

Elevated Temperature Formability of Precipitation Hardenable Sheet Aluminum Alloys

by

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Statement of Contributions

The following co-authors have contributed to the current work:

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Professor Cliff Butcher contributed in the numerical implementation of the ISO12004:2-2008 standard.

Dr. Massimo Di Ciano supported in the sample preparation and tensile testing for the AA6013 tensile aging studies, as well as the kinematics model calibration (Appendix C).

Co-Authors Farinez Khameneh and Kenneth Cheong completed the formability and stretch-bend testing of AA5182-O, respectively.

Negar Baghbanaghaie performed the Differential Scanning Calorimetry and Isothermal Calorimetry on AA6013 presented in Appendix C.

All named co-authors on the publications originating from this thesis contributed in the review process prior to publication.

The balance of the research is my own work.

Abstract

This thesis investigates the elevated temperature forming behavior of two age-hardenable aluminum alloys, AA6013 and a developmental 7000-series alloy (denoted AA7xxx), intended for automotive applications. Warm (150°C - 250°C) and hot ($\geq 300^\circ\text{C}$) forming conditions were considered as well as a range of starting tempers to ascertain the potential benefits of elevated temperature forming. A major outcome of this work was the development and detailed documentation of a physically motivated method to assess formability, termed the Enhanced Curvature Method (ECM). The ECM utilizes stereoscopic digital image correlation (DIC) to detect the onset of necking based on changes in surface curvature.

The ECM was validated at room temperature using AA5182-O sheet which exhibits a strong dynamic strain ageing effect that affects strain localization. The plane strain limit strains obtained using the ECM in Marciniak, Nakazima, and stretch-bend formability tests were consistent with two other formability characterization methods developed for room temperature, the ISO12004-2:2008 standard and the rate-dependent method developed by Volk and Hora (2011) termed the “linear best fit” (LBF) method. The average Marciniak plane strain limit strain for AA5182-O obtained using the ECM was 0.198, and in close agreement with the theoretical limit strain of 0.194 obtained from the instability model of Swift.

The ECM was subsequently applied to characterize the room and elevated temperature formability of 2 mm thick AA6013 and AA7xxx. Both alloys were tested using a 76.2 mm wide dog-bone specimen geometry in a Nakazima-type test, that produced near plane-strain conditions. For AA6013-T6, when characterized using the ECM, warm forming did not offer a meaningful improvement in formability with increasing temperature over the tested ranges. The plane strain forming limit at both 230°C and room temperature (RT) was measured as approximately 0.18 with comparable scatter bands.

To improve warm formability, AA6013 in a pre-aged (PA) temper was investigated. The PA temper was produced by solutionizing the material at 560°C for 10 minutes, water quenching, aging at 100°C for 4 hours then followed by a second water quench. The room temperature formability of the PA temper was significantly higher than that of the peak-aged temper with a major limit strain of 0.28. The elevated temperature formability of the PA temper, however, decreased relative to its room temperature formability. At 230°C, the major limit strain was reduced to 0.21 and attributed to aging occurring during the warm forming process. A significant outcome of the AA6013 work was the observation that a relatively short secondary aging cycle (less than 410 s) at 235°C after forming, combined with a pre-aged initial temper, could result in a stress-strain response similar to the baseline T6 temper, regardless of a subsequent paint-bake. This processing route offers a potential pathway to achieve T6 strength levels with a greater than 50%

improvement in formability when comparing the RT PA limit strain of 0.28 to the baseline T6 limit strain of 0.18.

The effect of initial temper on the warm formability and aging response of AA7xxx was also investigated for three tempers: pre-aged (PA), peak-aged (T6), and over-aged (T76), all tested using a 76.2 mm dog-bone specimen. The PA temper exhibited the highest formability at room temperature, with a major true limit strain of 0.21. At 150°C, the limit strain for the PA temper increased to 0.24, representing a modest improvement. At higher temperatures (175°C and 200°C), the results were not as clear, with measured dome heights decreasing and strains relatively unchanged. Similar to the AA6013 PA temper, the formability gains at elevated temperature for the AA7xxx PA temper were modest. Warm forming at 200°C, followed by a paint-bake, resulted in both the PA and T6 tempers achieving ductility and strength levels similar to those of the as-received T76 temper.

