

Effect of Paint Baking on the Fusion Boundary Softening and Fracture Behavior of Q&P 980 Steel Resistance Spot Welds

Dileep Chandran Ramachandran^{a, 1}, Olakunle Timothy Betiku^a, Mohammad Shojaee^a, Adib Salandari-Rabori^a, Abdelbaset R.H. Midawi^a, Ji-Ung Kim^a, Reza Bakhtiari^{a,b}, Andrew Macwan^c, Elliot Biro^a

^a Centre for Advanced Materials Joining (CAMJ), Department of Mechanical & Mechatronics Engineering, University of Waterloo, Waterloo, Ontario N2L 3G1, Canada.

^b Department of Materials and Textile Engineering, Faculty of Engineering, Razi University, Kermanshah, Iran

^c ArcelorMittal Global Research, 1390 Burlington Street East, Hamilton, ON L8N 3J5, Canada.

¹ Corresponding author: *Dileep Chandran Ramachandran*
Centre for Advanced Materials Joining (CAMJ),
Department of Mechanical & Mechatronics Engineering,
University of Waterloo, Waterloo, Ontario, Canada, N2L 3G1
E-mail: dcramachandran@uwaterloo.ca

Abstract

During automotive assembly, vehicles undergo low-temperature heat treatment (paint baking) to harden the paint. Although paint baking occurs at a relatively low temperature, it can remarkably affect the weld's mechanical and fracture behavior. This work studies how paint baking improving the strength and fracture behaviour of Q&P 980 spot welds exhibiting a halo ring; a low carbon enriched zone in the weld nugget. The mechanical behavior of the paint-baked welds reveals an increase in cross-tensile strength and absorbed energy when baked at 180°C for 27 minutes. Microstructural observation showed that the martensite present in the as-welded conditions started to decompose into tempered martensite with ϵ -carbide in the martensitic matrix. Gleeble thermo-mechanical simulations of the upper-critical heat-affected zone (UCHAZ) were produced to understand the mechanical and fracture micro-and macro-mechanisms, before and after the paint baking process, by widening the regions of UCHAZ. The transmission electron microscopic analysis of the Gleeble simulated sample reveals the segregation of C, Mn, Al, and Cr along the prior austenitic grain boundary which will change the nature of bonding at these boundaries. Nevertheless, paint baking treatment helps to redistribute the segregated elements from the grain boundary to the grain interior and to eliminate the solidified liquation formed at the grain boundaries during welding. The transformation of martensite to decomposed martensite, elimination of solidified liquation due to the enhanced atomic mobility and growth of surrounding grains, and the redistribution of C, Mn, Al, and Cr from the grain boundary to interior regions of grains, are the main reasons for the improvement of mechanical properties and fracture behavior of the spot welds.

Keywords: 3G-AHSS; Resistance Spot Welding; Paint Baking; Microstructure; Mechanical Properties

1 Introduction

Modern advanced high-strength steel (AHSS) for automotive applications possesses a high strength-to-weight ratio to achieve designs with high passenger safety and better fuel efficiency [1]. Quenched & partitioned (Q&P) third-generation AHSS (3G AHSS) has been suggested as a potential alternative to conventional high-strength steel due to the remarkable combination of strength and ductility [2,3]. These properties rely on the complex microstructures of 3G AHSS composed of ferrite, martensite, and retained austenite, developed by precisely designed alloying and thermo-mechanical processing steps.

Resistance spot welding (RSW) is the major manufacturing technique used in the assembly of AHSS sheets in body-in-white manufacturing [4]. Implementation of 3rd generation steels with such unique microstructure in automotive structures requires a systematic and detailed understanding of transformations that occurs during the RSW process, and their further effects on the failure behavior [5]. Due to the complex nature of the alloy design, unique challenges have arisen due to the increased hardenability and associated microstructures during the RSW process. The high heating and cooling rate inherent in the RSW process and the high alloying content of the 3G AHSS can lead to a martensitic microstructure in the fusion zone (FZ) and the upper critical heat-affected zone (UCHAZ). The brittle martensite structure in such areas of the weld can result in inferior mechanical properties of the spot welds, i.e., lower cross tension strength (CTS), and unfavorable failure modes [6,7]. Previous studies have shown that the presence of a transient softened zone (known as a halo ring [8,9]) along the fusion boundary, provides a low-energy fracture path following the fusion boundary. Hence, the CTS is found to be lower than similar sized welds in comparable materials that do not exhibit the halo ring [9–12]. In general, the mechanical properties of the spot welds during post-loading significantly depend on the microstructural features and how well these features contribute to accommodating the energy to increase the toughness.

Post-welded heat treatments have proven to be an efficient method to modify the microstructure, and subsequently improve the mechanical properties of the spot welds. Betiku *et al.* [13,14] showed that the application of a post-welded tempering pulse in spot welded press-hardened steel (PHS) can increase weld strength by 36%, compared to no post-heat treatment condition. The improvements in peak load were attributed to the transformation of martensitic lath structure into comparatively soft-tempered martensite structure with higher toughness. EftekhariMilani *et al.* [15] reported that the second current pulse partially anneals the solidified primary weld nugget by providing enough energy to redistribute the impurity element phosphorous, which changes the overall phosphorous distribution along the nugget boundary, improving the CTS performance. The authors' previous study [9] confirmed that the improvement of mechanical

properties of the double pulse welds (DPS) is related to the redistribution of alloying elements such as C, Mn, and Si in the fusion boundary, which could potentially eliminate the halo ring. Therefore, the manipulation of the microstructure via post-weld treatments is shown effective in enhancing the mechanical properties, while very limited works can be found dealing with such improvements in fracture behavior along with changes in the crack propagation path through the microstructure.

In car body manufacturing, paint baking (PB), as a low-temperature heat treatment, is a mandatory process that is carried out to harden the automotive paint. This process is also considered to be similar to a low-temperature tempering process [16]. EftekhariMilani *et al.* [6] reported that applying the baking cycle on the bainitic AHSS significantly improved the CTS and absorbed energy. However, they found no significant effect on the failure modes of the spot welds after cross-tension test (CTT). The improvements in the mechanical performance were attributed to the precipitation of ϵ -carbides in the martensite, as the precipitation of ϵ -carbide can improve the overall toughness, compared to the martensite [6]. Park *et al.* [17] investigated the effect of PB on spot welds made with medium Mn-TRIP and dual phase (DP) steels and showed that PB up to 210°C can improve the spot weld strength due to the nanoscale austenite reversion, softening the martensite. Furthermore, the segregation of Ti and Mo to prior-austenitic grain (PAG) boundaries that led to a reduction in boundary cohesion, which decreased after the PB. Shamsujjoha *et al.* [7] also observed an improvement in mechanical properties after the PB of 980 hot-formed (HF) steel. They theorized, evidence was not provided, that the improvement in CTS and toughness was due to the precipitation of carbides from the decomposition of martensite. Manladan *et al.* [18] reported that PB improved the energy absorption of the spot welds by changing the fracture path from the halo ring to the nugget area. Although PB improved mechanical properties, they failed to investigate the fracture characteristics and crack propagation mechanisms before and after the PB treatment. However, in all these studies the PB effect is mainly discussed in the context of CTS improvement with few studies dealing with toughness changes. Accordingly, they are falling short of comprehensively investigating the effects of bake hardening on the macro and micro-fracture behavior of the AHSS spot welds.

The authors' previous study showed that the mechanical properties of these resistance spot welds can be significantly improved by applying a double pulse welding schedule by implementing a relatively high current for a short time after the primary pulse [9,19]. However, there were no comprehensive studies to evaluate the macro and microfracture mechanisms before and after PB treatment of spot welds. Hence, this study focuses on the effect of PB on the microstructure, texture, and fracture behavior of the Zn-coated 3rd generation Q&P 980 steel resistance spot-welds. Furthermore, the authors explore the crack propagation mechanism in more detail by isolating the microstructure of UCHAZ, away from the surrounding

microstructure, by simulating the thermal history of this area of the HAZ using Gleeble thermo-mechanical simulation. By understanding these mechanisms, welding, and post-processing parameters may be modified in such a way as to eliminate the detrimental effects of the local microstructural features.

2 Experimental Procedure

2.1 Material

This study examines an industrially produced Zn-coated 1.2 mm thick Q&P 980 steel with a nominal chemical composition of 0.23 wt.% C, and other alloying elements, as shown in Table 1. The carbon equivalent of the material is 0.65 and was calculated according to the formula proposed by Yurioka *et al.* [20].

Table 1: Chemical composition (in wt.%) of the steel used in this study.

C	Mn	Si	Cr+Mo+Ti+Nb
0.23	2.01	1.05	0.22

The microstructure of the as-received steel consists of a combination of ferrite (F), martensite (M), tempered martensite (TM), and retained austenite (RA), as shown in Fig. 1. A detailed microstructural analysis was already reported in our previous work [9].

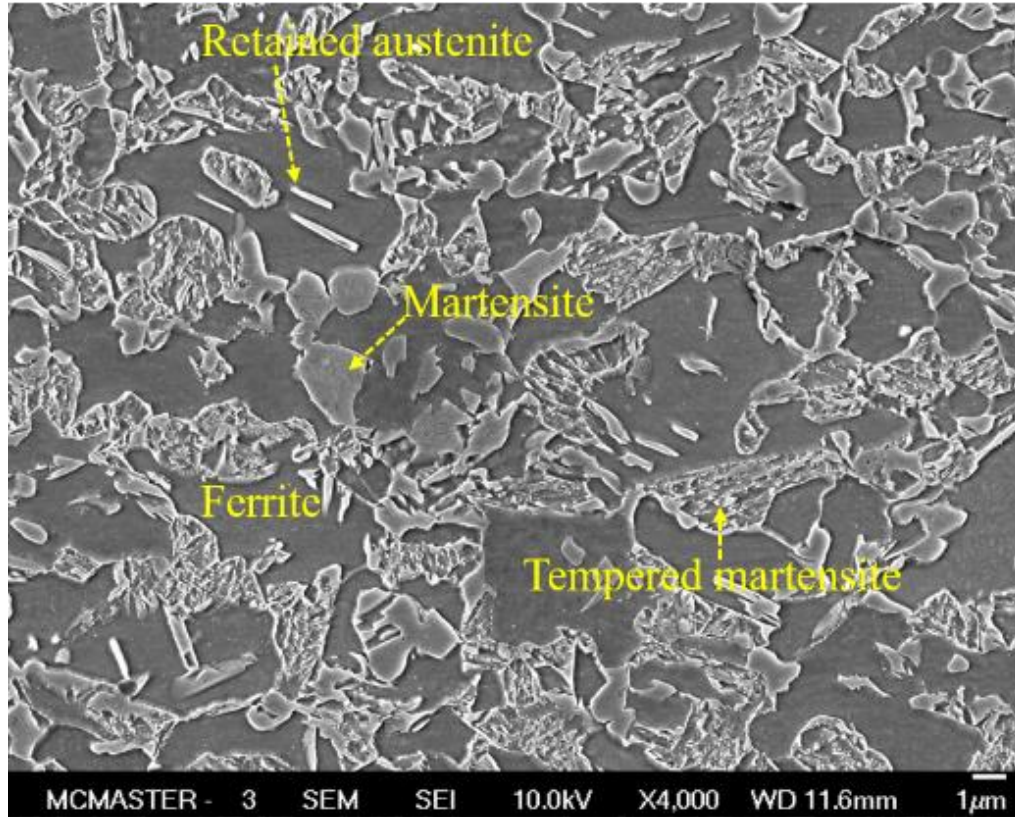


Fig. 1: SEM micrograph of as-received Q&P 980 steel.

2.2 Resistance Spot Welding

Resistance spot welding was carried out using a medium-frequency direct current (MFDC) robot with a C-type transformer and a Rexroth Bosh weld controller [21]. A dome-shaped Cu-Cr electrode with a flat face diameter of 6 mm was used for the welding. The welding schedules were chosen according to the AWS D8.9 specifications [22]. The detailed welding parameters used in this study were shown in Fig. 2. This study employed two different welding schedules: one is AWS-recommended single pulse welding (it will be named, AWS, hereafter) and the double pulse weld (DPS) as shown in Fig. 2. Optimization of these welding schedules and their effect on mechanical properties was already reported in our previous study [9]. An electrode force of 4 kN was used in all the experimented conditions. The average nugget diameter of both conditions was kept constant at 6.67 ± 0.11 mm.

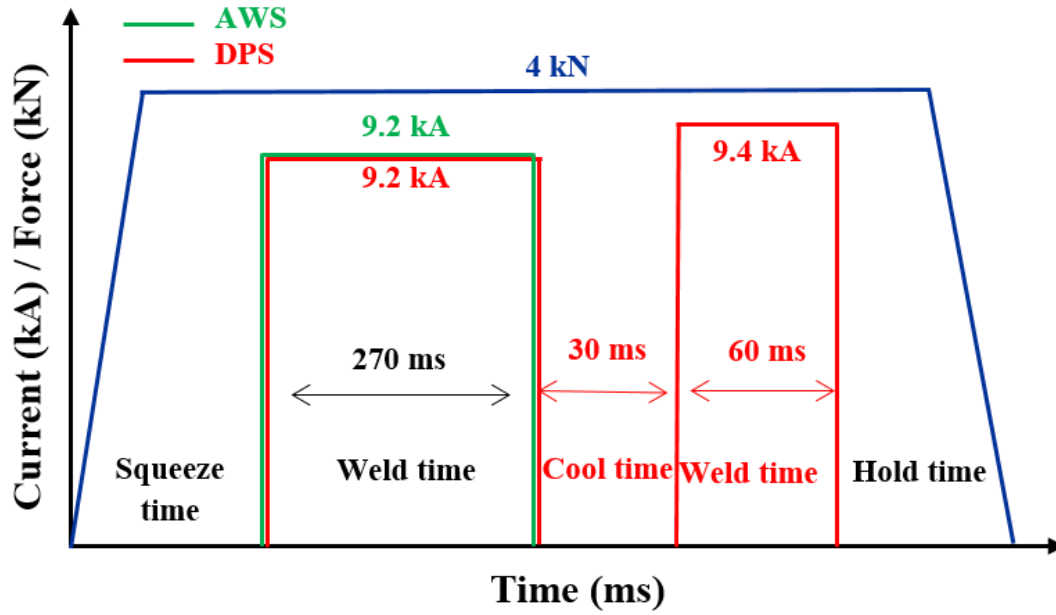


Fig. 2: The welding schedules (AWS and DPS) used in this study.

