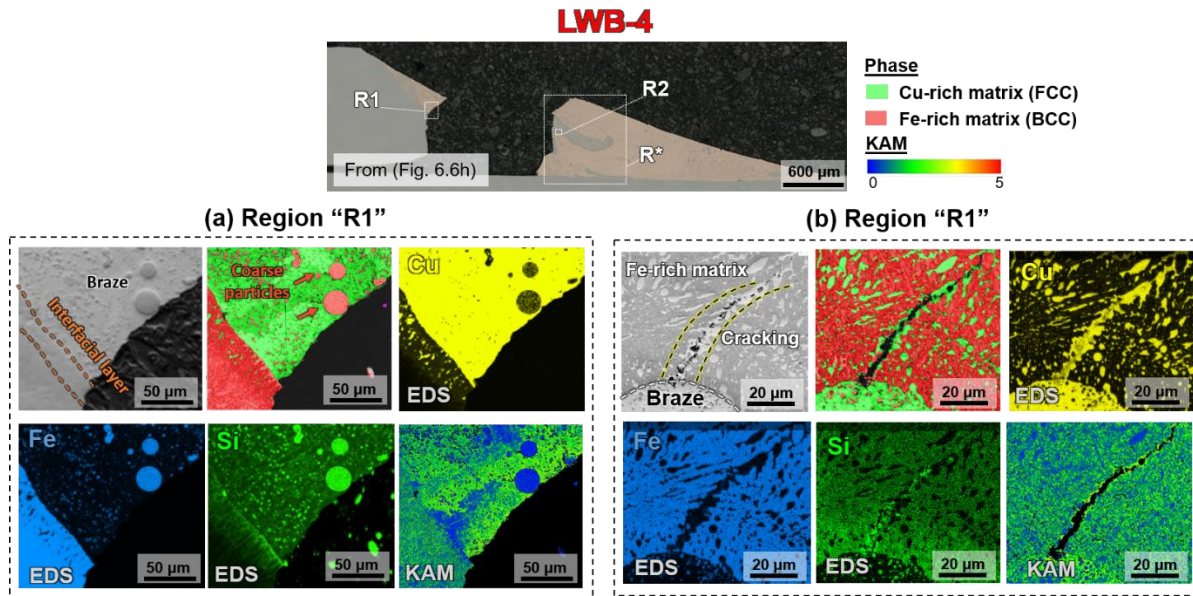


Fig. 6.10: EBSD mapping of the near crack propagation path of the LWB-4 joint after tensile fracture, consisting of the band slope micrographs, phase maps, elemental maps, and KAM maps that identify the SI-IMCs in regions "R1" and "R2".

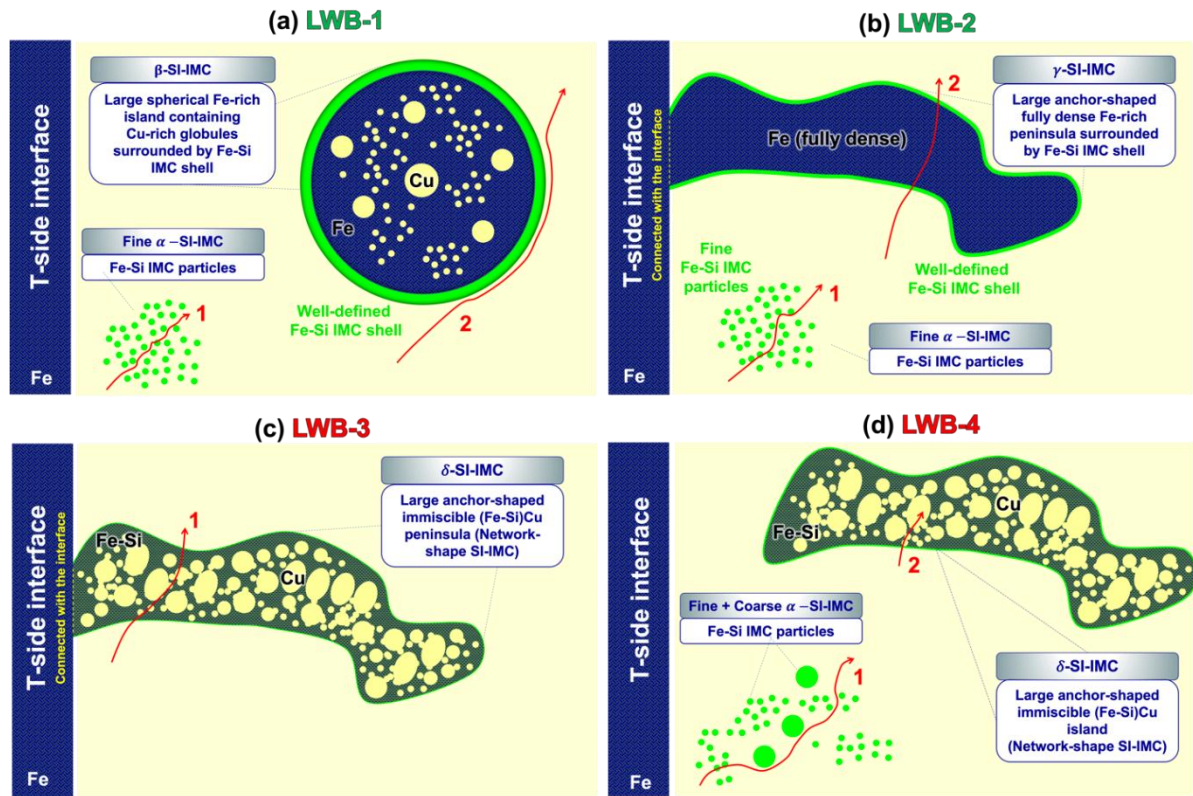


Fig. 6.11: Schematics diagram showing SI-IMCs with various morphologies and chemistry that appeared near/along the crack propagation path (indicated by red lines) of the LWB-1, LWB-2, LWB-3, and LWB-4 joints. Red lines numbered “1” refer to the path of the crack closer to the crack initiation site in (a, b, and d) and the fast occurring crack propagation in (c). Red lines numbered “2” refer to the regions away from crack initiation in (a and b), and internal cracking in (c).

6.4.4 Analyzing the Microstructure and Texture of the Braze and SI-IMCs of the Fractured Samples

Fig. 6.12 displays the phase and inverse pole figure (IPF) phase maps of the underlying microstructure of the fracture surface of the braze joints LWB-1, LWB-2, LWB-3, and LWB-4, along with their corresponding pole figure (PF) analysis from the Cu-rich matrix. The region of interest for this investigation is “R*” as defined in Fig. 6.7-6.10. In LWB-1 and LWB-2, the Cu-rich matrix displays a non-uniform grain distribution with a pronounced bimodal grain structure and favorable orientational alignment, as respectively displayed in Fig. 6.12a and b.

Although large pores and internal cracks with significant openings are present within the Cu-rich matrix, no internal cracking is observed at the B-side interface or at the interfaces between the Cu-rich and Fe-rich phases. This suggests that the interfacial adhesion between the Cu matrix, the braze/substrate, and the Fe-rich matrix regions is sufficiently robust to resist interface-induced cracking, influenced by the type, morphology, and orientational distribution of the SI-IMCs formed under these conditions.

By contrast, the Cu-rich matrix of LWB-3 and LWB-4 joints (Fig. 6.12c and d), exhibits a more uniform grain distribution and less pronounced bimodal structures, along with a weaker orientational distribution of grains and phases. In LWB-3, unlike in LWB-1 and LWB-2, no pores or internal cracks are present within the Cu and SI-IMC matrices. In LWB-4, while no pores are present, cracking initiates from the interface of the network-shape δ -SI-IMC peninsula, which highlights weakness of the interfacial bonding at the interface of the δ -SI-IMC with the Cu-rich matrix. Notably, both LWB-3 and LWB-4 exhibited very limited displacement due to the rapid initiation and propagation of cracks. This may explain why the Cu-rich matrix remains crack-free, experiencing limited deformation, and the internal crack within the SI-IMC does not propagate further. In fact, the internal cracking within the δ -SI-IMC didn't propagate further, not due to crack blunting, but rather because of premature failure from the main crack and brittleness of the IMCs, which were unable to accommodate strain.