The hot-stamping and die quenching (DQ) process was investigated as an alternative means to enhance the formability of AA7xxx. The DQ process resulted in a major limit strain of 0.76 in plane strain. Under an interrupted, isothermal (300°C) hot-stamping operation, a major limit strain of 0.53 in plane strain was observed. Both limit strains were significantly higher than the room temperature formability of 0.12 measured for the as-received T76 temper and that achieved using warm forming. Surface defects, however, such as orange peel, were observed for major strain levels in excess of approximately 0.45.

A major conclusion of this work is that while both AA6013 and AA7xxx alloys demonstrated only modest improvements in formability when warm formed in their as-received conditions, a better balance of strength and formability can be achieved by leveraging pre-aged starting tempers combined with post-stamping aging cycles. The die quenching process results clearly show that excellent formability of AA7xxx series alloys can be achieved with additional process complexity. The ECM, a key outcome of this work, offers an effective and robust method for quantifying formability at both room and elevated temperatures. Future work should explore faster warm forming processes to overcome the rapid aging effects observed in the PA tempers. Additionally, other aluminum alloys should be evaluated for warm forming and die quenching to identify alloys that may offer a greater positive response to elevated temperature forming.

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*“We do not inherit the earth from our ancestors; we
borrow it from our children” ~ Chief Seattle*

This work was inspired by everyone who came before me
and it is dedicated to everyone that will come after me.
Only together can we make the world better.

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1. INTRODUCTION

Stringent fuel economy and emissions regulations (Ministry of Ecology and Environment, 2018; Dornoff & Rodriguez, 2024; National Highway Traffic Safety Administration, 2024) imposed on the global automotive sector have put renewed pressure (Environmental Protection Agency, 2022) on automotive manufacturers to reduce vehicle weight while also increasing vehicle safety. To achieve these weight reductions, high-strength 6000- and 7000-series aluminum alloys are under consideration for the design of vehicle structural components (Li & Ghosh, 2004; Sotirov et al., 2012; Wang et al., 2012; Smith et al., 2017). The high strength of these alloys is attributed to the distribution of fine precipitates within their microstructures that is achieved through tailored heat-treatment processes known as artificial aging (Davis, 2001; Esmaeili et al., 2003; Banhart, 2016).

Heat-treatable aluminum alloys are considered to be at peak strength in the T6 temper, which is typically achieved through a multi-step aging process. During aging, the exposure of the alloy to different temperatures for varying amounts of time promotes microstructural changes that have a positive influence on the final strength of the alloy. In the peak aged (T6) condition, these alloys have specific strengths comparable to advanced and ultra-high strength steels but often poor room temperature formability, making the production of components with complex geometries difficult (Illinich & Luckey, 2014). Temperature dependent processing routes such as warm forming (Shehata et al., 1978; Bolt et al., 2001; Sotirov et al., 2012) and die quenching (also referred to as hot stamping) (Garrett et al., 2005; Omer et al., 2017; Mohamed et al., 2012) have the potential to improve the formability of these alloys.

Formability is a measure of the capability of a sheet metal to be plastically deformed to a desired shape without defects such as wrinkles or cracks. Warm forming of aluminum alloys is a processing route in which a sheet material is deformed at temperatures above 150°C, but below the recrystallization temperature of the alloy (Sotirov et al., 2012). In this temperature range, the ductility of aluminium alloys is believed to be improved, while the flow resistance generally decreases. The warm forming processing route has been studied in detail for non-heat-treatable 5000-series aluminum alloys (Naka et al., 2001; Abedrabbo et al., 2007), and has been shown to improve formability. The warm formability of age-hardenable aluminum alloys has also been studied; however, many studies do not provide so called “safe” strain measurements, or do not detail the methodologies used for determining safe formability strains (Cavaliere, 2002; Li & Ghosh, 2004; Wang et al., 2012; Sotirov et al., 2012). In contrast to warm forming, the die quenching process (Garrett et al., 2005; Omer et al., 2017) involves heating the blanks to a temperature at which the strengthening precipitates dissolve into the microstructure in a process known as solutionizing (Omer et al., 2017). The sheet material is then simultaneously quenched and formed in a

cooled die system. The high cooling rates during quenching suppress precipitation so the as-formed part is in a supersaturated condition that can be heat-treated to the desired temper (Abolhasani et al., 2017).