2.3 Thermo-mechanical Simulation

Rapid tempering heat treatments were carried out to understand the mechanical properties of the UCHAZ in the as-welded and paint-baked conditions (e.g., AWS and AWS-PB samples, respectively). The UCHAZ was simulated (G-AWS: Gleeble simulation of the sample welded by AWS-recommended single pulse condition) with a Gleeble thermo-mechanical simulator. The thermal cycle used for the heat treatment and the dimensions of the test coupon is illustrated in Fig. 3. To validate the ability of the heat treatment to reproduce the microstructure and mechanical properties of the UCHAZ in the weld, the hardness and grain size of the Gleeble simulated structure were assessed and matched to those of the weld (AWS sample). More details about the Gleeble simulations and discrepancies in the target and sample temperature were reported elsewhere [23]. The specimens were then machined into dog-bone-shaped samples for mechanical testing; the dimensions are shown in Fig. 3b.

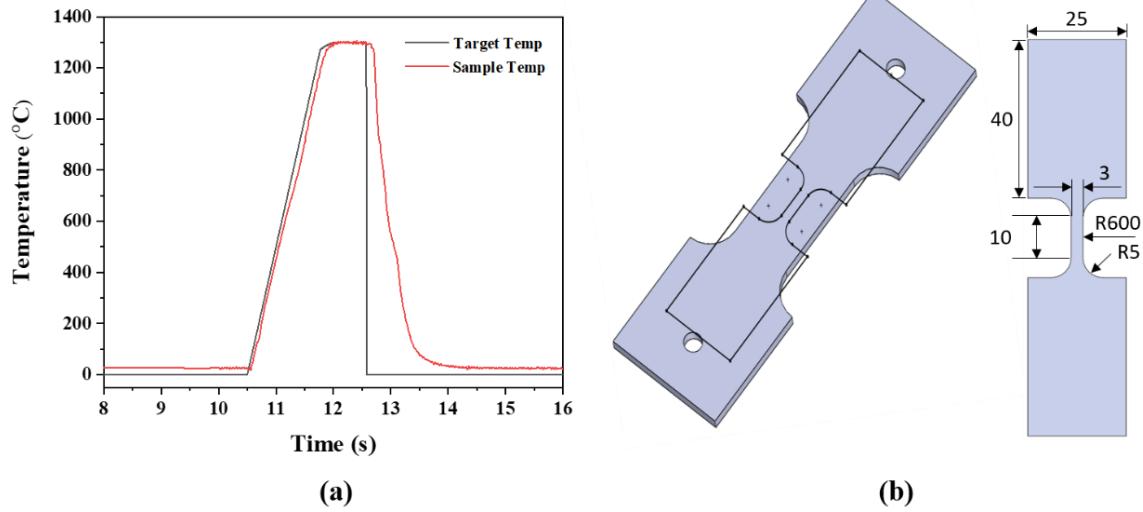


Fig. 3: (a) Gleeble simulation curve and (b) sample geometry before and after machining.

2.4 Paint baking

During automotive production, the car body will undergo a baking treatment in an oven to process the paint. To simulate those conditions in the laboratory, the welded samples were heat-treated in a muffle furnace at a temperature of 180°C for 27 min. Five samples were heat treated per welding condition to be used for mechanical testing and microstructural analysis. PB treatments were also carried out in the Gleeble samples (G-AWS PB) with the same heat treatment condition, as mentioned above.

2.5 Cross-Tension Test & Uniaxial Tensile Test

The mechanical performance of the joints was evaluated by doing CTT according to the AWS D8.9 standard [22]. Standard CTT coupons were used for the welding; the dimensions are reported elsewhere [1,19]. All mechanical tests were performed at a constant crosshead speed of 10 mm.mm⁻¹ using a 100 kN MTS frame universal tensile testing machine. The absorbed energy was calculated by integrating the area under the curve up to the peak load. All reported results are an average of three repetitions and the error bars show a 95% confidence interval (CI). To understand the fracture propagation behavior, interrupted CTT was also conducted (samples were loaded to ~90 % and ~75 % of the average peak load) for one sample per condition.

Additionally, uniaxial tensile tests (UTT) were conducted on the Gleeble samples to understand the mechanical properties of the samples after PB. The tests were conducted in a Shimadzu universal tensile

frame with a strain rate of $1 \times 10^{-3} \text{ s}^{-1}$ ($\sim 0.54 \text{ mm.min}^{-1}$) and elongation was measured with coupled digital image correlation (DIC) technique using a 10 mm virtual gauge length.

2.6 Microstructural Analysis

For metallographic analysis, the samples were prepared by conventional metallographic techniques up to 1 μm diamond suspension polishing followed by etching with 2% Nital solution for 10 s. Optical microscopy (Olympus BX51M) and scanning electron microscopy (SEM) were used to characterize the microstructure. A Zeiss UltraPlus FE-SEM with EDX was used to characterize the chemical composition. A JEOL JSM 7000F field emission SEM was used for the electron backscattered diffraction (EBSD) technique. An Aztec crystal software was used for the EBSD data post-processing. Standard metallographic techniques were used to prepare thin foils of transmission electron microscopy (TEM) samples followed by electro-polishing. The 3 mm discs were thinned using a TenuPol-5 twin jet electro-polishing equipment in a solution of 10% Perchloric acid and 90% Methanol maintained at -40°C , at a voltage of 15 V. A Talos 200 TEM at an accelerating voltage of 200 keV were used to analyze the samples. The mapping setup varies with the thickness of the sample and the spot size of the scanning-transmission electron microscopy (STEM) beam, as the intensity of the X-ray depends on the thickness and the beam current. Hence, the dwell time was maximized to obtain more accurate chemical distribution maps.

2.7 Microhardness

Hardness measurements were performed on unetched samples using an automated Clemex microhardness machine interfaced with a Vickers indenter. Hardness line scans were performed using a 0.2 kgf load and a dwell time of 10 s. The spacing between the consecutive indents was kept constant at 200 μm . The high-density hardness scans were performed with a load of 0.01 kgf with an average spacing of 100 μm . The error bars of the microhardness measurements represent the 95% confidence interval.

3 Results

The Q&P steel coupons were initially welded according to the AWS D8.9 recommendations [24], which had a single pulse current. In the authors' previous work [9], a double-pulse welding schedule was designed to improve the mechanical properties. Furthermore, the welds were tested using the CTT, and the resulting failure was observed in the cross-sectional view. It was shown that the AWS sample failed along the fusion boundary, resulting in a low-strength failure. The fusion boundary failure path, the halo ring, had a lower hardness. In what follows, the effect of paint baking on the mechanical properties, fracture behavior, and microstructural evolution will be presented.

3.1 Mechanical Properties

The CTT data and its representative load-displacement curves for AWS, DPS, AWS-PB, and DPS-PB samples are shown in Fig. 4a. The absorbed energy (area under the curve up to peak load) is representative of the crashworthiness of the weld, and is shown in Fig. 4b.

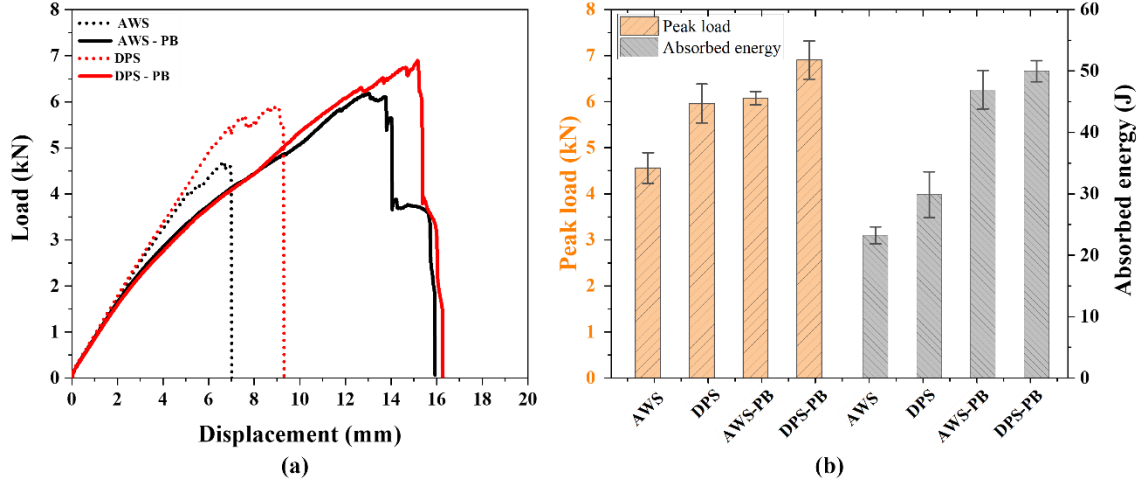


Fig. 4: (a) Load-displacement curves of CTT test and (b) absorbed energy up to peak load of all experimented conditions.

As shown, a notable trend can be seen where a tangible increase in peak load and absorbed energy can be realized by applying PB on both AWS and DPS samples. The average mechanical properties (CTS data) and its absorbed energy up to the peak load are summarized in Table 2. The highest CTS performance and absorbed energy are achieved by the DPS-PB sample with an average value of 6.9 ± 0.4 kN and 50.0 ± 1.7 J, respectively. So, it can be clearly understood that the AWS-PB sample shows more significant improvements in properties than the AWS sample. The increase in strength and absorbed energy after PB is 34% and 102%, respectively. Moreover, there is a significant difference in the average strength (6.1 to 6.9 kN) between AWS PB and DPS PB samples (see Table 2).

Table 2: Cross-tension test data and failure mode of the spot-welds.

Samples	Peak load (kN)	Absorbed energy (J)	Failure mode (AWS D8.1) [25]
AWS	4.6 ± 0.3	23.2 ± 1.4	Mode 3
DPS	5.9 ± 0.4	29.9 ± 3.7	Mode 1
AWS-PB	6.1 ± 0.2	46.9 ± 3.1	Mode 1
DPS-PB	6.9 ± 0.4	50.0 ± 1.7	Mode 1

Representative failure areas and their corresponding cross-sectional micrographs of different welds are shown in Fig. 5. A distinct change in failure modes between the as-welded (AWS and DPS samples) and as-welded and then paint-baked (AWS-PB and DPS-PB) samples can be observed after the CTT. AWS sample (welded according to AWS D8.1 [25]) shows partial thickness failure mode (known as Mode 3); in which the failure happens along the fusion boundary (see Fig. 5a). However, DPS, AWS-PB, and DPS-PB samples all show pull-out fractures through the UCHAZ (see Fig. 5b-d). Recently, Manladan *et al.* [18] reported a change in the failure mode from Mode 3 to Mode 2 after the PB treatment of 30MnB5 spot weld. However, no studies are reporting the changes in the failure mode from Mode 3 to Mode 1 due to the PB treatment. According to the literature, the pull-out failure mode exhibits the most satisfactory failure mode in the spot welds compared to the interfacial failure mode since it is associated with higher plastic deformation and absorbed energy before the fracture [7]. Thus, the application of PB after welding significantly eliminated the strength and toughness difference between the AWS and DPS conditions since both conditions (AWS-PB and DPS-PB samples) failed in a button pull-out mode, where the fracture path progressed through the UCHAZ. On the other hand, the PB treatment does not change the failure mode of the DPS sample after CTT. Furthermore, the DPS schedule eliminates the microstructural inhomogeneity of the fusion boundary by reheating the fusion boundary with the short second current pulse. This second current pulse homogenizes the microstructure near the fusion boundary, which mitigates the effect of the halo ring. A detailed study on the effect of DPS on microstructure and mechanical properties is reported elsewhere [9].