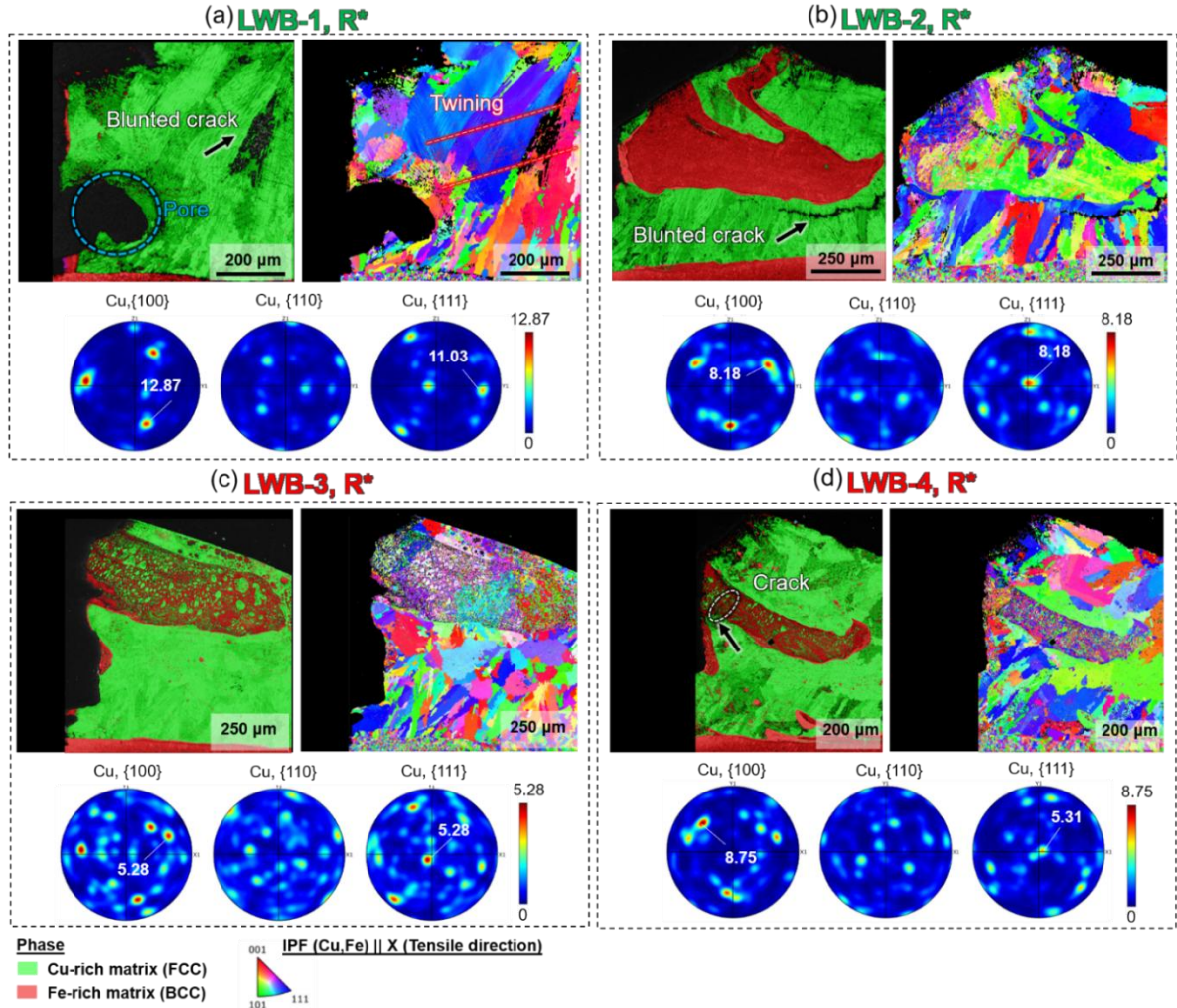


Fig. 6.12: Broad-view EBSD maps of the B-side fracture surface showing the underlying microstructure of the LWB-1, LWB-2, LWB-3, and LWB-4 joints. The phase maps indicate the Cu-rich (green) and Fe-rich (red) phases. The IPF maps are along the tensile direction (X-axis) for both the Cu-rich and Fe-rich matrices. The corresponding PFs show the orientation distributions of the {100}, {110}, and {111} planes in the Cu-rich matrix. The "R*" regions are defined in Fig. 6.7-6.10.

Notably, cracks within the Cu-rich matrix in the LWB-1 and LWB-2 samples did not propagate further due to sufficient crack blunting, likely influenced by the grain texture. The large crack in LWB-1 occurred in a grain, where adjacent to this grain, mechanical twins are widely formed, acting as an obstacle against further crack propagation (as seen from the IPF map in

Fig. 6.12a). Accordingly, the presence of internal cracks in the Cu matrix indicates strain localization in this region; moreover, crack blunting, accompanied by twinning, suggests that the Cu-rich matrix in LWB-1 has a high capacity to accommodate localized strain. Crack branching and blunting in the softer (Cu-rich) matrix of the braze is responsible for the observed delayed failure in LWB-1 and LWB-2. Meanwhile, the primary crack path leading to failure was closer to the T-side. It appears that the propagation of the primary crack was slow enough to allow further strain localization within the Cu-rich matrix at the B-side. The presence of distinct SI-IMCs, formed both adjacent to and within the primary crack path, may have contributed to this process.

The phase-level PF analysis of the Cu-rich (FCC) matrix reveals that the $\{100\}$ plane has the highest intensity in LWB-1 (12.87) and the lowest in LWB-3 (5.28), as shown in Fig. 6.12a and c, following the overall trend of $\{100\}$ intensity: LWB-1 > LWB-2 ~ LWB-4 > LWB-3. It should be noted that although the $\{100\}$ PFs of LWB-2 and LWB-4 show similar peak intensities of 8.18 and 8.75, respectively (Fig. 6.12b and d), they differ in texture, as observed from the hot zone distribution in their $\{100\}$ PFs, where a weaker texture has formed in LWB-4. The hot zones of $\{100\}$ PFs in each of LWB-1 and LWB-2 appear only in three distinct spots, indicating the development of a strong, aligned texture. By contrast, the hot zones of $\{100\}$ PFs in LWB-3 and LWB-4 appear duplicated, forming neighboring spots and resulting in a more scattered distribution, suggesting a more randomized and weaker texture. It is worth mentioning that the $\{100\}$ plane is a well-known solidification texture in FCC metals because grains with this orientation have the lowest interfacial energy with the liquid, making it the most energetically favorable growth-direction during solidification [232–234]. The observed trend in the $\{100\}$ PFs of Cu-rich matrix correlates strongly with the mechanical properties. A weak randomized texture limits effective stress redistribution, leading to a lower load-bearing capacity. Similarly, Ahmadikia et al. [235] reported that a randomized texture reduces strain localization by promoting isotropic deformation; however, it also led to inefficient load transfer and lower mechanical strength due to the lack of preferential grain alignment. Moreover, from the $\{111\}$ PFs of the Cu-rich matrix, a stronger texture is observed in LWB-1 and LWB-2 with

higher peak intensities (11.03 and 8.18, respectively) than in LWB-3 and LWB-4, with lower peak intensities (each with 5.30). The $\{111\}$ orientation in FCC metals is responsible for activating twinning mechanisms. A higher intensity of $\{111\}$ PFs in LWB-1 and LWB-2 facilitates the observed mechanical twins, responsible for internal crack blunting in LWB-1. Twinning is known to serve as a key deformation mechanism to delay crack propagation and improve toughness [236–239]. Conversely, the relatively lower $\{111\}$ intensity in LWB-3 and LWB-4 suggests lower capability of twinning, diminishing their mechanical resilience.

Fig. 6.13 shows the phase-level texture analysis for the Fe-rich matrix (BCC), which enables comparing the texture of various SI-IMC structures. For SI-IMC texture analysis in LWB-1, the T-side of the braze was analyzed to account for the β -SI-IMC that only appeared in this side of the fracture sample, while the B-side was analyzed in joints LWB-2, LWB-3, and LWB-4. The PFs demonstrated the development of a strong $\{100\}$ texture in LWB-1 and LWB-2, having peak intensities of 9.05 and 7.97, respectively, as shown in Fig. 6.13a. By contrast, LWB-3, as shown in Fig. 6.13c, exhibited a significantly weaker and more randomized $\{100\}$ texture, while LWB-4, shown in Fig. 6.13d, presented a moderate texture. As deformation of IMCs is not likely to occur, the texture of the SI-IMC structure primarily reflects its solidification history rather than deformation. The $\{100\}$ plane developed the strongest texture in both the Cu-rich and Fe-rich matrices, under conditions associated with improved mechanical performance (LWB-1 and LWB-2). Otherwise, strengthening the $\{100\}$ texture in the Fe-rich matrix enhances its alignment with the Cu-rich matrix by reducing misorientation between the two matrices, which aligns the texture of the SI-IMCs with the Cu-rich braze matrix. The alignment of the lattices of the two structures promotes strong interfacial bonding and efficient load transfer, preventing crack initiation from the SI-IMCs, as observed in LWB-1 and LWB-2. This contrasts with LWB-3 and LWB-4, which had weaker textures. In LWB-1, the combination of fine α -SI-IMC and β -SI-IMC, exhibited the highest texture alignment with the Cu-rich matrix. This alignment slightly decreases in LWB-2, where the SI-IMC structures were composed of a combination of fine α -SI-IMC and γ -SI-IMC. The δ -SI-IMC in

LWB-3 and the coarse α -SI-IMC in LWB-4 resulted in the weakest texture alignment between the matrices, leading to weaker interfacial bonding.