For continued adoption of precipitation hardenable sheet aluminum alloys within the automotive sector, formability characterization procedures must be developed for elevated temperatures to provide the material data needed to support the tooling and process design phase to stamp complex components. The current thesis aims to characterize the elevated temperature forming performance of heat-treatable sheet aluminum alloy AA6013-T6 and a developmental 7000-series alloy, AA7xxx-T76. A major result from this work is the detailed development, documentation, and implementation of a physically motivated methodology for quantifying formability at room and elevated temperatures, termed the Enhanced Curvature Method (ECM). The influence of the initial temper of each alloy on warm formability is experimentally characterized for a number of different temper conditions, as well as the associated aging response from the warm forming cycle and corresponding paint-bake response. The effect of temperature, strain-rate sensitivity, and time at temperature is explored for a limited number of temper and temperature conditions where the in-situ aging effects on the constitutive response is assessed. For the AA6013 alloy in a pre-aged (PA) temper, a model is calibrated to capture the secondary aging kinetics. For the AA7xxx alloy, the die-quench formability under plane strain loading was also characterized for different processing conditions in an effort to better understand the influence of temperature and strain-rate effects on formability.

1.1. Thesis Organization

The thesis was written in a manuscript-based format comprised of five papers and this synopsis to provide an overview of the research. The organization of the synopsis is as follows. Chapter 2 presents a review of published research relevant to the temperature-dependent characterization of aluminum alloys, formability characterization techniques, and the elevated formability of sheet aluminum alloys. Chapter 3 provides a concise statement of the objectives of this thesis. Chapter 4 provides discussion and a summary of all relevant results, with more detailed discussion and experimental details included in Appendices A-E. Chapter 5 presents the conclusions stemming from this research and recommendations for future areas of research.

2. BACKGROUND AND LITERATURE REVIEW

The use of aluminum alloys in U.S. automotive body-in-white and closure applications is projected to increase from 4% in 2010 to 26% by 2040 (Smith et al., 2017), as summarized in Figure 1. Of this, 10% of future aluminum sheet alloys are expected to be high-strength. To support this increase in usage of high-strength aluminum alloys, new forming processes must be developed to accommodate the relatively poor formability of high-strength aluminum alloys under conventional forming processes (Anyasodor & Koroschetz, 2017; Long et al., 2017). Temperature dependent forming processes may be a possible solution (Sotirov et al., 2012; Kumar & Ross, 2016; Degner et al., 2017). This section provides a concise review of the literature relevant to warm forming of sheet aluminum alloys and identifies potential gaps within the literature that will be addressed in this thesis.

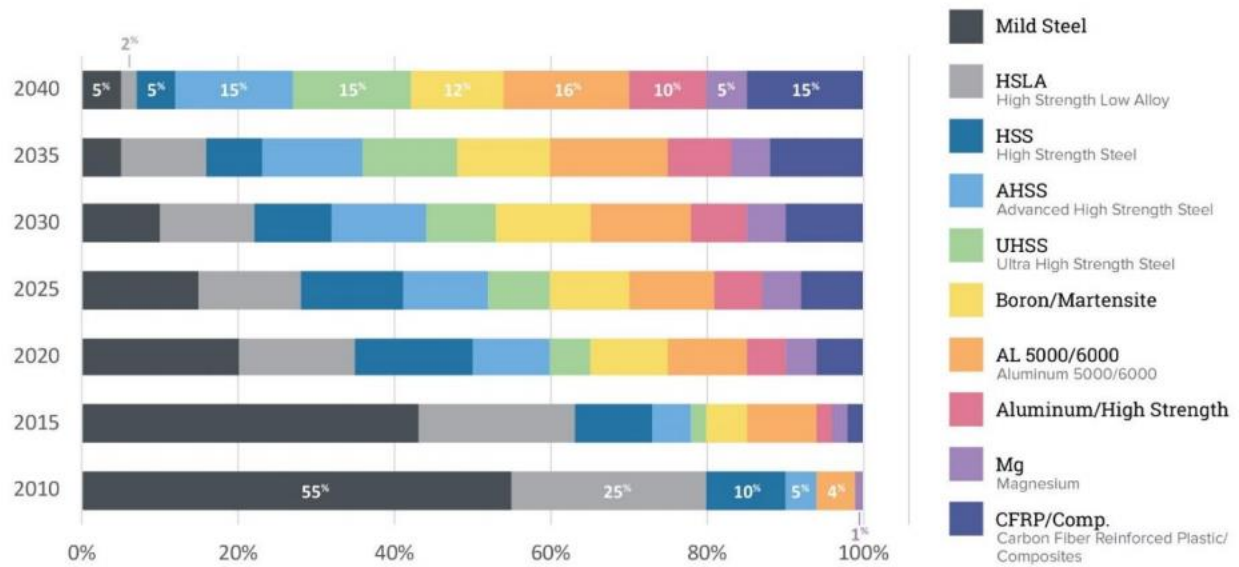


Figure 1: Estimated material distribution in the U.S. Fleet (Body-in-White Plus Closures), 2010 to 2040 (Smith et al., 2017).