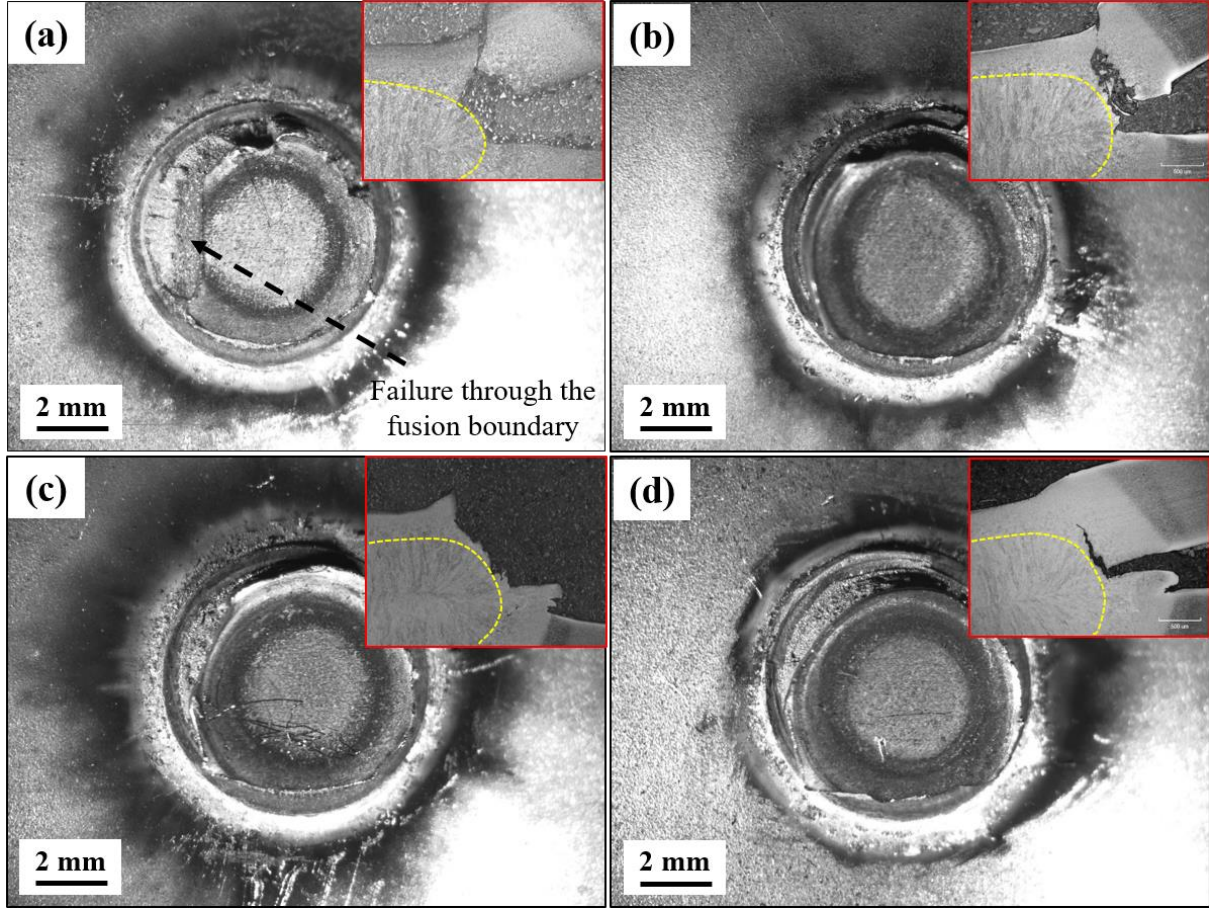


Fig. 5: Representative failure modes and their corresponding cross-sectional micrographs of samples (a) AWS (Mode 3), (b) AWS-PB (Mode 1), (c) DPS (Mode 1), and (d) DPS-PB (Mode 1).

Often in a simplified form, changes in mechanical properties (e.g., yield strength) can be measured as changes in hardness. From the microhardness profile of the studied weld samples, the average nugget microhardness of the AWS and AWS-PB samples are 541 ± 5.5 and 506 ± 4.0 $HV_{0.1}$, respectively. The average nugget hardness of the DPS and DPS-PB is reduced compared to the AWS and AWS-PB conditions and are 518 ± 5.0 , and 495 ± 5.5 $HV_{0.1}$, respectively (see Fig. 6). The average microhardness of the heat-affected zone (HAZ) for AWS and AWS-PB samples are 532 ± 11.0 and 530 ± 7.5 $HV_{0.1}$, respectively, and are 544 ± 8.5 and 516 ± 9.5 $HV_{0.1}$ for DPS and DPS-PB samples, respectively. The error bars of the microhardness measurements represent the 95% confidence interval. From the hardness profile, it may be seen that the average hardness decreases after PB. A decrease in hardness is typically associated with a decrease in material strength. So, it is believed that the change in hardness resulting from PB is due to the microstructural changes which can thereby lead to a different joint response during tensile testing, as well. Detailed discussions on the material's tensile strength will be addressed in the later sections.

Here, it should be noted that the PB also dramatically mitigates the halo ring, formed in the AWS sample. After PB, the average hardness of the region representing the halo ring increased from 417 to 447 HV_{0.1}. Moreover, the difference in hardness between the halo ring and surrounding microstructure (nugget and HAZ) decreased significantly, which reduces the possibility of stress concentration in the fusion boundary. For the DPS sample, however, the halo ring is eradicated and further exposure of samples to the PB treatment (DPS-PB) reduces the hardness differences at the fusion boundary to a higher extent (Fig. 6b).

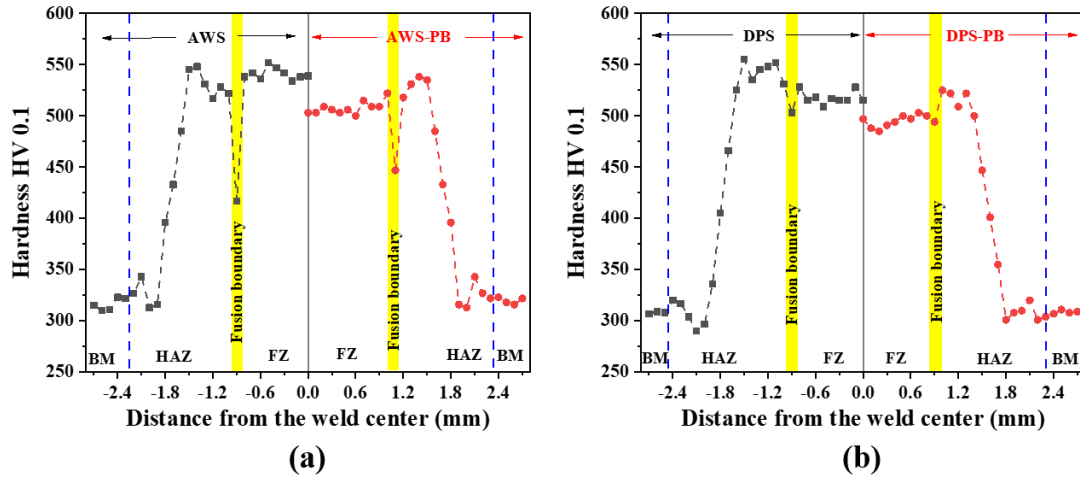


Fig. 6: Hardness profile of the samples (a) AWS and AWS-PB samples, (b) DPS and DPS-PB samples (where halo rings are not formed).

To further substantiate the results of CTT and microhardness tests, the UCHAZ, where the failure happens in the spot-welded samples, was simulated using Gleeble. Then, uniaxial tensile tests were performed on the Gleeble samples, namely, G-AWS and G-AWS PB (performing PB on the simulated UCHAZ Gleeble sample). Engineering stress-strain curves are shown in Fig. 7. As can be seen, after PB the ultimate tensile strength (UTS) and the elongation to fracture are noticeably improved by 7.6 % and 833%, respectively. Furthermore, the area under the stress-strain curve, up to the fracture, represents the toughness which is dramatically larger in the case of the PB. The G-AWS sample failed at around 1.1 % strain, whereas the G-AWS PB sample can accommodate the strain up to 11%. The absorbed energy of G-AWS and G-AWS PB samples is 9.4, and 151.0 J.m⁻³, respectively. It is also clear from the DIC images that the G-AWS sample experienced a brittle behavior as compared to the G-AWS PB sample, which shows a necking behavior implying a higher amount of energy absorbed during the tensile loading.

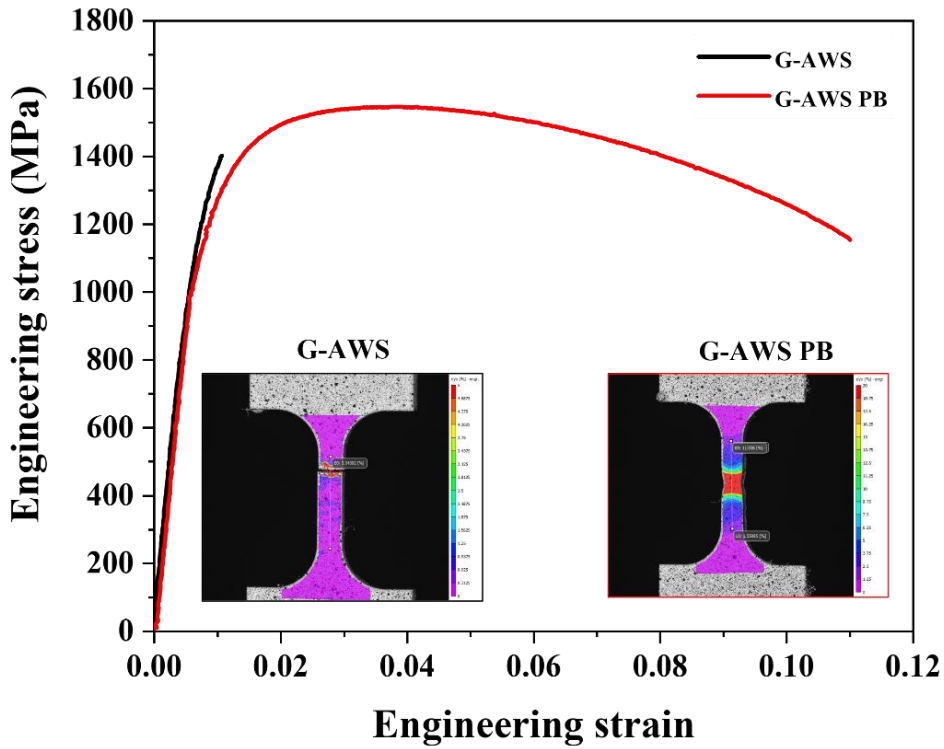


Fig. 7: Tensile engineering stress-strain curves of G-AWS and G-AWS PB samples.

Here, it is worth mentioning that the yield strength (YS) is slightly lower for the G-AWS PB sample, compared to the G-AWS sample (1161 vs. 1278 MPa) which agrees with the hardness measurements, as well. Tensile mechanical properties are shown in Table 3.

Table 3: Mechanical properties of the G-AWS and G-AWS PB samples after uniaxial tensile test.

Samples	YS (MPa)	UTS (MPa)	Absorbed energy (J.m^{-3})
G-AWS	1278 ± 2.8	1414 ± 81.4	9.4
G-AWS PB	1161 ± 19.6	1521 ± 20.0	151.0

To understand the reason behind the fracture, fracture surface analysis on the actual spot welds and the Gleeble samples are thoroughly done, and the results are summarised below.

3.2 Fracture Behavior

Interpretation of the failure locations in the spot welds is important to understand the underlying fracture mechanisms. Optical micrographs (OM) of the weld cross-sections (showing the hardness indents) for the AWS and AWS-PB samples are shown in Fig. 8a and b, respectively.

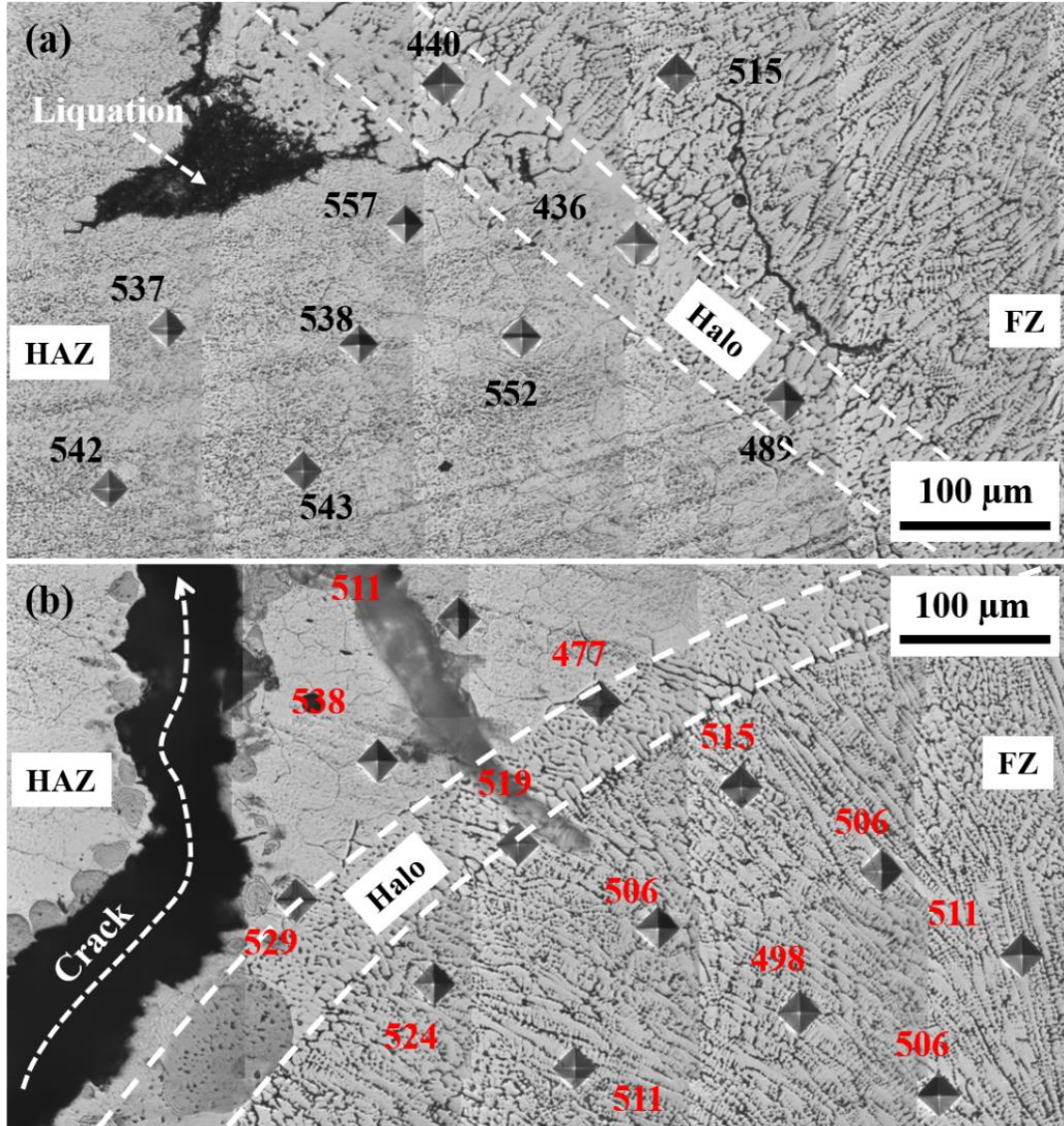


Fig. 8: OM cross-section of the fracture surface and detailed hardness values of (a) AWS and (b) AWS-PB samples (all the hardness measurements are in HV_{0.2}).