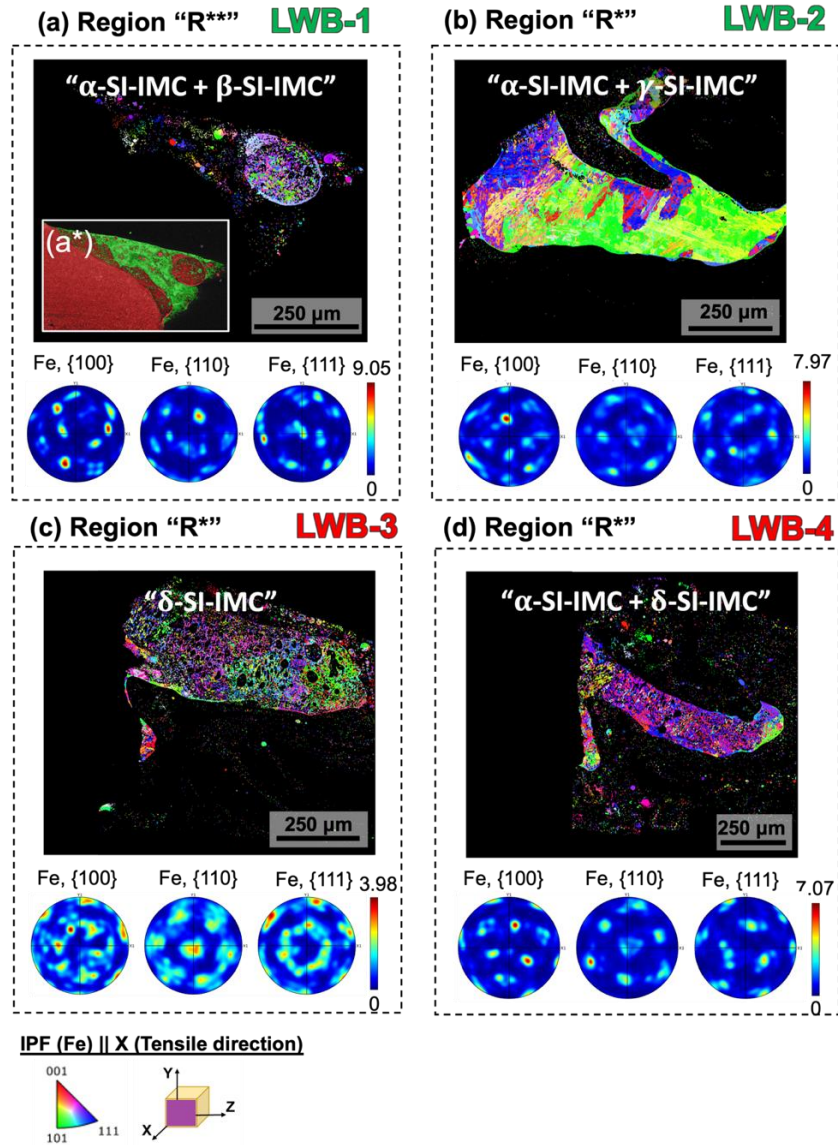


Fig. 6.13: Broad-view phase level IPF and PF analyses of the Fe-rich matrix from the fracture surface showing the underlying microstructure of the LWB-1, LWB-2, LWB-3, and LWB-4 joints. The IPF maps are along the tensile direction (X-axis). The corresponding PFs show orientation distributions of {100}, {110}, and {111} planes in the SI-IMC structures. (a*) corresponding phase map of (a). The "R**" and "R*" regions are defined in Fig. 6.7-6.10.

The brazing process and subsequent tensile testing both introduce dislocations within the braze microstructure. To further elucidate how the texture alignment between the SI-IMCs and the Cu-rich matrix correlates with the mechanical performance, dislocation activity in both the Cu-rich matrix and Fe-rich matrix (SI-IMCs) was analyzed. A geometrically necessary dislocation (GND) analysis was conducted using AZtecCrystal software to quantify dislocation activity by estimating the density of dislocations required to accommodate local lattice curvature. GND density enables assessing the distribution of plastic deformation at a finer scale from orientation gradients obtained from the EBSD data. Fig. 6.14 presents the GND distribution maps of the braze after it fractured during the tensile testing. The GND maps indicate that the braze region, consisting of Cu-rich and Fe-rich matrices, had dislocation densities ranging from 0 to $6.45 \mu\text{m}^{-2}$, depending on the area. Notably, the analysis was conducted on both the T-side and B-side brazes for accuracy in LWB-1 due to β -SI-IMC only being present on the T-side of the failure, while the B-side braze was considered for the remaining samples. Considering that the Fe-rich matrix is less prone to deformation due to the brittleness of IMCs formed in this matrix, it is assumed that the generated dislocations are primarily attributed to solidification, as it contains a high fraction of fine martensitic and/or bainitic grains formed during LWB prior to the mechanical test. Due to the brittle nature of Fe-Si SI-IMCs formed at the periphery or within the SI-IMCs structures, the chance of deformation is lower in the Fe-rich matrix compared to the Cu-rich matrix of the braze. Therefore, deformation-induced dislocations within the Fe-rich matrix were neglected. By contrast, GNDs in the Cu-rich matrix are induced by both the solidification during the brazing process, due to stresses from thermal contraction, and by deformation. The high dislocation density observed within the Cu-rich matrix surrounding the Fe-rich regions and along the crack propagation path is mainly attributed to tensile testing. This increase in dislocation density in the Cu-rich matrix is associated with localized strain in areas experiencing high plastic deformation. Although dislocations in the Fe-rich matrix or SI-IMCs might not form during deformation, their presence during tensile loading can influence strain accommodation in the Cu-rich matrix. To assess the combined effect of the Fe-rich and Cu-rich matrices on failure, a unified dislocation

density measurement tool is necessary. Since the dislocation density measured using GND analysis in EBSD represents the length of dislocations per unit volume for each matrix, and the mapping areas may vary across different LWB conditions, the effect of mapping area must be eliminated. To achieve this, a unitless parameter called the *GND.Ratio* is defined. This allows for direct comparison of the Cu/Fe relationship within a single sample and across different samples, independent of braze area variations. To calculate the *GND.Ratio*, the midpoints (M_i) of each dislocation density range ($i = 1, \dots, n$) for each matrix are calculated using Equation 6.4. The number of the dislocation density ranges detected in the present work is $n = 10$. Fig. 6.14c shows the color-coded ranges obtained from the legend of the GND maps.

$$M_i = \frac{LB_i + UB_i}{2} \quad (6.4)$$

Where, LB_i and UB_i are the lower bound and upper bound, respectively, in an identified dislocation density range. To determine the total dislocation density for matrix “X” (GND_X) in each LWB condition, the M_i values and their corresponding volume fractions in each matrix are inserted into Equation 5 as follows:

$$GND_X = \text{Total dislocation density} = \sum_i (M_i \times VF_i) \quad (6.5)$$

By defining a linear relationship between one matrix ($X1$) and another ($X2$), as shown in Equation 6.6, a unitless parameter called *GND.Ratio* ($X1/X2$) is created. This ratio eliminates the influence of the non-uniform braze areas and serves as a reliable metric to assess the role of dislocation activity across various brazing conditions.

$$GND.Ratio_{(X1/X2)} = \frac{GND_{X1}}{GND_{X2}} \quad (6.6)$$

A strong correlation is observed between the *GND.Ratio* (Cu/Fe) and the maximum Hencky strain obtained from the DIC maps before the failure stage during tensile testing, as shown in Fig. 6.14d for all conditions. In LWB-1, the *GND.Ratio* (Cu/Fe) reaches 1.03, indicating a well-balanced dislocation density between the Cu-rich and Fe-rich matrices. In LWB-2, with a higher volume fraction of the Fe-rich matrix, the *GND.Ratio* (Cu/Fe) decreases to 0.67. This ratio

still allows the Cu-rich matrix to accommodate strain, as reflected in its corresponding maximum Hencky strain value.

In LWB-3 and LWB-4 with weaker $\{100\}$ texture alignment between SI-IMCs and Cu-rich matrix than LWB-1 and LWB-2, the $GND.Ratio_{(Cu/Fe)}$ further decreases to 0.33 and 0.53, respectively. The coincidence of the data points in the graphs of the $GND.Ratio_{(Cu/Fe)}$ and Hencky strain (see Fig. 6.14d) suggests that the GNDs generated in the Cu-rich matrix of LWB-3 and LWB-4 during tensile deformation were insufficient to overcome the pre-existing GNDs in the Fe-rich matrix. This results in reduced dislocation accommodation, manifested as lower maximum Hencky strain values. The critical role of the morphologies of SI-IMCs within the braze joints in influencing dislocation activity, strain accommodation, and ultimately mechanical performance becomes evident.

It is important to note that solely relying on the GNDs in the Cu-rich matrix does not accurately correlate with the maximum Hencky strain, due to the inconsistency of the trends observed from GND_{Cu} and maximum Hencky strain in Fig. 6.14d. This highlights the functionality of the $GND.Ratio_{(Cu/Fe)}$ as a robust parameter for regulating dislocation densities across different phases within the braze. The $GND.Ratio_{(Cu/Fe)}$ is directly proportional to the Cu-rich matrix's ability to accommodate strain during deformation; a higher $GND.Ratio_{(Cu/Fe)}$ enhances the load-bearing capacity of the braze joint. The maximum Hencky strain is the highest logarithmic strain the sample experiences before failure during the tensile test. In other words, as the braze joint increases dislocation activity in a way that raises the $GND.Ratio_{(Cu/Fe)}$ up to failure, delaying crack propagation, greater displacement is achieved, enhancing the joint's toughness.