This section begins with an overview of age-hardenable sheet aluminum alloys (Section 2.1). Section 2.2 provides an in-depth review of the research concerning development of forming limit diagrams (FLDs), including the current application of digital image correlation (DIC) techniques to formability characterization, relevant testing procedures, and current limitations. In Section 2.3, the warm formability of aluminum alloys is discussed with respect to non-heat-treatable alloys and heat-treatable aluminum alloys, as well as a limited review of the application of die quenching in sheet aluminum alloys. In Section 2.4, a discussion of modelling efforts regarding the formability of aluminum alloys as a function of temperature is presented and emphasizes the need for accurate experimental data for model inputs.

2.1. Sheet Aluminum Alloys

Pure aluminum is a face centred cubic (FCC) structured metal with a density of approximately 2,800 kg/m³. The FCC structure has 12 slip systems to accommodate plastic deformation. When small amounts of alloying elements are added to pure aluminum the resulting alloy can have strength-to-weight ratios that are competitive with ultra high-strength steels. Other positive properties of aluminum alloys include good recyclability, conductivity, and, in select tempers and alloys, good corrosion resistance and ductility (Bagheriasl, 2012). This combination of properties has been a driving force for the integration of sheet and cast aluminum alloys within the automotive sector to improve vehicle light-weighting (National Highway Traffic Safety Administration, 2011; Lewis et al., 2014).

Sheet aluminum alloys can be broadly categorized as age hardening or work hardening. Age-hardenable aluminum alloys have improved strength through multi-step, heat-treatment cycles. An example of a heat-treatment cycle for a 7000-series aluminum alloy is shown in Figure 2, with the inclusion of warm forming and paint-baking steps. The basic heat-treatment cycle consists of three steps: solutionizing, quenching, and natural or artificial aging. These steps and their importance are discussed in more detail in Section 2.1.1.

Age hardening aluminum alloys are encompassed in the 2000-, 6000-, 7000-, and 8000- series of aluminum alloys, as defined by ANSI standard H35.1 (Aluminum Association, 2017). Work hardening alloys fall under the series of 1000-, 3000-, 4000-, and 5000-. Work hardening alloys do not contain alloying elements that form precipitates and derive their limited strength from conventional work hardening techniques. Figure 3 summarizes the primary alloying elements for the listed wrought aluminum alloy series. The remainder of this section will focus on discussion of age-hardenable aluminum alloys.

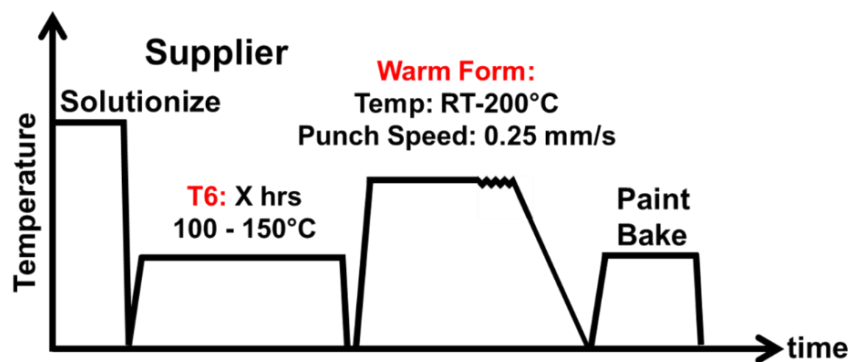


Figure 2: Example heat-treatment and forming process for an AA7000 series alloy.