It can be seen that the fracture occurs through the fusion boundary (halo) and propagates through the solidified liquation region in the AWS sample (Fig. 8a). Nonetheless, surprisingly, the crack path deviates to UCHAZ in the sample subjected to the PB treatment, and solidified liquation is not seen in that cross-section (see Fig. 8b). Furthermore, the decrease in hardness along the fusion boundary, compared to the FZ and HAZ is clear in Fig. 8a. It is also worth noting that existence of a high hardness region around the halo (higher hardness of FZ and UCHAZ/HAZ than the halo), makes a sharp hardness gradient across the fusion boundary of the spot weld, which promotes a low energy failure path or stress concentration region during the mechanical testing. However, after PB, this gradient in the hardness from FZ to the HAZ is mitigated

as shown in Fig. 8b. In other words, due to a sharper reduction in the hardness of the HAZ side of the fusion boundary, an increase in the hardness at the fusion boundary and a slight reduction in the hardness of the FZ (compared to the AWS sample), the sharp hardness gradient significantly reduced (Fig. 8b).

The fracture surfaces of the AWS and AWS-PB samples are shown in Fig. 9. As shown in Fig. 9a, the AWS sample fractured in a brittle manner, and showed evidence of intergranular fracture (Fig. 9c) near the notch, and cleavage fracture (Fig. 9b) throughout the other areas. However, the fracture surface of the AWS-PB sample shows an overall ductile mode of fracture (Fig. 9d). Although the fracture initiated near the fusion boundary (close to the sheet-to-sheet interface) was like the AWS sample, the crack grew predominantly in a ductile manner (Fig. 9e). Here, a clear change in the fracture macro-mechanism can be seen from an intergranular to the trans-granular propagation of the crack. This further led to a change from a brittle to a ductile fracture, as shown by dimples on the fracture surface of the AWS-PB sample (Fig. 9e). Moreover, it can be noted that this observation is in good agreement with the reported mechanical properties since the change from cleavage to dimple fracture provides significant improvements in strength and energy absorption capacity, as presented in section 3.1.

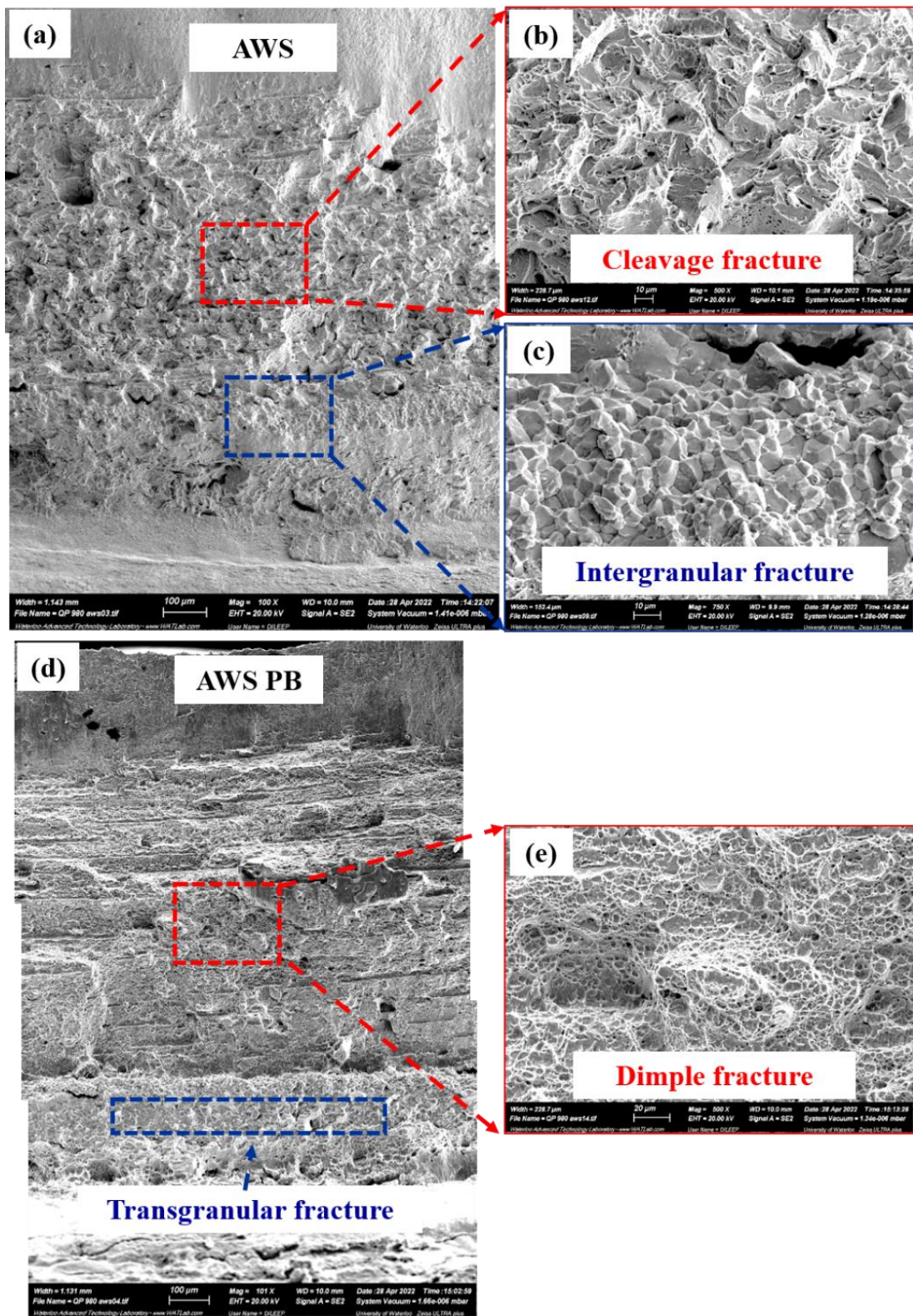


Fig. 9: SEM micrographs showing fracture surface of (a) AWS sample showing (b) cleavage and (c) intergranular fractures and (d) AWS-PB sample with (e) ductile fracture.

The scanning electron micrographs of the fractured sample (cross-sectional view) for the AWS-PB sample after the CTT are shown in Fig. 10. The plastic deformation occurs in the microstructure near the crack propagation area (Fig. 10a). Close observation depicts that the microstructure present in that region is the partially disintegrated martensite (Fig. 10a). This partially disintegrated martensite can absorb comparatively more energy during deformation; thereby increasing the energy required to propagate the crack. A typical example of partially disintegrated martensite is also shown in Fig. 10b.

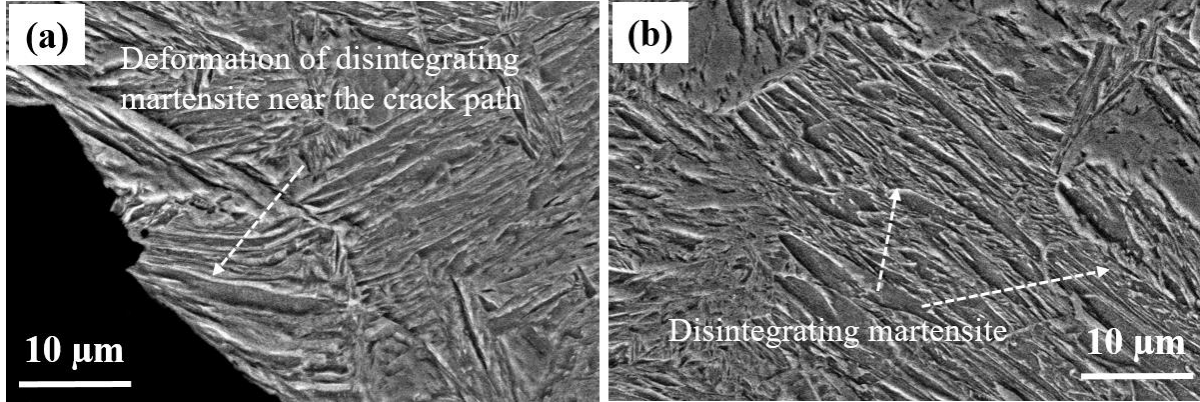


Fig. 10: SEM micrograph showing the (a) deformation of disintegrating martensite near the crack path, and (b) typical disintegrating martensite, in the AWS-PB sample after CTT.

To understand the difference in fracture behavior of the UCHAZ region in AWS and AWS-PB samples (where the failure happens) simulated via Gleeble (G-AWS and G-AWS PB samples), fractographic analyses were conducted after the tensile tests. The fracture surfaces of both G-AWS and G-AWS PB are compared in Fig. 11. As seen in Fig. 11a, in G-AWS samples also the crack propagation proceeds along the PAGBs with an intergranular type of fracture (in agreement with the AWS sample, see Fig. 9c). Moreover, limited trans-granular cleavage fracture characteristics can be also seen in Fig. 11a. It could be due to the presence of different types of grain boundaries with different energies. On the other hand, the specimens after the PB (G-AWS PB) show relatively a ductile fracture behavior including dimples (in line with observations in Fig. 9e). More importantly, Fig. 11b shows the presence of liquid film along the grain boundaries, pointed by arrows.

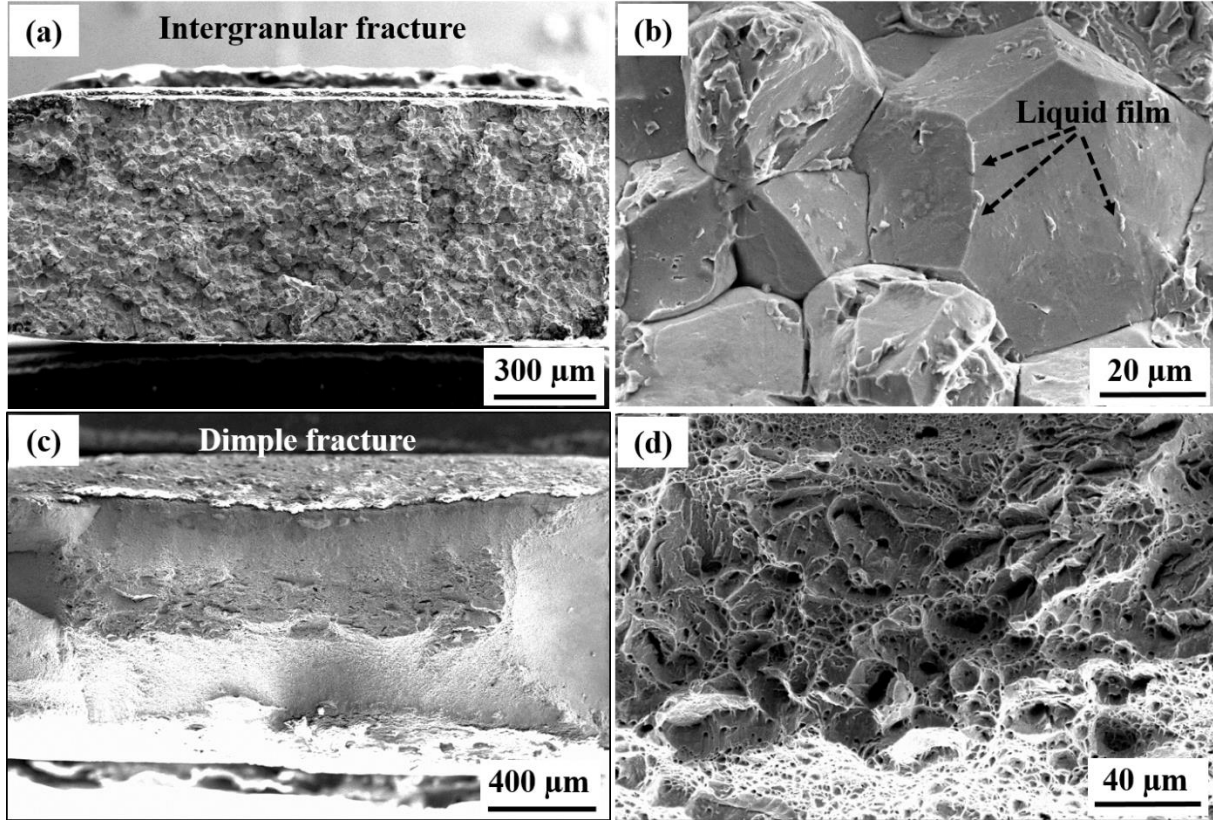


Fig. 11: SEM fractography micrographs of the tensile samples (a and b) G-AWS showing Intergranular-brittle fracture, (c and d) G-AWS PB showing dimple fracture.

Fig. 12 shows the difference in fracture behaviour from the crack beneath the surface (secondary crack analysis). Fig. 12a shows more evidence of intergranular fracture in the G-AWS sample. The PAGBs are marked in yellow color. On the contrary, the G-AWS PB sample shows a trans-granular failure with deformed regions on the microstructures (see Fig. 12b). These observations agree with Fig. 11. EBSD IQ and IPF maps (see Fig. 12c and d) on the G-AWS sample show a typical secondary crack propagating through the PAGB (an intergranular mechanism). It can be summarized that the resistance spot weld and its corresponding Gleeble samples are in acceptable agreement in terms of mechanical properties and fracture characteristics. Overall, the G-AWS sample shows brittle fracture behaviour where the crack is propagating intergranular manner, whereas a transition to ductile behavior is seen that is governing the fracture surface of the G-AWS PB sample with a trans-granular crack propagation mechanism.