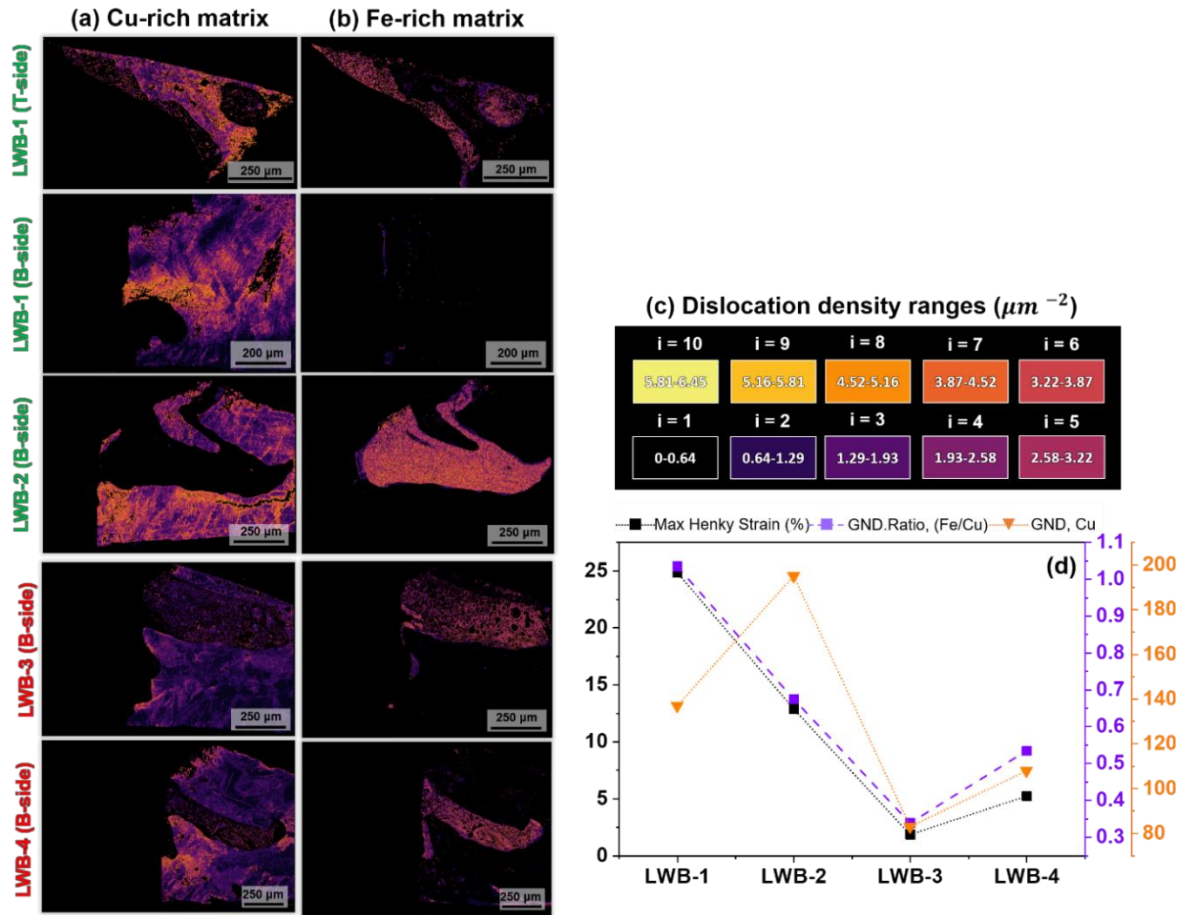


Fig. 6.14: GND maps of (a) Cu-rich and (b) Fe-rich matrices of the braze joints after failure for T-side LWB-1 and B-side LWB-1, LWB-2, LWB-3, and LWB-4 fracture samples. (c) Color coding of the dislocation density distributions in the GND maps. (f) Comparative graphs showing proportional relation of the maximum Hencky strain of the tensile test with the GND.Ratio (Cu/Fe) and the non-proportional relation with GND_{Cu} .

6.4.5 Crack Propagation Path Microstructure and Texture Analyses

6.4.5.1 Effect of SI-IMCs on Deformation Mechanisms

To understand how the $GND.Ratio_{(Cu/Fe)}$ controls strain accommodation in the braze joints during tensile testing, the deformation mechanisms during failure were investigated. The activated deformation mechanisms during tensile testing are directly related to the deformation texture observed at failure. To assess how the braze matrix and IMCs each contribute to the

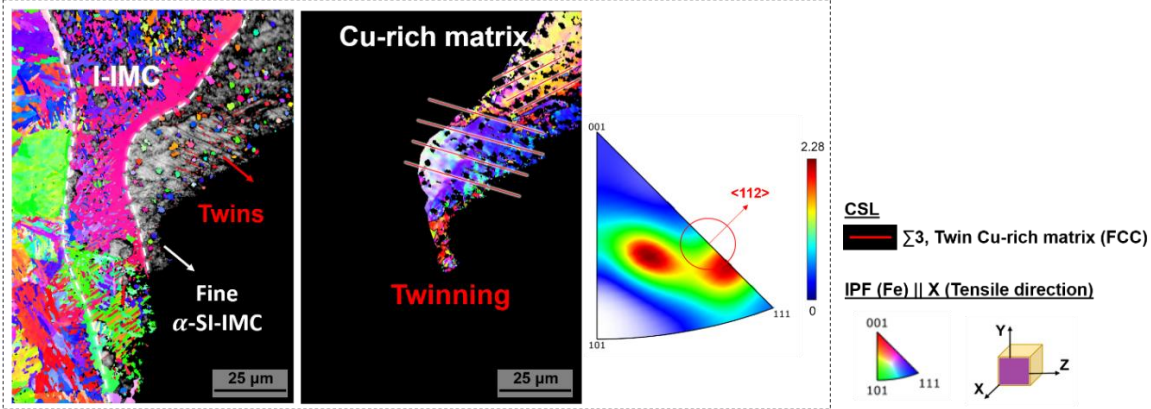
mechanical response, phase-level micro-texture and coincident site lattice (CSL) boundary analyses were conducted along the crack propagation path.

A high fraction of deformation twins, identified by $\Sigma 3$ boundaries, forms in the Cu-rich matrix along the crack path across all regions of LWB-1 and LWB-2, as displayed in Fig. 6.15 and Fig. 6.16. Twinning is most pronounced at two key locations: (1) intersections with fine α -SI-IMC that can be found in regions “R1” and “R2” in LWB-1, and “R1”, “R2”, and “R3” in LWB-2, and (2) intersections with the Fe-Si IMC shell of β -SI-IMC, particularly in “R2” in LWB-1, and “R2” and “R3” regions in LWB-2.

In LWB-1 and LWB-2, where twin boundaries form, the planes of the Cu-rich matrix predominantly align with the $\langle 112 \rangle$ direction, as evident from the IPF triangles relative to the tensile loading direction (X-axis) in Fig. 6.15 and Fig. 6.16. This orientation corresponds to the $\{111\}\langle 112 \rangle$ twinning system in FCC materials, confirming twin formation, as it is highly favorable for twinning [34,122,123]. The twin partitioning and localized plastic deformation are evident from the twin morphology observed in the IPF maps (Fig. 6.15 and Fig. 6.16). In fact, the twins formed in the braze matrix are promoted via a “SI-IMC-induced-twinning” mechanism. The intersection of the twin boundaries with the SI-IMCs in the Cu-rich matrix indicates that the adjacency of fine α -SI-IMC and β -SI-IMC impose constraints along the crack path, as these phases are non-deformable. This constraint alters the grain orientation of the surrounding Cu-rich matrix, promoting twinning.

(a) Region “R1”

LWB-1



(b) Region “R2”

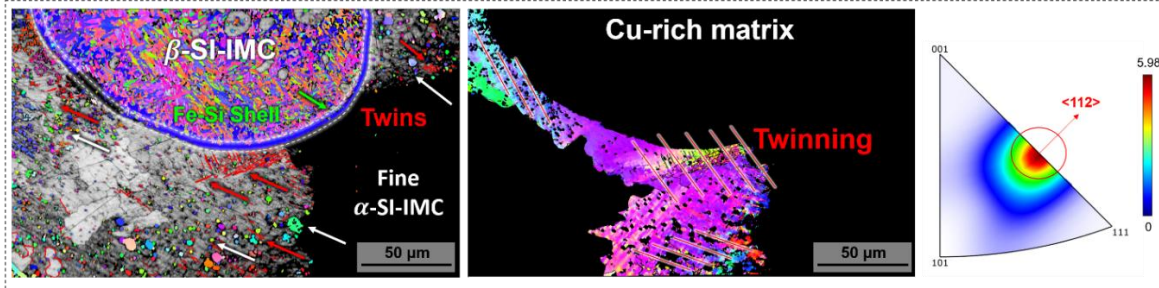


Fig. 6.15: Micro-texture analysis near the crack propagation path of LWB-1 showing SI-IMC (white arrows) induced twinning (red arrows). (a and b) The IPF map of the Fe-rich matrix combined with the band slope images of the Cu-rich matrix and the $\Sigma 3$ twin boundaries (red lines), corresponding to the Cu-rich matrix IPF maps and the Cu-rich matrix IPF triangles along the tensile direction (X-axis). The “R1” and “R2” regions are defined in Fig. 6.7-6.10.