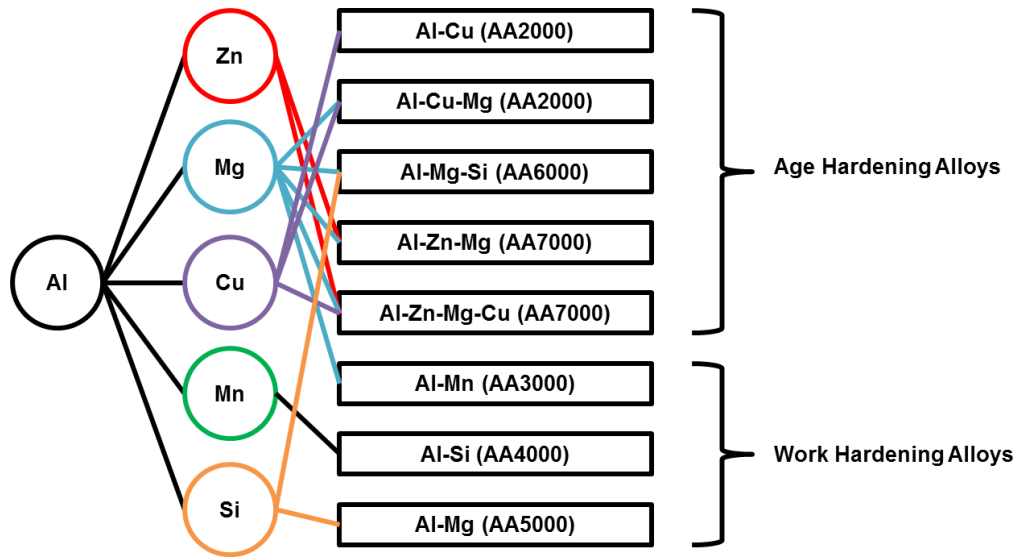


Figure 3: The principal aluminum alloys, reproduced from Davis (1993).

2.1.1. Age-Hardenable Sheet Aluminum Alloys

Age-hardenable or precipitation hardenable aluminum alloys experience changes to their yield and ultimate tensile strengths with time at elevated temperature. Precipitation hardening involves the generation of a fine distribution of precipitate particles within the base alloy that act to impede the motion of dislocations and increase the strength of the alloy. The driving force for the development of precipitates is that solute atoms have complete solubility at high temperature but limited solubility at room temperature, resulting in room temperature super-saturation of solute atoms within the base material phase (Ardell, 1985).

In the first step of precipitation hardening, the aluminum alloy is brought to the solutionizing temperature where all precipitates are dissolved and the solute atoms enter solution. Following solutionizing, the alloy is then quenched and the solubility of the solute atoms within the base material dramatically decreases. The solutionizing processing path results in a metastable condition in which the solute atoms experience a high driving force towards precipitation as a result of the super-saturation within the base alloy. At this stage, the alloy can then be heat-treated using natural (room temperature) or artificial (elevated temperature) aging techniques to achieve the desired final distribution of precipitates that dictate the final strength, corrosion resistance, and alloy stability, among other properties (Smith, 1981).

Specific heat-treatments are based on a careful balancing of the driving force for precipitation and the thermal activation energy available in the system, both of which are sensitive to the alloy composition (Kaufman, 2000). Aluminum alloys are at peak strength and stability in the T6 temper (Smith, 1981). More

generally, heat-treatable aluminum alloys are used with final tempers that are stable over reasonable periods of time, usually at least six months, to facilitate the manufacturing of vehicle components. The stability of an aluminum temper can be influenced by exposure to elevated temperatures, resulting in a degradation of both strength and ductility of the alloy in a process known as over-aging (Esmaeili et al., 2003); thus, the service environment of heat-treatable aluminum alloys should be considered in the design process. The influence of elevated temperature exposure on heat-treatable aluminum alloys is also a consideration in the forming process, as will be discussed in Section 2.3.

In the case of an Al-Cu alloy (Jacobs, 1999), copper atoms (solute) first cluster together in local groups to form Guinier-Preston (GP) zones, which are on the order of 10 nm in size. Guinier-Preston clusters will occur through solutionizing and quenching followed by natural aging; however, peak-strength in age-hardenable alloys is achieved by artificially aging the alloys at elevated temperatures to allow stable growth of the GP zones and the development of β'' precipitates. If the material is over-aged, the β'' will decay into β' or β (fully stable), based on the amount of time at temperature in a process known as Ostwald ripening. Edwards et al. (1998) observed a similar precipitation sequence in an Al-Mg-Si alloy.