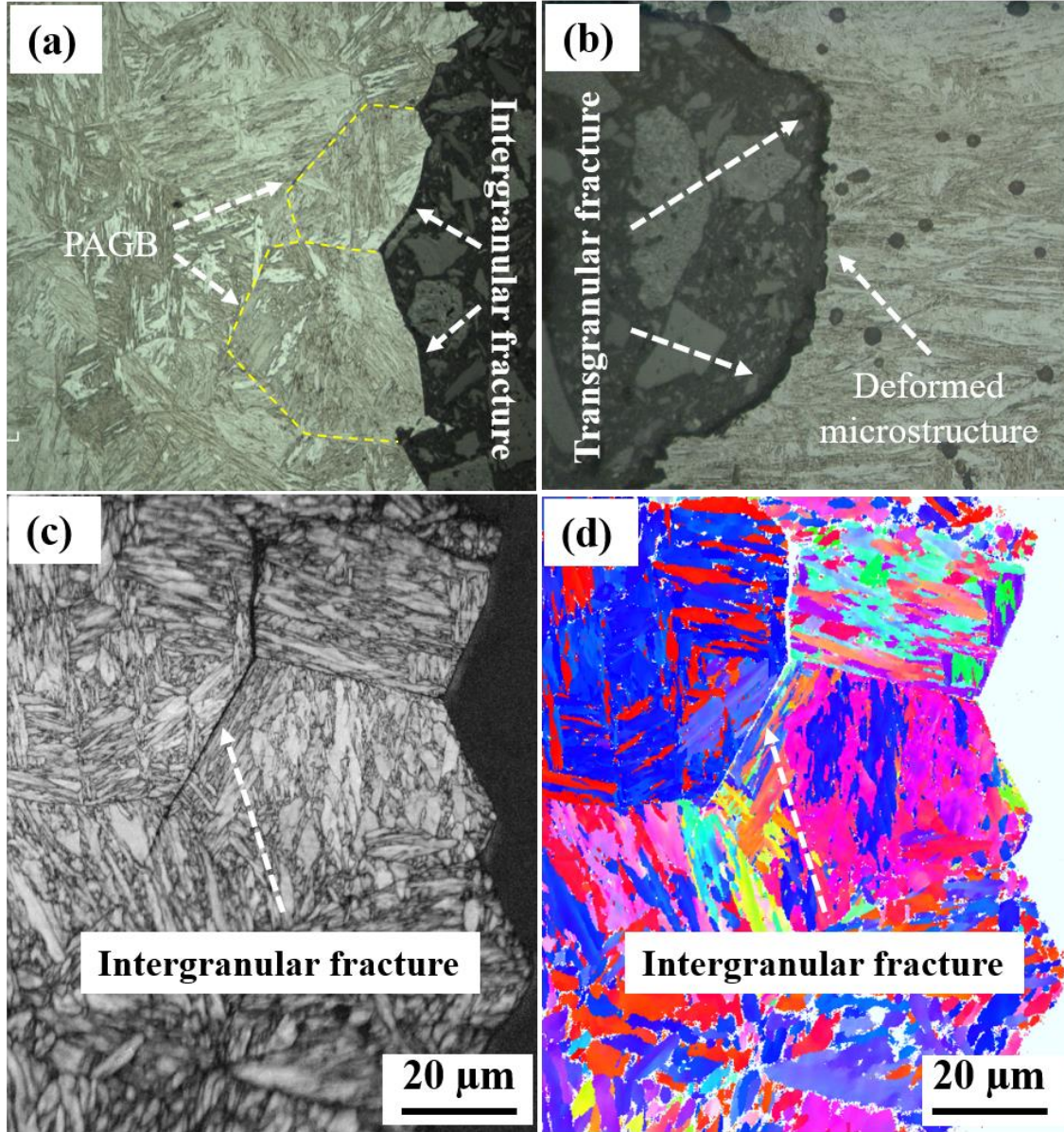


Fig. 12: Secondary crack analysis of the samples showing (a and b) OM of the crack beneath the fracture surface of G-AWS and G-AWS PB, respectively, and (c and d) EBSD IQ and IPF map showing the intergranular fracture in G-AWS sample.

3.3 Microstructural Inhomogeneities

So far, the mechanical and micro- (fractography) and macro-fracture (crack propagation) behavior of welded samples were assessed before and after PB. In the following, the differences in fracture behavior will be discussed in terms of microstructural evolution. The OM & SEM microstructures of the AWS sample, which failed in an intergranular fashion, are shown in Fig. 13. As seen in Fig. 13a, there is solidified

liquation product seen in this weld. The presence of liquation in the spot-welded Q&P 980 steels has been reported before [26]. This phenomenon is attributed to the segregation of Mn, Si, and P, at grain boundaries [27]. Alizadeh-Sh and Marashi [28] observed the liquation during the dissimilar RSW of austenitic/duplex stainless steel. They proposed a grain boundary sweeping mechanism for grain boundary liquation. This mechanism proposes that upon heating, grain boundaries sweep the solute and/or impurity atoms as the grain growth proceeds. Elements that have high surface activity and/or exhibit low solubility in the matrix would have the highest affinity for being swept into and assimilated with grain boundaries. Consequently, during the HAZ grain growth certain solute atoms, Al in their case, assimilated with moving boundaries. Afterward, these solute atoms acted as melting point depressant elements and preferential melting occurred along the grain boundaries during the HAZ thermal cycle [28].

The same grain boundary liquation mechanism is active here during the RSW. Such liquation can be seen near the fracture path of the AWS sample, as shown in Fig. 13b. Evidence of small cracks near the fracture path can also be seen in Fig. 13c, where the crack propagates through grain boundaries. As shown in Fig. 13d, the EDS elemental analysis across the fracture shows the segregation of Mn and Si, with an average composition of 4.6 and 2.3 wt.%, respectively. Saha *et al.* [29] pointed out that the liquation is formed in alloys of Al, Mg, and Fe, where a comparatively larger window of solidus and liquidus temperatures exist; therefore these alloys will exhibit a wider solidification range. While solidifying, the liquid film that had a comparatively lower solidus temperature, will remain in the liquid state along the grain boundaries (liquation). This will lead to the deterioration of grain boundary bonding. It should be worth noting that, the liquation present in the AWS sample solidifies between the grain boundaries, providing a crack-free bond between the grains. This solidified liquation product can further initiate cracking during post-weld loading since the liquation regions are low-strength regions, susceptible to stress concentration [26]. According to Li u *et al.* [26] and Jing *et al.* [30], the formation of liquation cracks was due to phosphorous segregation to grain boundaries during the welding. However, in the current study, segregation of phosphorous in this region was not observed, only Mn and Si. Still, cracks formed during CTT originating from the solidified liquation product, as may be seen in the AWS sample (Fig. 12).

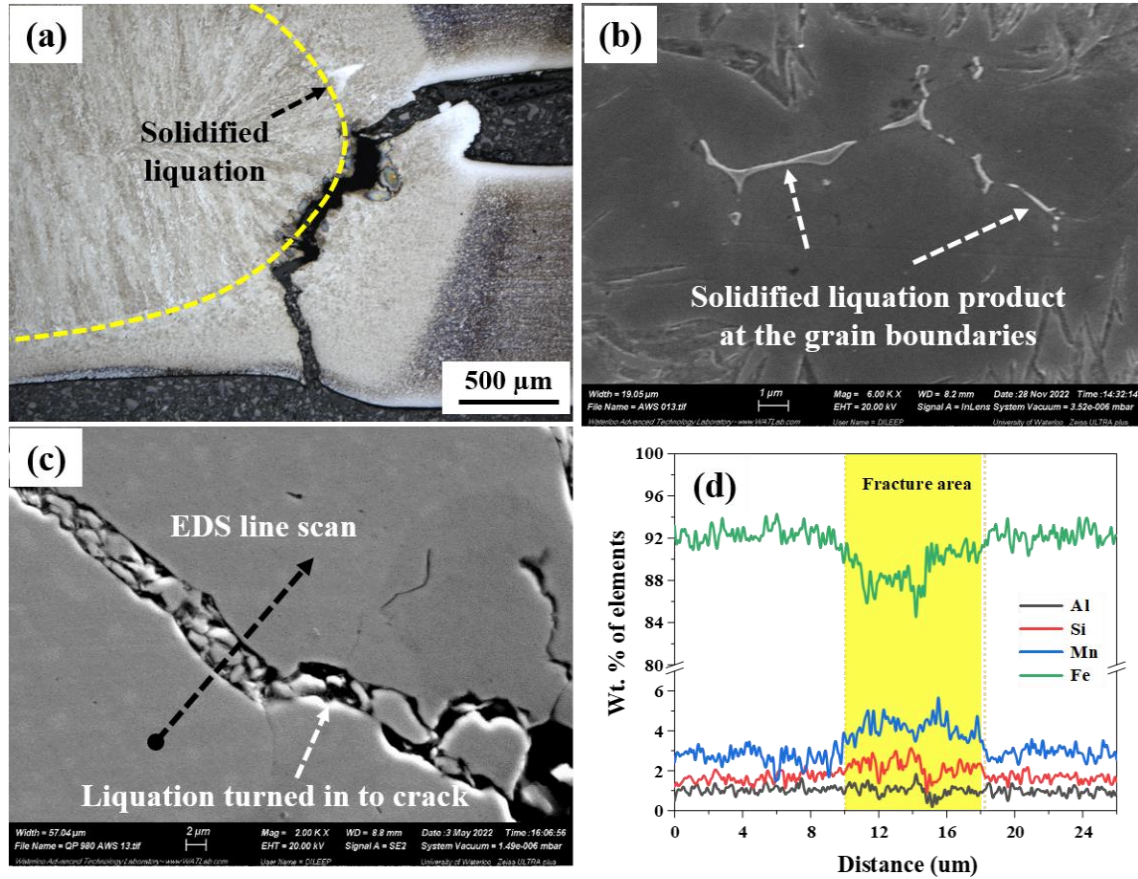


Fig. 13: Microstructure and elemental analysis of the solidified liquation product present in the AWS sample after the CTT: (a) OM micrograph showing the fracture path and solidified liquation product, (b) SEM micrograph of a typical solidified liquation product at grain boundary after fracture, and (c and d) representative SEM image and the EDS line scan across a solidified liquation, which transformed into a crack during the CTT.

Detailed EBSD analysis of the solidified liquation in the AWS sample (after welding) can be seen in Fig. 14. Fig. 14a shows the image quality (IQ) map of the weld with FZ, halo ring, and HAZ with a typical solidified liquation product. The width of the solidified liquation near the fusion boundary is approximately 90 μm . The inverse pole figure (IPF) map shows the presence of smaller grains compared to the surrounding microstructure, in the solidified liquation region (see Fig. 14b). The phase map shows the presence of austenite grains at the edge of this solidified liquation region (Fig. 14c). Combining the EDS line scans, and the presence of austenite, it can be interpreted that the austenite grains were stabilized to room temperature due to the presence of high Mn content at the tip of the solidified liquation, which greatly improves the stability of austenite at the lower temperature. As mentioned above, the presence of Mn, and

Si along the grain boundaries can pin the movement of grain boundaries, which could be the reason for the presence of smaller grains in the solidified liquation region. During the application of the load, the solidified liquation product can also initiate some micro-cracks, since the lattice mismatch due to the crystallography between the austenite and martensite grain causes stress concentration in its common boundaries [31,32]. It is worth noting that the solidified liquation was not observed in the AWS-PB sample (solidified liquation exists after welding but will be disappeared by PB treatment). It is believed that PB treatment enhances atomic mobility and grain growth. Heating at 180°C for 27 minutes provides significant time for the solidified liquation structures to diffuse out of the grain boundaries into the grain interiors.

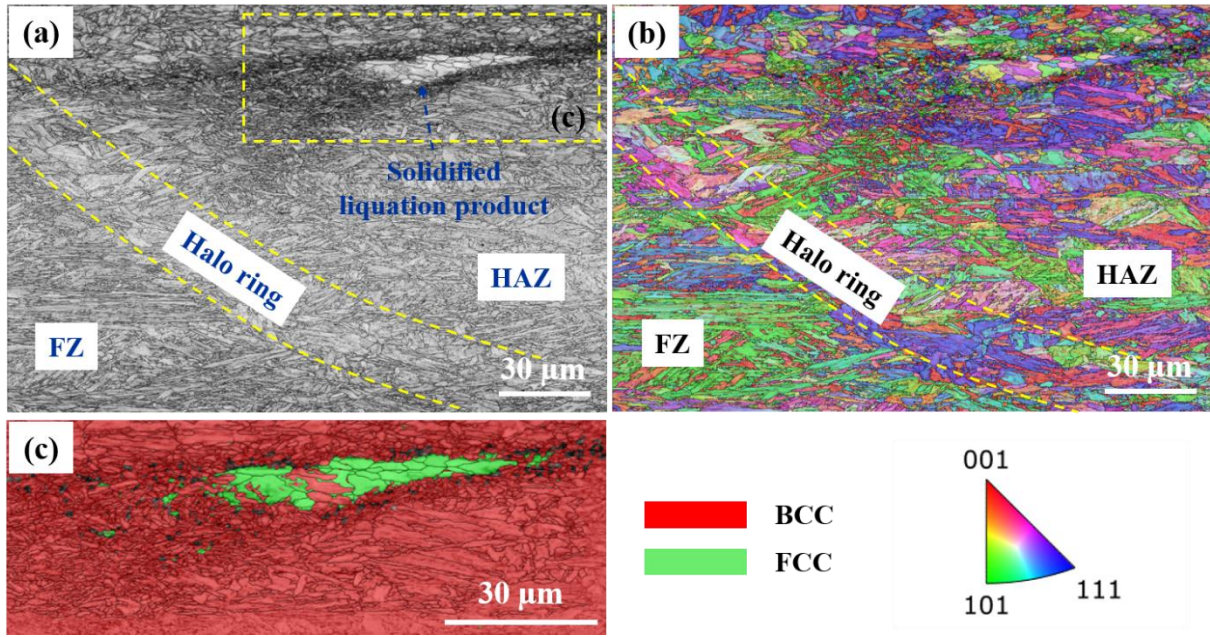


Fig. 14: Characterization of solidified liquation product in AWS condition (a) Image quality map, (b) IPF map, (c) phase map showing austenite in the solidified liquation.