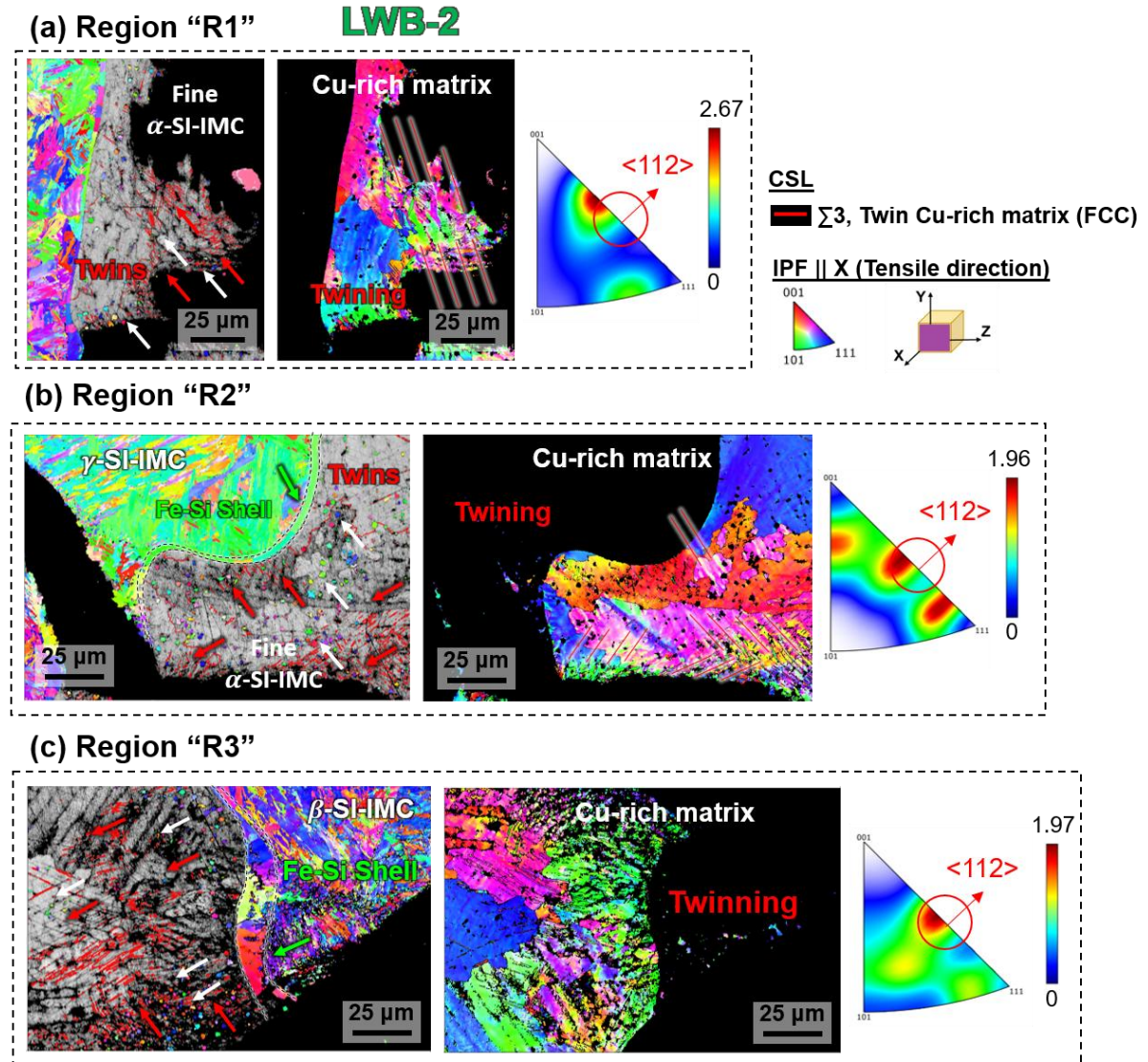


Fig. 6.16: Micro-texture analysis near the crack propagation path of LWB-2 showing SI-IMC (white arrows) induced twinning (red arrows). (a and b) The IPF map of the Fe-rich matrix combined with the band slope images of the Cu-rich matrix and the $\Sigma 3$ twin boundaries (red lines), corresponding to the Cu-rich matrix IPF maps and the Cu-rich matrix IPF triangles along the tensile direction (X-axis). The “R1”, “R3” and “R3” regions are defined in Fig. 6.7-6.10.

In contrast to LWB-1 and LWB-2, no significant twinning is observed in the Cu-rich matrix along the crack propagation path in LWB-3 and LWB-4, where SI-IMCs are present (Fig. 6.17). In LWB-3 (Fig. 6.17a), the Cu-rich matrix planes are oriented toward $\langle 101 \rangle$ and $\langle 111 \rangle$,

with no grains aligned towards the $\langle 112 \rangle$ direction. The δ -SI-IMC peninsula formed in LWB-3 was unable to transfer strain to the Cu-rich matrix, suppressing twin formation and limiting plastic deformation, which accelerated failure progression. In LWB-4 (Fig. 6.17b), no twin boundaries are detected within a region of Cu-rich matrix, where located between the I-IMC interface and a coarse α -SI-IMC that exists adjacent to the crack surface and closer to the crack initiation site. Additionally, the crystallographic planes of the Cu-rich matrix are predominantly oriented between the $\langle 101 \rangle$ and $\langle 111 \rangle$ directions rather than $\langle 112 \rangle$. Despite the high intensity of $\{111\}$ orientations, no twins are observed, suggesting that the dominant stress component in these grains favored dislocation slip rather than twinning. Although twinning in FCC metals follows the $\{111\}\langle 112 \rangle$ system, the presence of a strong $\{111\}$ texture alone does not necessarily indicate twinning activation. Twinning requires a critical shear stress component along the $\langle 112 \rangle$ direction on the $\{111\}$ plane.

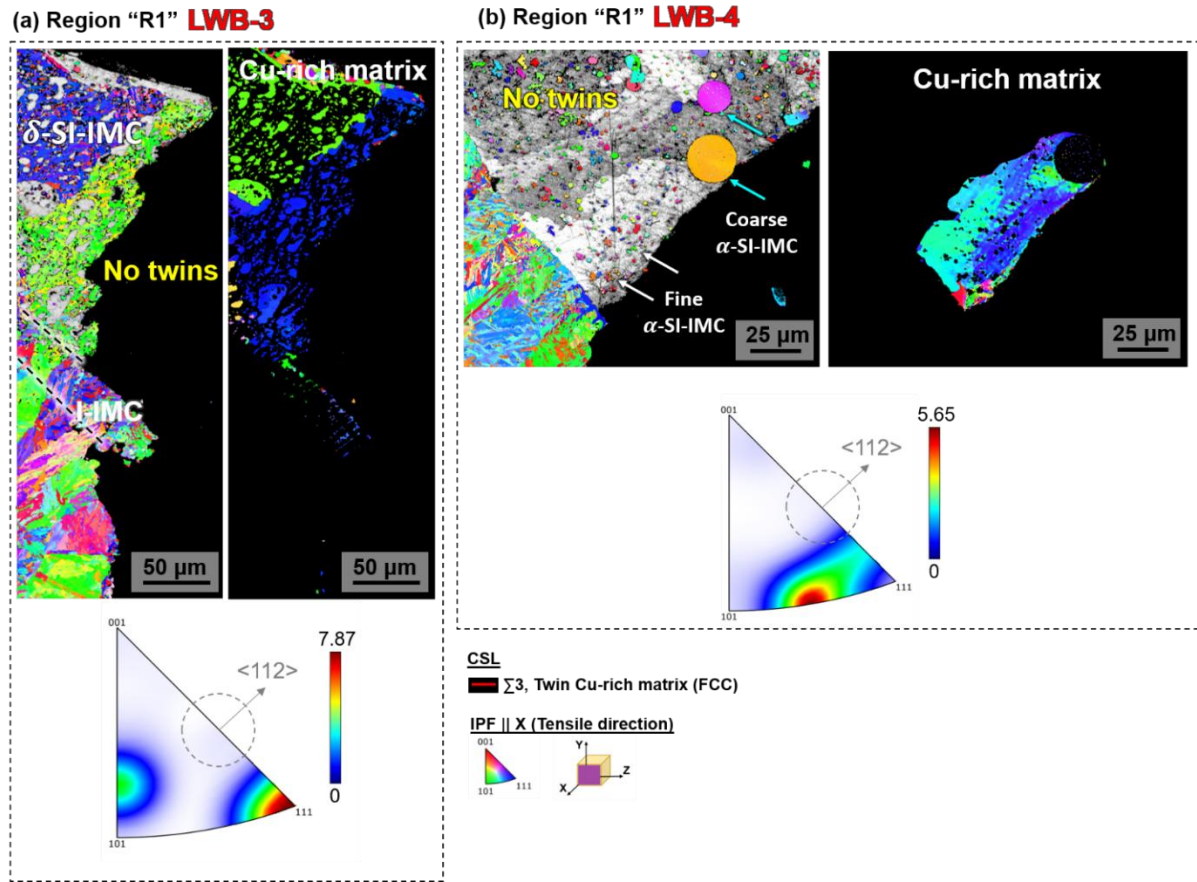


Fig. 6.17: Micro-texture analysis near the crack propagation path of LWB-3 and LWB-4. (a and b) the IPF map of the Fe-rich matrix combined with the band slope images of the Cu-rich matrix and the $\Sigma 3$ twin boundaries (red lines that are absent), corresponding to the Cu-rich matrix IPF maps and the Cu-rich matrix IPF triangles along the tensile direction (X-axis). The "R1", "R3" and "R3" regions are defined in Fig. 6.7-6.10.

Initiation of nano-twinning via precipitates and/or fine IMCs has been recently reported [121], showing improved plasticity through shear-band formation in the Cu-rich matrix of arc brazed joints under indentation testing. It is important to note that twinning behavior is highly dependent on the deformation mode. According to Schmid's law [122], twin-favorable grains differ between tension (during tensile testing in the present work) and compression (during indentation testing in [121]). The absence of twins in the regions containing fine α -SI-IMC in LWB-4 suggests that the mere presence of the IMC particles does not necessarily lead to the widespread twin formation in the Cu-rich matrix that was achieved in LWB-1 and LWB-2

during tensile loading. Twins observed in this study are micro-scale, providing a direct strengthening effect during tensile testing. In contrast, nano-twins under compression serve as a precursor to a secondary deformation mechanism, shear banding, which enhances plasticity. Although it is possible that nano-twins may form in LWB-4, their detection would require a high-resolution transmission electron microscope, which is planned for future work. Even if present, owing to the degraded mechanical performance of LWB-4, the probable nano-twins did not significantly contribute to strengthening under tensile loading.

The development or lack of twins in the joints in this work was striking. On one hand, there were twins present in the Cu-rich matrix of LWB-1 and LWB-2, and on the other, their absence in LWB-4, despite the existence of SI-IMC particles, indicates that the morphology, elemental distribution, and texture of the SI-IMCs strongly influence the driving force required for twinning when the joints are loaded in tension. It was found that the coarse α -SI-IMC particles do not generate sufficient driving force to promote twinning near their interfaces. In contrast, Fe-Si IMCs in the form of fine α -SI-IMC, β -SI-IMC, and γ -SI-IMC in LWB-1 and LWB-2 effectively constrained the Cu-rich matrix, enabling crystal plane rotation along the twinning direction, $\langle 112 \rangle$.

The role of twinning in enhancing strain hardening is well-documented, as it elevates tensile strength and ductility in FCC metals under deformation [240,241]. Twinning accommodates elevated strain by increasing the plasticity when dislocation accumulation is high, allowing further deformation through twin formation. Hence, in addition to the constraint effect of SI-IMCs with twin-favorable morphologies, interfacial bonding strength of the interface of SI-IMC shells with the Cu-rich matrix had to be higher than the interface of coarse α -SI-IMC and δ -SI-IMC to allow for imposing a higher strain to activate twins by reorienting the grains towards the twin-favorable directions.