In other words, Mn, and Si trapped at the grain boundaries can redistribute into the grains by diffusion during heat treatment. Moreover, the solidified liquation also can be consumed by the surrounding grains during the grain growth. These two metallurgical phenomena can eliminate the intergranular planes of weakness (liquation) in the AWS-PB samples. Since, these weak liquation regions, can act as crack initiation sources during post-weld loading, their elimination by the PB process will noticeably improve energy absorption and fracture behaviour.

Detailed microstructural analysis of the G-AWS and G-AWS PB samples are shown in Fig. 15. The IQ and corresponding IPF maps of the G-AWS sample are described in Fig. 15a and b.

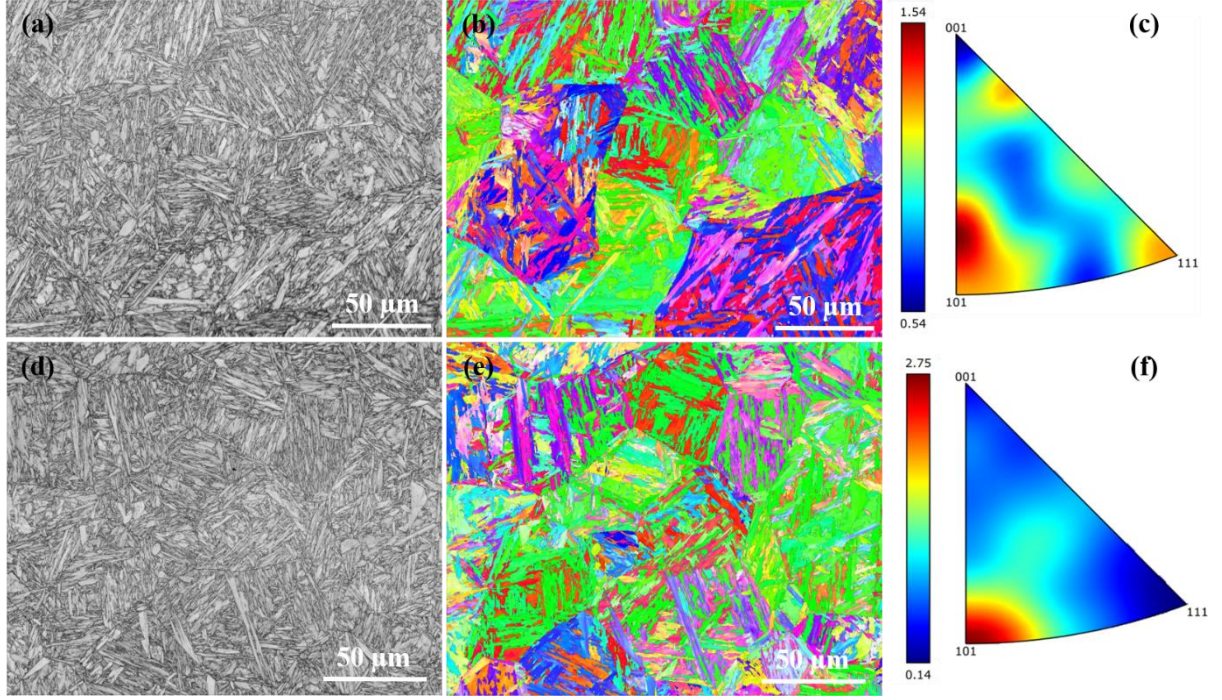


Fig. 15: IQ, IPF map, and IPFs for (a-c) G-AWS and (d-f) G-AWS PB samples. The IPF in (c) and (f) are in a rolling direction.

The martensitic microstructure is present throughout the matrix for the G-AWS sample (Fig. 15a). However, decomposed martensite is dominant in the G-AWS PB sample, as seen in Fig. 15d and e. It is worth noting that there is no tangible change in the grain size after the paint-baking treatment. Moreover, the microstructure and the grain size after the Gleeble simulation show a good agreement with the actual welded samples. As shown in Fig. 15c, a comparatively random texture can be seen after welding where a strong component is formed between $\langle 101 \rangle$ and $\langle 001 \rangle$ peaks with a weak component between $\langle 001 \rangle$ and $\langle 111 \rangle$. This is the common welding texture with a reduced intensity (1.54). However, after PB (Fig. 15f), the strong component formed between $\langle 101 \rangle$ and $\langle 001 \rangle$, peaks are moving toward the $\langle 101 \rangle$ direction which is implying that for most of the grain, their $\langle 101 \rangle$ direction is parallel to the rolling direction (RD). The texture intensity also increased after PB (from 1.54 to 2.75) which can be due to the fact that the grains with their $\langle 101 \rangle$ direction slightly deviated from RD (in the AWS sample), start to reorient and reduced their angle from RD until their $\langle 101 \rangle$ direction is parallel to $\langle 111 \rangle$. This texture is the favorable orientation for BCC materials. The formation of such texture components is believed to affect the crack propagation during post-welding loading, which will be discussed in the following.

4 Discussion

4.1 Effect of PB on Mechanical Properties

As shown in the previous section, the mechanical properties of resistance spot welds of Q&P 980 steels were remarkably improved after the PB treatment. The AWS sample exhibited intergranular fracture and failed through the PAGB (see Fig. 9 and Fig. 11), and the load-bearing capacity of the spot weld increased significantly after PB. To investigate the reason for this behavior, a CTT test was interrupted in the sudden increase in load-bearing capacity regime (see Fig. 16a) and its corresponding cross-section is shown in Fig. 16b. The cross-section of the interrupted CTT test of the AWS-PB showed that the failure proceeds through the corona bond at the faying surface, and the failure in the interrupted test arrested at the fusion boundary, as shown in Fig. 16c. The corona bond turned into a crack at the sheet-sheet interface during cross-tensile loading (where the fracture initiated in both the AWS and AWS-PB samples) and as it may be seen, the cracks blunted at the nugget boundary. A high-magnification SEM image shows the presence of decomposed martensite (tempered martensite) at the blunted crack tip (see Fig. 16d). Crack propagation during loading is indicated by a small drop in the load-displacement curve. However, crack propagation occurs at higher displacements in the paint-baked sample as compared to the AWS sample. As discussed earlier, the higher displacement in the AWS-PB sample before the crack propagation was due to the change in the crack propagation path, which ran through the UCHAZ region rather than through the halo ring. This change in cracking direction results in pull-out fracture in the AWS-PB weld. Furthermore, here the interrupted tests show the blunting of the crack at the fusion boundary. Accordingly, it can be speculated that the reason for the sudden change in the slope of the load-displacement curve after PB (increase in loading) is more likely due to the change in the crack path where the propagation is diverted.

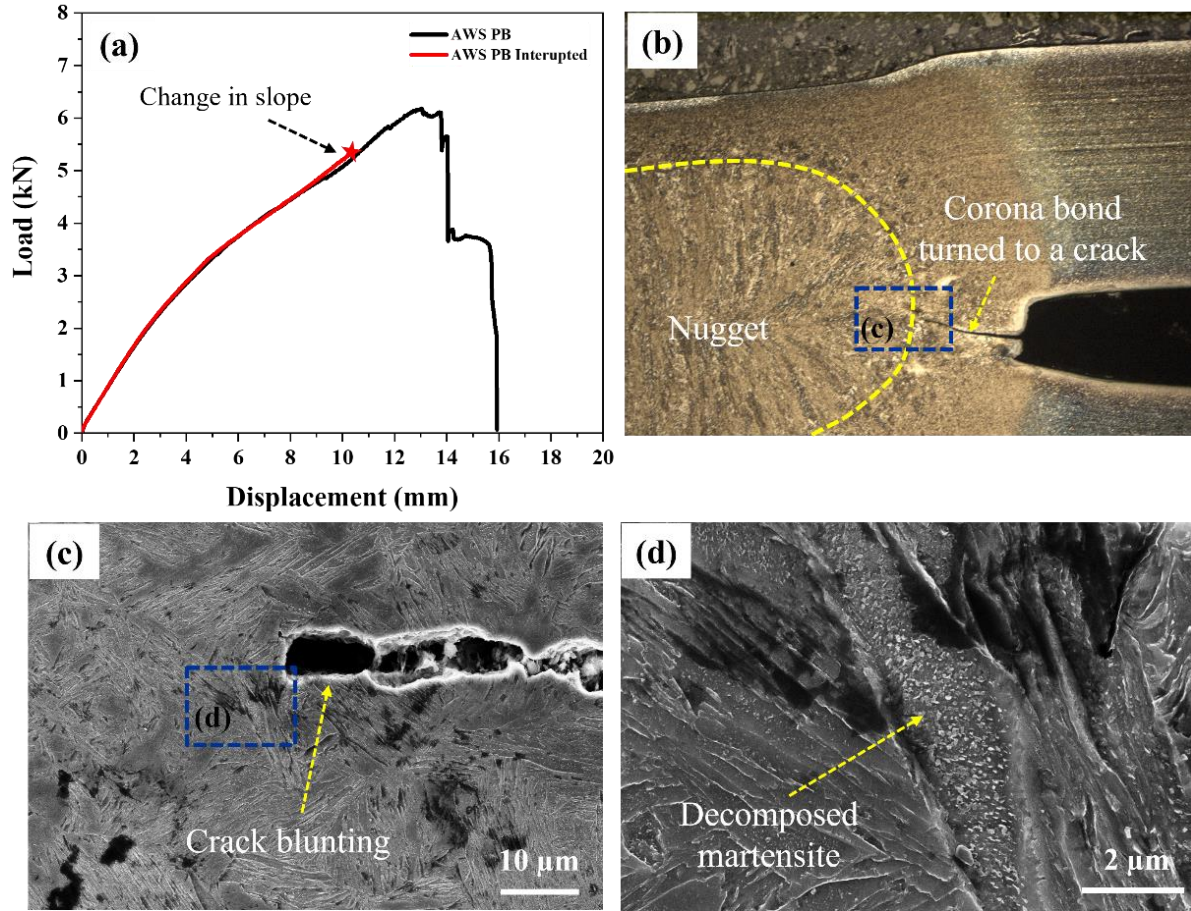


Fig. 16: (a) load-displacement curve for interrupted CTT showing the sudden increase in the slope, (b) cross-sectional OM image after the interrupted CTT for the AWS-PB interrupted sample, and (c and d) high magnification SEM images showing blunting of crack and presence of decomposed martensite, respectively, for the AWS-PB interrupted sample.

Then, the crack will change its path from the fusion boundary toward the UCHAZ which requires more energy, and therefore an increase in the loading will be seen in the load-displacement curve, after the PB. Here it is worthwhile mentioning that the texture is also another factor playing a key role in diverting the crack propagation [33,34]. As shown in Fig. 15d, the texture intensity enhanced after the PB where a strong $\langle 101 \rangle$ texture was seen in RD. It can be inferred that once a crack advances through the microstructure, a large portion of grains have orientations ($\langle 101 \rangle$ direction along RD) unfavorable for the crack propagation which thereby reduces the crack propagation rate. For further propagation, the load should increase which will help the crack to change its path toward regions (toward UCHAZ) with less intensified texture so it can propagate easily. Additionally, along with the grains' orientation, the surrounding microstructure is also another decisive factor in crack propagation behavior.

The mechanical properties of the spot welds are directly related to their local microstructure. It has been reported that the disintegration of martensite present in the microstructure, is one of the reasons behind the increase in cross-tensile properties [6]. The detailed microstructure of the Gleeble samples is shown in Fig. 17. The presence of lath martensite was observed in the G-AWS sample and its corresponding selected area diffraction (SAD) pattern was shown in the inset (see Fig. 17a). There was no other reflection than the BCC in the SAD pattern. G-AWS PB sample, on the other hand, consisted of a fully decomposed martensite microstructure, that can be confirmed by the corresponding SAD pattern (see Fig. 17b). The presence of ϵ -carbide in the martensitic matrix was indexed from the SAD pattern. This ϵ -carbide is assumed to be the reason for the improved CTS after the paint baking.

A significant amount of research has been conducted by different researchers, and they came up with different theories regarding load-bearing improvements after PB. Tumuluru [35] investigated the effect of PB in DP and TRIP steels and explained that the formation of the Cottrell atmosphere is the reason for improving the mechanical properties. Although he tried to explain the reason for the improvement in properties after PB, he failed to clarify the source for the geometrically necessary dislocations (GNDs) density to form this Cottrell atmosphere. After PB, due to the reduced microstructural inhomogeneities, as discussed in section 3.3, a lower GNDs density is expected. EftekhariMilani *et al.* [6] reported that the possibility of an increase in the CTS could be due to the metallurgical transformation that happens in martensite or retained austenite. Moreover, they discussed that the high-carbon martensite present in the steel can transform into tempered martensite by virtue of carbide formation during the PB, which leads to significant improvements in both CTS and fracture toughness.