6.4.5.2 Effect of SI-IMCs on Fracture Mechanisms

The deformation behavior of the braze joint is ultimately reflected in the fracture surface morphology, which may exhibit ductile and/or brittle fracture mechanisms. The SEM micrographs of LWB-1 and LWB-2 in Fig. 6.18 and LWB-3 and LWB-4 in Fig. 6.19 revealed that their fracture surfaces were significantly influenced by the presence or absence of SI-IMC particles. The IMC particles, fine α -SI-IMCs, were the only discernible evidence of the presence of SI-IMCs that could be observed within the fracture surface, due to their distinguishable fine spherical morphology.

In LWB-1 and LWB-2, ductile fracture features, such as dimples and crack blunting, are evident from Fig. 6.18 b, e, and f. The dimples, formed around the IMC particles, are referred to as “particle-induced dimples” and are driven by void nucleation, growth, and coalescence. The presence of SI-IMC particles in these regions promotes localized strain hardening by deviating the crack from its propagation path within the Cu-rich matrix to provide sufficient crack blunting, thereby delaying the final fracture. Notably, in the absence of the SI-IMC particles (Fig. 6.18c), the dimples are shallower, resulting from a reduction in capacity for plastic deformation. In contrast, LWB-3 and LWB-4 predominantly exhibit brittle fracture characteristics, such as cleavage and intergranular fracture, with minimal dimple formation (Fig. 6.19). In these conditions, the SI-IMC particles responsible for dimple formation are absent from the fracture surface. The observed cleavage and intergranular fracture modes are attributed to rapid crack propagation through the Cu-rich matrix of the network-shaped (Fe-Si)Cu phase in the δ -SI-IMC, which is not visible from the fracture surface. The transition from ductile failure in LWB-1 and LWB-2 to brittle failure in LWB-3 and LWB-4 correlates with the higher twinning activity near the crack propagation path in the former compared to the latter. Although SI-IMCs other than the fine α -SI-IMCs were not detectable within the fracture surface, their influence on strain accommodation are well understood. In LWB-1 and LWB-2, the presence of Fe-Si IMC shells of β -SI-IMC and γ -SI-IMC promoted twin formation, enhancing strain accommodation, and delaying fracture. In contrast, LWB-3 and

LWB-4, where twinning was absent or significantly reduced, showed limited strain accommodation, leading to premature failure.

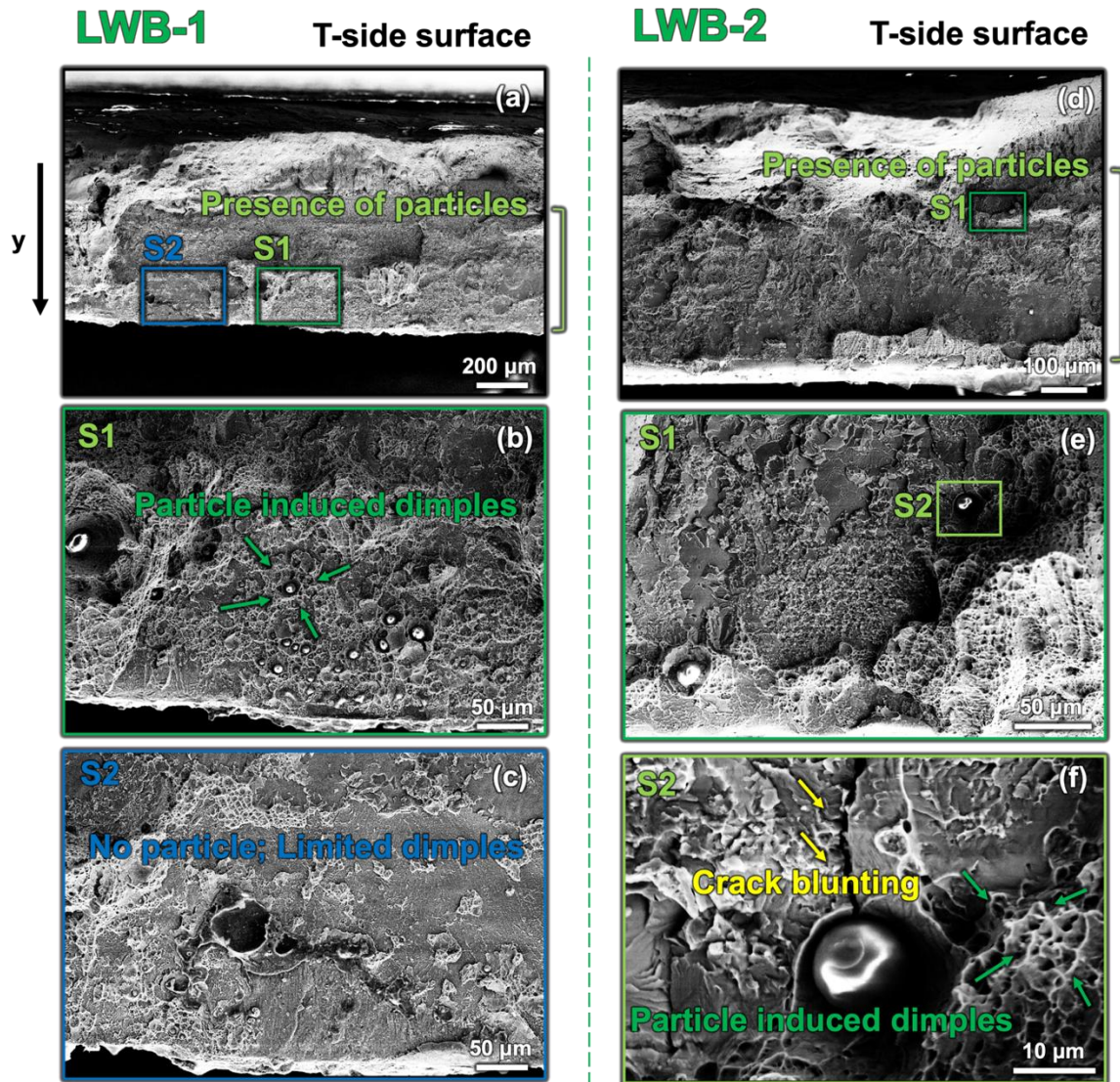


Fig. 6.18: Fractography analysis on T-side fracture surfaces of (a-c) LWB-1 and (d-f) LWB-2 joints after tensile testing, showing ductile fracture mechanisms.

This study offers a new perspective on the role of IMCs in fracture behavior, addressing a gap in the literature. While previous studies on the fracture mechanisms of brazed joints containing IMCs [21,104,116] primarily attributed failure to the mere presence of IMCs within the crack

propagation path or on the fracture surface, the present study offers a more nuanced understanding. This work demonstrates that the constraint effect of SI-IMCs, governed by their morphology, plays a crucial role in influencing deformation and fracture mechanisms. The findings highlight that in brazed joints, where the Cu-rich matrix is typically softer than the second-phase SI-IMCs, the morphology of SI-IMCs dictates the mechanical response by altering the local stress distribution and activating strengthening mechanisms, such as twinning and crack blunting. Notably, IMCs do not solely promote embrittlement; the SI-IMC structures, with favorable morphologies, texture, and distribution, can enhance strain accommodation and delay failure, thereby improving the joint's overall mechanical performance.

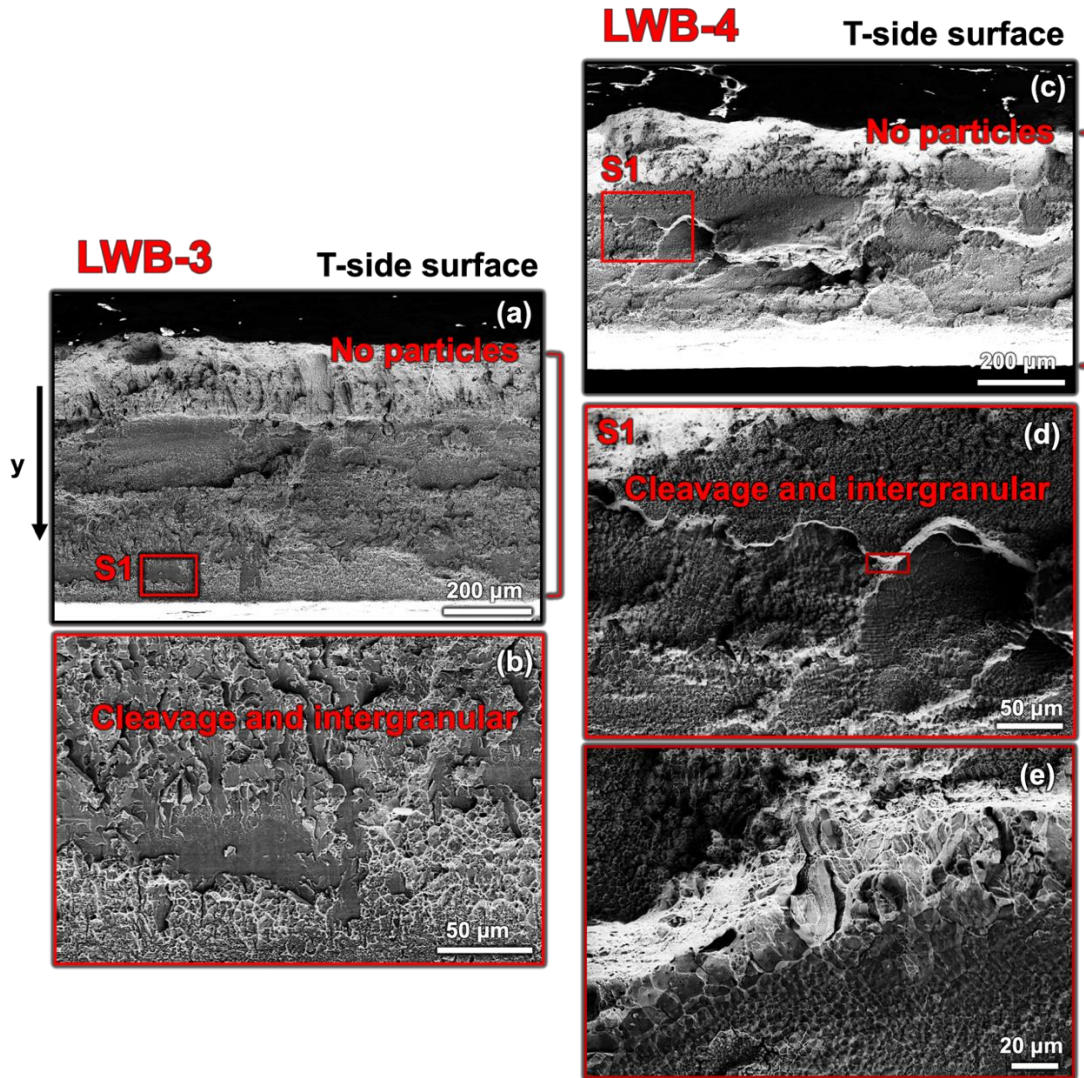


Fig. 6.19: Fractography analysis on T-side fracture surfaces of (a-b) LWB-3 and (c-e) LWB-4 joints after tensile testing, showing brittle fracture mechanisms.

6.5 Summary

In this study, the unknown effect of IMCs forming within the braze matrix of LWB joints, called SI-IMC structures, on the mechanical performance and fracture behavior was disentangled. This was achieved by optimizing the LWB process for Zn-coated Gen 3 Q&P980 steel substrates using a modified wire-adjusted HI approach for a lap joint configuration to

cause failure within the braze in the presence of various SI-IMC structures. Comprehensive examination of the fracture microstructure, elemental segregation, dislocation, and texture of the LWB joints after tensile testing, provided critical insights into optimizing IMC/braze-matrix interactions to enhance mechanical performance of the joints. The main conclusions could be drawn as follows:

1. The LWB process was successfully optimized to achieve braze joints with various SI-IMCs structures across four elevated HIs with $\Delta HI_{Relative}$ levels of 2.3 J/mm in LWB-1, 7.5 J/mm in LWB-2, 13 J/mm in LWB-3, and 17.5 J/mm in LWB-4 above the minimum required brazing HI in the experimental lap joint configuration (32.52 J/mm).
2. Mechanical testing under tensile loading conditions established a direct relationship between $\Delta HI_{Relative}$ and the joint mechanical performance. LWB-1 and LWB-2 resulted in higher mechanical performance (peak load $> 1308.18 \pm 181.82$ N, displacement $> 0.3118 \pm 0.041$ mm), whereas LWB-3 and LWB-4 degraded the mechanical performance (peak load $< 685.38 \pm 96.81$ N, displacement $< 0.0805 \pm 0.016$ mm).
3. Crack propagation was found to be significantly affected by the SI-IMC morphologies. Five distinct classes of SI-IMCs were identified near/along the crack propagation path, classified based on their elemental distribution and morphologies, affecting joint fracture behavior: fine α -SI-IMC, β -SI-IMC and γ -SI-IMC (delaying failure), as well as δ -SI-IMC and coarse α -SI-IMC (accelerating failure).
4. IMCs within the SI-IMC structures were identified by Fe-Si segregation with near-zero KAM. The α -SI-IMC consisted of Fe-Si IMC particles with fine size ($< 3 \mu\text{m}$) in all conditions except LWB-3 and their coarse size ($> 3 \mu\text{m}$) only appeared in LWB-4. The β -SI-IMC referred to a SI-IMC structure (that was found only in LWB-1) with a spherical Fe-rich island containing Cu-rich globules surrounded by Fe-Si IMC shell. The γ -SI-IMC was formed as an anchor-shaped Fe-rich peninsula surrounded by an Fe-Si IMC shell in LWB-2. The δ -SI-IMC was characterized by a large immiscible (Fe-Si)Cu (network-shape) island and peninsula in both LWB-3 and LWB-4.

5. Strong {100} texture alignment between SI-IMCs and the Cu-rich matrix enhanced interfacial bonding, effectively shifting strain localization to the softer Cu-rich phase. This texture alignment increased strain accommodation, thereby improving crack resistance and overall joint toughness.
6. A novel $GND.Ratio_{(Cu/Fe)}$ was introduced as a key parameter for controlling dislocation density across the different phases in the braze joint. A higher $GND.Ratio_{(Cu/Fe)}$ improved strain accommodation in the Cu-rich matrix, enhancing the joint's load-bearing capacity. As dislocation activity increased this ratio at failure, greater displacement was achieved, contributing to improved joint toughness.
7. SI-IMC-induced twinning in the Cu-rich matrix was identified as a critical mechanism for enhancing strain accommodation and strain hardening, thereby improving mechanical performance. Twinning was activated by the fine SI-IMC particles (α -SI-IMC) near the crack propagation path and was further facilitated by the constraint effect of the bulky SI-IMCs in LWB-1 and LWB-2, which rotated the Cu-rich grains toward $\langle 112 \rangle$, a favorable orientation for twinning. In contrast, bulky SI-IMCs in LWB-3 and LWB-4 lacked this capability.
8. The fracture mechanisms of the braze joints were primarily dictated by the SI-IMCs, influencing the transition between ductile and brittle failure modes. Ductile fracture was characterized by particle-induced dimple formation in twinned regions (LWB-1 and LWB-2), whereas cleavage and intergranular fracture dominated in the absence of twinning (LWB-3 and LWB-4).

Chapter 7. Conclusion and Future Recommendations

7.1 Conclusion

Sub-surface defects and IMC embrittlement undermine the integrity and mechanical performance of LWB joints. Detecting defects and producing reliable brazes with high strength remain major challenges, limiting wider adoption of LWB. This research advanced LWB process for Zn-coated automotive steels by developing novel LWB optimization strategies that facilitated in-line AI-enhanced NDT inspection, and by innovatively tailoring IMC formations maximized joint's mechanical performance. This research primarily developed fundamental understanding that shows laser processing, with its higher power density, modifies substrate surface morphology and segregation in a way that improves wetting and braze quality over arc processes. Furthermore, this study was able to predict the formation of features in the braze (i.e. IMC, pore, BMM, LoA) using a novel heat input metric that not only accounted for linear power delivered by the laser as well as accounting for heat extracted from the system to melt the filler wire. This metric was key to producing robust AI-friendly LWB joints with customized defect conditions required to train AI for in-line real-time UT monitoring that was unachievable before. This study offered new insights into IMC formation by introducing previously unreported SI-IMC structures, modulated by the novel heat input metric in this work, that revolutionized the braze microstructure. This research developed a new definitive understanding of IMC dispersion in joint mechanical performance by delineating the contributions of various SI-IMC structures in altering crack propagation paths, deformation mechanisms, and texture, critical factors that ultimately govern joint strength, displacement, toughness, and fracture behavior. These findings, for the first time, revealed the critical role of IMC/braze-matrix interactions during loading, and showed that their optimization significantly maximize joint mechanical performance. Altogether, this work makes both scientific and practical contributions toward the design of high-performance, AI-compatible brazed joints for next-generation multi-material vehicle structures. The following conclusions can be drawn from the present study:

1. LWB was found to produce a more uniform brazing bead with reduced spattering compared to GMAB. LWB had twice the joining speed and minimizing heat absorption in GI- and GA-coated DP600 steels compared to the arc process. In LWB, the HAZ dominated over the PMZ, and a 25% decrease in contact angle improved wettability through Zn accumulation at the bead edge. Unlike GMAB, where tail formation occurred only in GI-coated substrates, LWB enabled it in both GI- and GA-coated samples, with GI showing a 50% smaller tailing angle. Wetting differences between LWB and GMAB were primarily driven by laser-solid/liquid substrate interaction and heat conduction. GA-coated substrates, with higher roughness, smoothed under laser heating due to phase transformation, mitigating roughness effects. The smaller tail in GA-coated samples correlated with curve-shaped Fe-Si segregation near the bead edge, formed due to restricted molten wire flow and a higher temperature gradient.
2. A novel wire-adjusted HI approach was introduced to strategically optimize LWB process to set direct correlation between heat input, internal defects, and SI-IMC formation. This method incorporated wire deposition into the heat input calculations. The wire-adjusted heat input was represented in 3D by $C_{i=1,2,3,...,n}$: ($HI_{Nominal} = \frac{P}{TS}$, $HI_{Wire} = \frac{P}{WFS}$, $\Delta HI_{Relative} = (\frac{P}{TS+WFS})_i - (\frac{P}{TS+WFS})_{baseline}$), accurately enabled predictive navigation into brazing features. The $\Delta HI_{Relative}$ boundaries facilitated distinguishing the required HI for producing a defect-free braze seam, modulate SI-IMC fraction and morphologies, and identify conditions where internal defects occur. Using this optimization strategy, an AI-friendly process window was established for LWB of GI-IF steel with a trapezoidal geometry used in automotive roof joints, enabling the production of braze seams with specific internal defects, including LoA, pores, and BMM, while allowing classification based on defect size relative to the specified I.U.B. for each type.
3. Pore defects, ranging from below to above the I.U.B. (0.2 mm for pores), exhibited greater variation in $\Delta HI_{Relative}$ levels compared to LoA and BMM defects, requiring