During CTT, two different mechanisms are concurrently competing, the plastic deformation by dislocation multiplication and their further movement, and crack propagation. Whenever the dislocation movement happens during the application of stress, plastic deformation will occur, and that will dominate rather than the crack propagation mechanism in the early stages of the CTT, as seen in Fig. 16a. Once the hardening enters the recovery stage, due to the dynamic recovery processes (e.g., substructure formation, dislocation annihilation, etc.), the overall dislocation density will decrease, and the material will not undergo more dislocation multiplication. Then the crack propagation starts. It is widely accepted that martensite has a higher dislocation density than decomposed martensite, which means due to the dislocation multiplications and entanglement, the hardening and recovery stages can be reached very fast. However, the ϵ -carbides present in the decomposed martensite can interact with dislocations and can absorb more energy by pinning the dislocations and impeding their movement [23]. Therefore, the AWS sample did not have enough room for deformation during CTT; since it had a martensitic microstructure, thereby easily

reaching a point where crack propagation dominates within a comparatively lower strain value. AWS-PB, on the other hand, has a microstructure consisting of decomposed martensite with ϵ -carbides, which can delay crack propagation by absorbing more energy via delaying the dislocation movement. Accordingly, this will lead to strength and toughness improvement after PB.

It was also reported that the stacking fault energy (SFE) of decomposed martensite is lower than that of fresh martensite. On the other hand, it is well established that the SFE of ferrite is lower than that of martensite [36]. Hence, it can be assumed that the decomposed martensite is considered a transition phase towards the transformation of ferrite and cementite from the fully martensite microstructure. It can be concluded that the improvement in CTS after PB is due to the presence of ϵ -carbide in the martensitic matrix, which can pin the movement of dislocations and delay crack propagation. Nonetheless, Speich *et al.* [37] reported that tempering at temperatures higher than 200°C can activate the transformation of retained austenite into martensite in some particular conditions. However, there was no clear evidence found in their study for transformations happening in retained austenite. More investigations will be done to explore the effect of retained austenite during the paint-baking process on peak load and toughness improvement, and the micro-mechanisms behind mechanical properties advancement in the presence of such transitional phases, in our future studies.

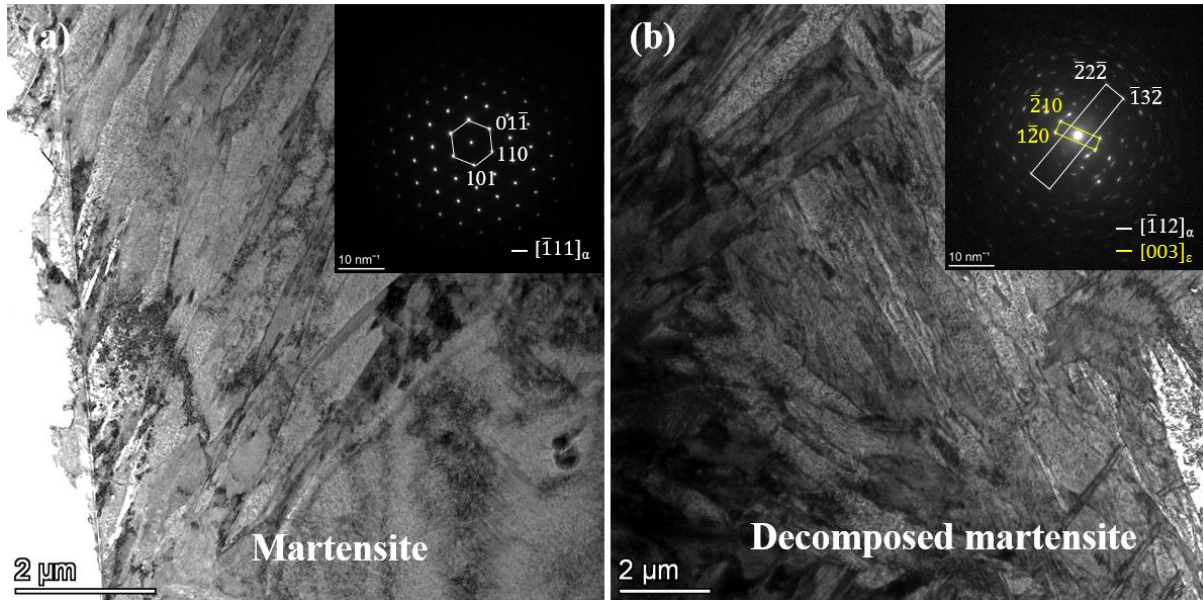


Fig. 17: TEM images and their corresponding SAD pattern showing the microstructure in (a) G-AWS and (b) G-AWS PB samples.

4.2 Fracture Macro-mechanism (Intergranular vs. Trans-granular)

It was shown previously that the failure for the AWS sample is a partial thickness (Mode 3) type through the halo ring and it is a pull-out failure (Mode 1) through the UCHAZ in the AWS-PB sample. To reveal the reason behind the intergranular fracture for the AWS sample, a detailed TEM-EDS analysis was conducted on the Gleeble samples. EDS analysis on typical PAGBs of G-AWS and G-AWS PB are shown in Fig. 18. It can be seen that (Fig. 18a) the G-AWS sample has a clear segregation of C, Mn, Al, and Cr to its PAGBs. It's also worth noting that P segregation hasn't been observed in the G-AWS condition. Furthermore, PB treatment improves the grain boundary properties by redistributing the alloying elements to the matrix (See Fig. 18b) reducing or eliminating their segregation. Hence, the heat treatment improves the grain boundary characteristics by diffusing out C, Mn, Al, and Cr to the grain interior, thereby improving the crack propagation properties by changing the crack propagation path from an intergranular to a transgranular mechanism.

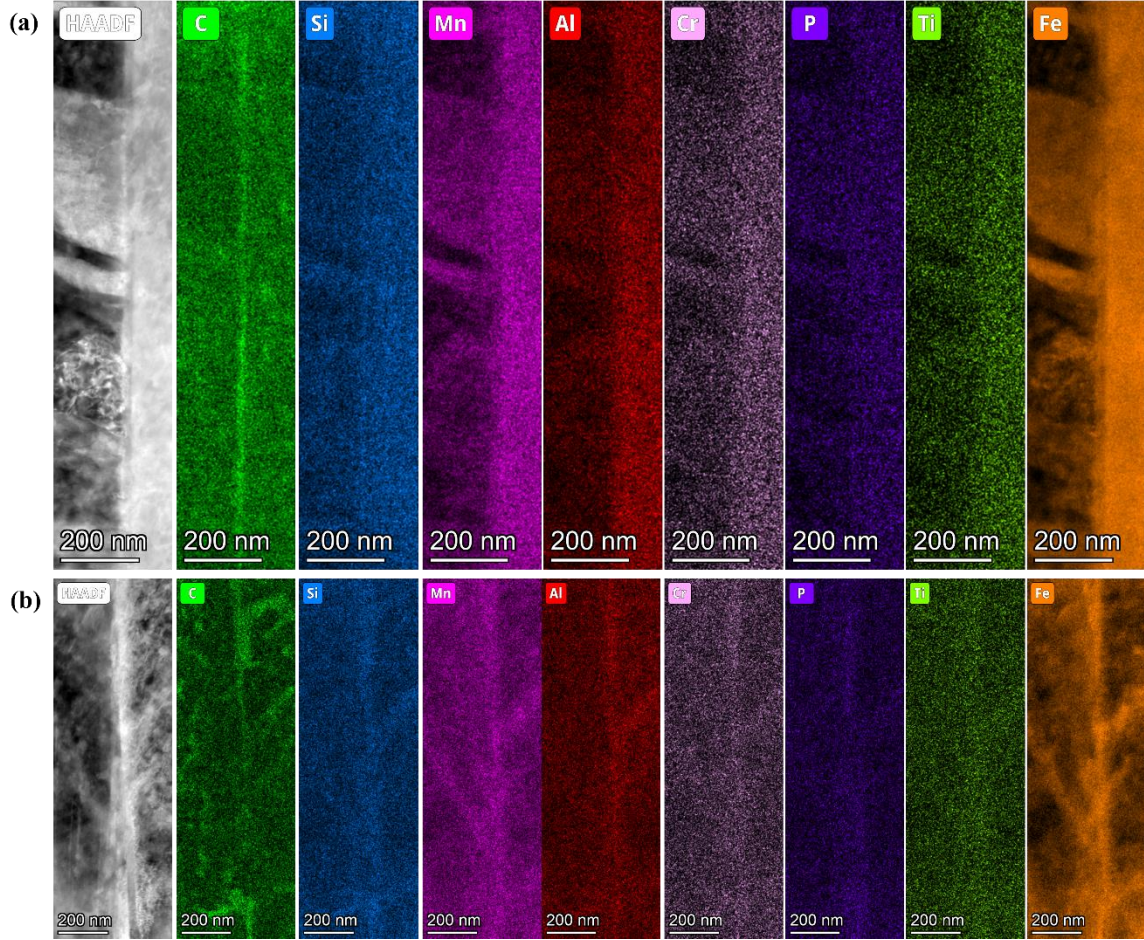


Fig. 18: TEM-EDS images showing (a) elemental segregation in the grain boundaries for the G-AWS sample, and (b) no clear/reduced segregation for the G-AWS PB sample.

It has been reported that during the transformation of austenite into martensite (diffusion-less transformation), the solubility limit of carbon largely exceeds. This thermal activation triggers carbon to segregate to the grain boundaries since the grain boundaries are the favorable sites for defects, such as dislocations to nucleate, which further increase the residual stresses [38]. It is also worth noting that the high carbon content in the grain boundary also can stabilize the austenite during transformation, thereby forming a thin film of retained austenite at the grain boundaries. This presence of thin austenitic film can cause embrittlement of the grain boundaries during loading. The presence of austenite also implies a higher amount of Mn in the grain boundaries. Park *et al.* [17] reported that C and other substitutional alloying elements can segregate to the grain boundary during welding, further deteriorating the fracture behavior. However, PB treatment was found to facilitate the redistribution of C and Mn from the grain boundaries to the grain interiors. Furthermore, the segregation of C and other alloying elements to the grain boundary

mainly depends on the activity of each of these elements. We believe that the rate of diffusion and the factors favouring this diffusion are out of the scope of this work and will be addressed in follow-up work.

It is well reported that regardless of the welding schedules, the crack will always start forming at the sheet-sheet interface due to the stress concentration during CTT [4]. Therefore, for both AWS and AWS-PB samples, the preliminary crack initiation mechanism remains the same. However, the microstructure surrounded by the sheet-sheet interface (UCHAZ and FZ) serves a significant role in the crack propagation behavior, thereby determining the mechanical and fracture properties of the spot welds. In general, cracks will only propagate when the stress applied at the crack tip exceeds the material's fracture toughness [39,40]. In this study, in the AWS sample, the crack also initiates at the solidified liquation products and propagates through the halo ring. It is believed from the results of the G-AWS sample that the halo ring and the region near the halo (UCHAZ) have the segregation of C, Mn, Al, and Cr, which decreases the grain boundary energy (bonding nature of the grain boundary) thereby experiences a low energy brittle fracture. On the contrary, the PB treatments change the fracture mode to pull out through the UCHAZ by modifying the grain boundary characteristics and altering the crack propagation mechanism from intergranular to transgranular, due to the evolution of the surrounding microstructure and texture. That means the change in crack propagation path in the PB sample was due to the transformation of martensite to decomposed martensite, with ϵ -carbide, and the redistribution of C, Mn, Al and Cr from the grain boundary to the grain interior. Both microstructural changes, delays the crack propagation and improves the maximum strength and toughness of the PB samples. In other words, the PB treatment enhances atomic mobility, and grain growth. This will allow the redistribution of alloying elements as well as the disappearance of the solidified liquation from the grain boundaries, thereby changes the brittle (for the AWS sample) to the ductile fracture. The delayed crack propagation is also partially supported by the change in the microstructure and texture of the UCHAZ since the toughness of the decomposed martensite is higher than that of the high-carbon martensite (AWS sample). Therefore, PB is shown to be very essential in terms of mechanical and fracture behavior improvement for the welded samples.

5 Conclusions

The effect of paint baking on spot-welded Q&P 980 steel under cross-tension loading was studied. Moreover, Gleeble simulations were done to simulate the upper-critical heat-affected zone in the weld samples, followed by uniaxial tensile testing to gain insights into the effects of PB on material properties and fracture micro- and macro-mechanisms. The following conclusions were drawn from this study:

- 1) Post-weld PB heat treatment significantly improved the load-bearing capacity of the Q&P 980 steel spot welds in the cross-tension loading conditions. The application of PB increases the weld strength (peak load) by 34 and 17% for AWS and DPS samples, respectively. The absorbed energy was improved by the order of 102% and 67 %, for AWS-PB and DPS-PB samples, respectively. PB significantly decreased the differences in mechanical properties between the AWS and DPS welds. Unless strength needs to be maximized further, the DPS schedule is not required under normal manufacturing conditions to mitigate sub-optimal joint strength resulting from the halo.
- 2) PB changes the fracture mode from Mode 3 (partial thickness fracture) to Mode 1 (button pull) in AWS welds. However, DPS weld failed with Mode 1 fracture regardless of whether the welds were subjected to PB or not. AWS welds failed in a brittle manner whereas the AWS-PB failed in a ductile manner.
- 3) The presence of solidified liquation regions at the grain boundaries in the AWS sample also serves a role in crack initiation. Cracks can be initiated from the solidified liquation regions during loading due to the presence of retained austenite structures and the strain inhomogeneity between austenite and martensite, culminating in stress concentrations. However, due to the enhanced atomic mobility and grain growth during the PB, the weak solidified liquation regions are consumed by surrounding grains and the redistribution of Mn and Si to the grain interiors. This further improves energy absorption due to the elimination of crack initiation sources.
- 4) Microstructural analysis revealed that PB resulted in the decomposition of the martensite and formation of ϵ -carbide, which is associated with an increase in the fracture toughness and is likely the major factor improving the CTS of Q&P 980.
- 5) PB was associated with a change in the micro-texture where an intensified texture with the $\langle 101 \rangle$ direction of most grains parallel to the rolling direction was observed. This texture was found to be less favorable for crack propagation, increasing the energy required for a crack to continue its movement leading to the change in the crack path which was accompanied by an increase in the loading rate during CTT.
- 6) The presence of segregation of C, Mn, Al, and Cr was observed along the grain boundaries in G-AWS conditions, which was further redistributed throughout the grains after the PB treatments. The presence of C and Mn in the grain boundaries decreases the grain boundary cohesion thereby decreasing the fracture toughness of the grain boundaries.

Acknowledgment

The authors would like to acknowledge the financial support from the Arcelor-Mittal Dofasco G.P. in Hamilton, Canada, and the Natural Sciences and Engineering Research Council (NSERC). The electron microscopy was performed at the Canadian Centre for Electron Microscopy (also supported by NSERC and other government agencies).

Credit author statement

Dileep Chandran Ramachandran: Conceptualization, Methodology, Investigation, Data curation, Visualization, Validation, Writing - original draft. **Olakunle Timothy Betiku:** Data curation, Writing - review & editing. **Mohammed Shojaee:** Writing - review & editing. **Adib Salandari-Rabori:** Writing - review & editing. **Abdelbaset R.H. Midawi:** Data curation. **Ji-Ung Kim:** Data curation. **Reza Bakhtiari:** Visualization, **Andrew Macwan:** Project administration, Resources, Funding acquisition, Writing - review & editing. **Elliot Biro:** Supervision, Validation, Resources, Funding acquisition, Writing - review & editing.

References

- [1] M. Shojaee, A.R.H. Midawi, B. Barber, H. Ghassemi-Armaki, M. Worswick, E. Biro, Mechanical properties and failure behavior of resistance spot welded third-generation advanced high strength steels, *J. Manuf. Process.* 65 (2021) 364–372. <https://doi.org/10.1016/J.JMAPRO.2021.03.047>.
- [2] X.D. Liu, Y.B. Xu, R.D.K. Misra, F. Peng, Y. Wang, Y.B. Du, Mechanical properties in double pulse resistance spot welding of Q&P 980 steel, *J. Mater. Process. Technol.* 263 (2019) 186–197. <https://doi.org/10.1016/j.jmatprotec.2018.08.018>.
- [3] W. Zhang, S. Yang, Z. Lin, W. Tao, Weld morphology and mechanical properties in laser spot welding of quenching and partitioning 980 steel, *J. Manuf. Process.* 56 (2020) 1136–1145. <https://doi.org/10.1016/J.JMAPRO.2020.05.057>.
- [4] M. Pouranvari, S.P.H. Marashi, Critical review of automotive steels spot welding: Process, structure and properties, *Sci. Technol. Weld. Join.* 18 (2013) 361–403. <https://doi.org/10.1179/1362171813Y.00000000120>.
- [5] B. Figueredo, D.C. Ramachandran, A. Macwan, E. Biro, Failure behavior and mechanical

- properties in the resistance spot welding of quenched and partitioned (Q&P) steels, *Weld. World.* 65 (2021) 2359–2369. <https://doi.org/10.1007/s40194-021-01179-z>.
- [6] P. Eftekharimilani, E.M. van der Aa, R. Petrov, M.J.M. Hermans, I.M. Richardson, Understanding the Effect of a Paint Bake Cycle on the Microstructure–Mechanical Properties Relationship of a Resistance Spot Welded Advanced High Strength Steel, *Metall. Mater. Trans. A Phys. Metall. Mater. Sci.* 49 (2018) 6185–6196. <https://doi.org/10.1007/s11661-018-4912-9>.
- [7] M. Shamsujjoha, C.M. Enloe, A.C. Chuang, J.J. Coryell, H. Ghassemi-Armaki, Mechanisms of paint bake response in resistance spot-welded first and third generation AHSS, *Materialia*. 15 (2021) 100975. <https://doi.org/10.1016/j.mtla.2020.100975>.
- [8] O. Sherepenko, O. Kazemi, P. Rosemann, M. Wilke, T. Halle, S. Jüttner, Transient softening at the fusion boundary of resistance spot welds: A phase field simulation and experimental investigations for al–si-coated 22mnb5, *Metals (Basel)*. 10 (2020). <https://doi.org/10.3390/met10010010>.
- [9] D.C. Ramachandran, B. Figueredo, O. Sherepenko, W. Jin, Y. Do Park, E. Biro, A study on improving the mechanical performance by controlling the halo ring in the Q&P 980 steel resistance spot welds, *J. Manuf. Process.* 75 (2022) 320–330. <https://doi.org/10.1016/j.jmapro.2022.01.019>.
- [10] D.C. Ramachandran, B.D.N. Figuerdo, E. Biro, The Fusion boundary, a potential failure path in the resistance spot welding of Q&P steels, *Sheet Met. Weld. Conf. XIX. 2003* (2021) 3A–2.
- [11] O. Sherepenko, S. Jüttner, Transient softening at the fusion boundary in resistance spot welded ultra-high strengths steel 22MnB5 and its impact on fracture processes, *Weld. World.* 63 (2019) 151–159. <https://doi.org/10.1007/s40194-018-0633-3>.
- [12] A. Mohamadizadeh, E. Biro, M. Worswick, Shear band formation at the fusion boundary and failure behaviour of resistance spot welds in ultra-high-strength hot-stamped steel, *Sci. Technol. Weld. Join.* 25 (2020) 556–563. <https://doi.org/10.1080/13621718.2020.1773057>.
- [13] O.T. Betiku, M. Shojaee, O. Sherepenko, A.R.H. Midawi, A.M. Chertov, H. Ghassemi-Armaki, R.G. Maev, E. Biro, Optimizing post-weld performance of press-hardened steel resistance spot welds by controlling fusion zone porosity, *Weld. World.* 1 (2022) 1–14. <https://doi.org/10.1007/S40194-022-01332-2/FIGURES/15>.
- [14] O.T. Betiku, D.C. Ramachandran, A. Ghatei-kalashami, C. DiGiovanni, O. Sherepenko, H. Ghassemi-Armaki, E. Biro, Improving the mechanical performance of press-hardened steel

- resistance spot welds via in-situ grain refinement, *J. Mater. Process. Technol.* 320 (2023) 118122. <https://doi.org/10.1016/J.JMATPROTEC.2023.118122>.
- [15] P. Eftekharimilani, E.M. van der Aa, M.J.M. Hermans, I.M. Richardson, Microstructural characterisation of double pulse resistance spot welded advanced high strength steel, *Sci. Technol. Weld. Join.* 22 (2017) 545–554. <https://doi.org/10.1080/13621718.2016.1274848>.
- [16] G. Park, K. Kim, Y. Lee, S. Uhm, · Changhee Lee, Kinetics Study on Low-Temperature Tempering of Martensitic Phase in Medium Mn Steel Weldment During Paint-Baking Heat Treatment, *Met. Mater. Int.* 28 (2022) 1157–1168. <https://doi.org/10.1007/s12540-021-00985-w>.
- [17] G. Park, K. Kim, S. Uhm, C. Lee, Remarkable improvement in resistance spot weldability of medium-Mn TRIP steel by paint-baking heat treatment, *Mater. Sci. Eng. A.* 766 (2019) 138401. <https://doi.org/10.1016/j.msea.2019.138401>.
- [18] S.M. Manladan, Y.J. Jang, Y. Do Park, Effect of paint baking on the halo ring and mechanical behavior of 30MnB5 hot-stamped steel resistance spot welding joints, *J. Mater. Res. Technol.* 24 (2023) 4756–4761. <https://doi.org/10.1016/j.jmrt.2023.04.144>.
- [19] D.C. Ramachandran, E. Biro, Improving the mechanical performance of Q&P980 steel resistance spot welds by pulsing technique, *Sheet Met. Weld. Conf. XIX.* (2021) 3A – 4.
- [20] N. Yurioka, T. Kasuya, Carbon Equivalent and Multiplying Factor for Hardenability of Steel, *Weld. Res. Suppl.* 6 (1993) 263–268. https://doi.org/https://app.aws.org/wj/supplement/WJ_1993_06_s263.pdf.
- [21] X. Han, Effect of Surface Condition on Resistance Spot Welding of Advanced High Strength Steel., University of Waterloo, 2019. <https://uwspace.uwaterloo.ca/handle/10012/15149?show=full>.
- [22] AWS D8 Committee on Automotive Welding, Test methods for evaluating the resistance spot welding behaviour of automotive sheet steel materials, 2012.
- [23] D.C. Ramachandran, A.R.H. Midawi, M. Shojaee, O. Sherepenko, H. Ghassemi-Armaki, E. Biro, A comprehensive evaluation of tempering kinetics on 3rd generation advanced high strength steels, *Materialia.* 26 (2022) 101644. <https://doi.org/10.1016/j.mtla.2022.101644>.
- [24] AWS, AWS D8.9M:2012 Test Methods for Evaluating the Resistance Spot Welding Behavior of Automotive Sheet Steel Materials, (2012) 1–106.

- [25] AWS, AWS D8.1M:2007 Specification for Automotive Weld Quality-Resistance Spot Welding of Steel, 2007.
- [26] X.D. Liu, Y.B. Xu, R.D.K. Misra, F. Peng, Y. Wang, Y.B. Du, Mechanical properties in double pulse resistance spot welding of Q&P 980 steel, *J. Mater. Process. Technol.* 263 (2019) 186–197. <https://doi.org/10.1016/j.jmatprotec.2018.08.018>.
- [27] H. Li, W. Yan, Z. Li, B. Mariusz, J. Senkara, Y. Zhang, Numerical and experimental study of the hot cracking phenomena in 6061/7075 dissimilar aluminum alloy resistance spot welding, *J. Manuf. Process.* 77 (2022) 794–808. <https://doi.org/10.1016/J.JMAPRO.2022.04.001>.
- [28] M. Alizadeh-Sh, S.P.H. Marashi, Resistance spot welding of dissimilar austenitic/duplex stainless steels: Microstructural evolution and failure mode analysis, *J. Manuf. Process.* 28 (2017) 186–196. <https://doi.org/10.1016/j.jmapro.2017.06.005>.
- [29] D.C. Saha, I. Chang, Y. Do Park, Heat-affected zone liquation crack on resistance spot welded TWIP steels, *Mater. Charact.* 93 (2014) 40–51. <https://doi.org/10.1016/J.MATCHAR.2014.03.016>.
- [30] Y. Jing, Y. Xu, D. Wang, L. Lu, J. Li, Y. Yu, Improving mechanical properties of welds through tailoring microstructure characteristics and fracture mechanism in multi-pulse resistance spot welding of Q&P980 steel, *Mater. Sci. Eng. A.* 843 (2022) 143130. <https://doi.org/10.1016/J.MSEA.2022.143130>.
- [31] D.C. Ramachandran, J. Moon, C.-H. Lee, S.-D. Kim, J.-H. Chung, E. Biro, Y.-D. Park, Role of bainitic microstructures with M-A constituent on the toughness of an HSLA steel for seismic resistant structural applications, *Mater. Sci. Eng. A.* 801 (2021). <https://doi.org/10.1016/j.msea.2020.140390>.
- [32] D.C. Ramachandran, S.D. Kim, J. Moon, C.H. Lee, J.H. Chung, E. Biro, Y. Do Park, Classification of martensite-austenite constituents according to its internal morphology in high-strength low alloy steel, *Mater. Lett.* 278 (2020). <https://doi.org/10.1016/j.matlet.2020.128422>.
- [33] C. Meng, Z. Zhang, J. Yang, W. Zhuang, H. Shi, Effect of laser bionic treatment on fatigue crack growth behavior of TIG welded AZ31B magnesium alloy joint, (2022). <https://doi.org/10.1016/j.matlet.2022.132334>.
- [34] Z. Liu, F. Li, P. Xia, S. Bai, Y. Gu, D. Yu, S. Zeng, Mechanisms for Goss-grains induced crack deflection and enhanced fatigue crack propagation resistance in fatigue stage II of an AA2524

- alloy, (2014). <https://doi.org/10.1016/j.msea.2014.12.021>.
- [35] M. Tumuluru, Effects of baking on the structure and properties of resistance spot welds in 780 MPa dual-phase and TRIP steels, *Weld. J. (Miami, Fla)*. 89 (2010).
- [36] Y.K. Lee, C.S. Choi, Driving force for $\gamma \rightarrow \epsilon$ martensitic transformation and stacking fault energy of γ in Fe-Mn binary system, *Metall. Mater. Trans. A Phys. Metall. Mater. Sci.* 31 (2000) 355–360. <https://doi.org/10.1007/s11661-000-0271-3>.
- [37] G.R. Speich, W.C. Leslie, Tempering of Steel., *Met. Trans.* 3 (1972) 1043–1054. <https://doi.org/10.1007/BF02642436>.
- [38] L. Morsdorf, A. Kashiwar, C. Kübel, C.C. Tasan, Carbon segregation and cementite precipitation at grain boundaries in quenched and tempered lath martensite, *Mater. Sci. Eng. A*. 862 (2023). <https://doi.org/10.1016/J.MSEA.2022.144369>.
- [39] D.J. Radakovic, M. Tumuluru, Predicting resistance spot weld failure modes in shear tension tests of advanced high-strength automotive steels, *Weld. J. (Miami, Fla)*. 87 (2008) 96–105.
- [40] A. Chabok, E. Galinmoghaddam, J.T.M. De Hosson, Y.T. Pei, Micromechanical evaluation of DP1000-GI dual-phase high-strength steel resistance spot weld, *J. Mater. Sci.* 2018 542. 54 (2018) 1703–1715. <https://doi.org/10.1007/S10853-018-2886-Z>.