- verification of their consistent distribution along the braze seams as characterized by MT. Full-scale μ CT analysis confirmed the repeatability of these pores in LWB conditions classified as pore condition. A comparison of maximum pore diameters in MT and the upper 95% in μ CT showed a minor deviation (+0.12 mm) for defect-free and pore-containing conditions, without affecting defect classification. However, even in defect-free conditions, pores below the I.U.B. (0.2 mm for pores) with sizes of 0.02–0.12 mm may still be present. These AI-friendly braze seams generated extensive datasets of internal defect occurrence in UT-readable brazed geometries and successfully enabled AI training for in real-time inspection.
4. In roof joint configuration of LWB for GI-IF steel, a $\Delta HI_{Relative}$ of 12.44 J/mm relative to a defect-free braze condition, was found to elevate the heat input at a controlled condition which was highly favorable to generate high density of SI-IMC with $38.2 \pm 16.9\%$ area fraction, while BMM kept below the I.U.B. (0.2 mm for BMM). The SI-IMCs formed composed of a bulky (Fe-rich)-Cu eutectic phase covered by shell-like Fe-Si layer with flat/dendritic morphologies and dispersed/clustered Fe-Si particles. SI-IMC formation mechanism was found to be similar to growth mechanism that occur during layer thickening in I-IMC type. SI-IMCs were found to be promoted under a controlled elevated heat input where TG and CR reached their peak levels within the braze, since the necessary Marangoni flow for an increased diffusion of Fe into the braze to form Fe-Si IMCs, without causing substantial melting of the base metal. This high density of SI-IMC formation significantly improved the tensile peak load (648.3 ± 14.1 N), displacement (4.3 ± 0.5 mm), and toughness (2.5 ± 0.1 J) were achieved by 14%, 350%, and 525% increment, respectively, when comparing to low SI-IMC density samples with $1.2 \pm 2.4\%$ area fraction.
 5. A wide verity of SI-IMC structures was achieved during LWB process for Zn-coated Gen 3 Q&P980 steel coupons across four elevated HI with $\Delta HI_{Relative}$ levels of 2.3 J/mm, 7.5 J/mm, 13 J/mm, and 17.5 J/mm of above 32.52 J/mm, the minimum required brazing HI in the experimented lap joint configuration. The mechanical test

under tensile loading condition established a direct relationship between $\Delta HI_{Relative}$ and joint mechanical performance. It was found that brazes formed within 2.3-7.5 J/mm significantly enhanced mechanical performance by 91.7% in peak load ($> 1308.18 \pm 181.82$ N), 319.84% in displacement ($> 0.3118 \pm 0.041$ mm), and 1200% in toughness ($> 0.26 \pm 0.06$ J) compared to brazes formed at higher $\Delta HI_{Relative}$. Brazed formed within 13-17.5 J/mm with degraded mechanical properties resulted in significantly lower peak load ($< 685.38 \pm 96.81$ N) and displacement ($< 0.0805 \pm 0.016$ mm). Fracture microstructure analysis from the tensile samples revealed that the differences in morphologies and chemical compositions of SI-IMC structures were responsible.

6. Five distinct classes of SI-IMCs were identified near/along the crack propagation, based on their elemental distribution and morphologies, affecting joint fracture behavior. It was found that fine α -SI-IMC, β -SI-IMC, γ -SI-IMC formed at 2.3–7.5 J/mm delayed the crack propagation (high mechanical performance), while δ -SI-IMC and coarse α -SI-IMC formed at 13-17.5 J/mm accelerated it (low mechanical performance). IMCs within SI-IMC structures identified by Fe-Si segregation had near-zero KAM. The α -SI-IMC consisted of Fe-Si IMC particles with fine size ($< 3 \mu\text{m}$) appeared at any $\Delta HI_{Relative}$ above 2.3 J/mm except at 13 J/mm. The coarse size α -SI-IMC ($> 3 \mu\text{m}$) only appeared at 17.5 J/mm. The β -SI-IMC formed at 2.3 J/mm with spherical Fe-rich island with Cu-rich globes surrounded by Fe-Si IMC shell. The γ -SI-IMC was formed at 7.5 J/mm with an anchor-shaped Fe-rich peninsula surrounded by an Fe-Si IMC shell. The δ -SI-IMC was characterized by a large immiscible (Fe-Si)Cu (network-shape) island and peninsula formed at above 13 J/mm.
7. The texture relationship between the Cu-rich matrix of the braze and SI-IMCs structures was discovered to significantly dictate their impact on mechanical performance. Strong {100} texture alignment between SI-IMCs and the Cu-rich matrix enhanced interfacial bonding, effectively shifting strain localization from SI-IMCs to the softer matrix (Cu-rich), which demonstrated an increased capacity to accommodate

strain. A novel *GND.Ratio* _(Cu/Fe) was defined to elucidate the contribution of texture alignment into strain accommodation. A higher *GND.Ratio* _(Cu/Fe) improved strain accommodation in the Cu-rich matrix, enhancing the joint's load-bearing capacity. As dislocation activity increased this ratio at failure, thereby improving crack resistance and greater displacement was achieved, contributing to improved joint toughness.

8. To enhance mechanical performance, the critical deformation mechanism that increase strain accommodation and strain hardening was SI-IMC-induced twinning in the Cu-rich matrix, which could only occur in presence of desirable SI-IMCs. The fine α -SI-IMC particles under the constraint effect of the fully-dense bulky β -SI-IMCs and γ -SI-IMCs led to rotation of the Cu-rich grains toward $\langle 112 \rangle$, a favorable orientation for twinning, inducing high plasticity delaying the failure. The coarse α -SI-IMC particles and δ -SI-IMC with network-shape bulky morphology, however, were found to lack this capacity provide the constraint effect to activate twin formation. SI-IMCs governed the transition between ductile and brittle failure modes, with ductile fracture exhibiting particle-induced dimples in twinned regions, while cleavage and intergranular fracture prevailed in the absence of twinning. The fracture mechanisms of the braze joints were primarily dictated by SI-IMCs, influencing the transition between ductile and brittle failure modes. Ductile fracture was characterized by particle-induced dimple formation in twinned regions, whereas cleavage and intergranular fracture dominated in the absence of twinning.

7.2 Areas of Future Research

While this study has addressed key gaps in the literature on LWB of Zn-coated steel joints, it also opened several new avenues for further investigations. Future work should focus on the following aspects.

This research established a predictive framework for anticipating sub-surface defects and modulating SI-IMC formation by developing the wire-adjusted HI approach. Future studies could further refine this approach by incorporating additional factors, such as the diameter and

specific heat capacity of the brazing filler wire, to measure the exact energy absorbed by the wire. Determining the exact heat consumed during LWB to melt the deposited filler wire and complete wetting, regardless of joint configuration (e.g., roof joint, lap joint), is essential for generalizing the established LWB optimization method across various manufacturing scenarios.

This study explicitly disentangled the previously unknown effects of IMCs dispersion on mechanical performance and fracture behavior. Future studies could explore the effect of heat input on phase transformations occurred to form fine α , coarse α , β , γ , δ SI-IMC structures using high-resolution techniques, such as EPMA and TEM analyses, to determine their precise chemical composition and crystal structures. Nano-segregation of elements beyond Si, such as Mn, is expected to be present, offering critical insights into the nucleation mechanisms of Fe-Si IMCs as well as the formation of other IMCs, including Cu-based phases.

Future studies could further delve into understanding the kinetics of SI-IMC formation in desired LWB conditions. This can be achieved by simulating LWB condition in Gleeble testing, to find the required time and temperature for development of a certain SI-IMCs. This experiment will provide a basic diffusion couple for braze/substrate system to investigate the elemental diffusion and phase transformation sequences using interrupted approach. Interrupted Gleeble tests will enable validation of thermodynamic models that can predict phase transformation based on phase diagram of responsible elements. Understanding SI-IMC formation mechanisms along with the knowledge of their thermal history will facilitate the strategic modulation of IMCs in various processing scenarios, enabling precise microstructural control for optimizing joint performance.

The SI-IMC-induced twinning characterized in this study represents a deformation mechanism that can be activated in the presence of desirable SI-IMCs. To gain deeper insight into this phenomenon, high-resolution TEM can analyze the interaction of twins with SI-IMC interfaces and particles, determining their crystal structure, orientation relationships, and atomic arrangement. This will clarify how SI-IMCs induce twinning in the Cu-rich matrix through

constraint effects. Moreover, to further investigate the local mechanical behavior of SI-IMCs, ball indentation testing can be conducted to measure hardness and yield strength of the bulk of SI-IMC structures as well as regions where twining is anticipated to occur during tensile test. Moreover, microhardness mapping across the entire braze area, interface, and HAZ in the base metal can help assess the hardness gradient in the joint and its correlation with ultimate tensile properties. Local mechanical properties assessment is particularly valuable, to directly characterize the differences in activated deformation mechanisms and strengthening capacity differences between tensile and indentation testing methods in the joints containing SI-IMCs.

This study investigated LWB of various Zn-coated steels for different study purposes. Future studies can explore the effect of base metal chemistry on SI-IMC formation mechanisms, as it significantly influences phase stability in Si-bronze brazes. The IF steel favors Fe-Cu solid solutions and Fe-Si IMCs due to its predominantly ferritic structure and low alloying elements. Although Fe-Si IMCs were characterized in the case of Gen 3 Q&P980 steel, due to its martensitic phases and higher Mn, Si, and C content, Fe-Si interactions and phase equilibria are expected to be affected. This is because, Mn modifies Fe solubility in Cu, while C diffusion may promote carbide formation, which may impact the Fe-rich phase distribution in the braze.

The AI-friendly process window successfully enabled acquiring reliable UT signals and AI-enhanced UT inspection of braze geometry, and the AI-based defect detection is an ongoing work. Beyond defect and geometry inspection, AI implementation in the LWB process can be extended to microstructural assessments of defect-free joints. The microstructure can identify the joint overall mechanical performance by correlating micromechanical properties (i.e., hardness, yield strength) with macromechanical performance (i.e., tensile peak load, displacement). Establishing a framework that links microstructural features, such as SI-IMC structures which appear darker than the braze-matrix and brighter than the voids in 2D slices of μ CT scans, to train AI to predict the overall mechanical performance of the joint. The full integration of AI for defect detection, geometry conformity assessment, and mechanical property prediction will enable real-time optimization of braze mechanical performance, achieving zero-defect manufacturing with elevated joint performance. If successfully

implemented, this could mark a significant breakthrough in manufacturing efficiency, sustainability, and safety, setting the stage for next-generation smart manufacturing systems.

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