In the work of Esmaeili et al. (2003), the precipitation hardening behaviour of aluminum alloy AA6111 was quantified for both natural and artificial aging processes. GP zones were present in the early stages of isothermal artificial aging at 180°C, but peak strength was attributed to a combination of β'' precipitates and Q' precipitates. These observations are captured with bright field TEM images in Figure 4, in which clustering present after 1 hour of aging at 180°C disappears in favour of β'' and Q' precipitates after 7 hours of aging at 180°C. Di Ciano et al. (2018) explored the coarsening response of AA6013-T6 under warm forming conditions, an alloy with a higher copper content than the AA6111 studied by Esmaeili et al. (2003). They found that for temperatures below 300°C at times up to 330 seconds and strains of 8% that there was a negligible effect on the subsequent room temperature yield and ultimate tensile strengths.

Werinos et al. (2016) studied the impact of trace amounts of Sn (tin) on the artificial aging process of an aluminum-magnesium-silicon (Al-Mg-Si) alloy known as AA6061 when treated at high temperatures exceeding 210 °C under aging times of $10^0 - 10^4$ seconds. The research demonstrated that the addition of Sn resulted in accelerated aging kinetics and a higher peak hardness compared to the alloy without Sn. The positive effects were attributed to the formation of Sn-vacancy pairs, which contribute to faster diffusion, and the ability of Sn to hinder the elimination of vacancies during the aging process.

The 2000-series alloys are made heat-treatable with the addition of copper, which forms GP zones and eventually β'' precipitates within the microstructure for dislocation pinning. These alloys offer strengths that range from 190 to 430 MPa and are typically used in aerospace components. The addition of copper

within these alloys has been shown to increase susceptibility to stress corrosion cracking (SCC) (Kaufman, 2000).

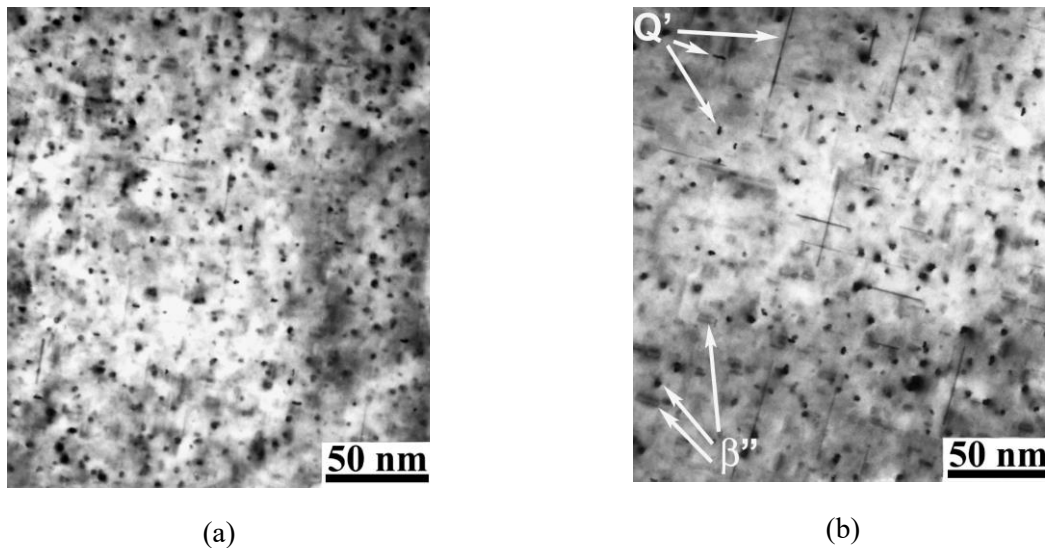


Figure 4: Bright field TEM micrographs of AA6111 being artificially aged at 180°C for 1 hour (a) and 7 hours (b) (Esmaeili et al., 2003).

The 6000-series alloys are primarily made heat-treatable through the addition of magnesium and silicon, although copper is also sometimes used. 6000-series alloys have strengths comparable to that of the 2000-series, but typically are more cost effective to manufacture (Kaufman, 2000). Applications within the automotive sector include outer panels, inner panels, reinforcement beams (Miller et al., 2000), and crush structures (Kohar et al., 2016). 6000-series alloys may have tensile strengths on the order of 300 MPa which would yield a strength-to-weight ratio of 111 MPa/(g/cm³) – in line with high strength dual-phase steels.

The 7000-series alloys are primarily made heat-treatable with the addition of zinc and copper. These alloys have been reported to have UTS values in excess of 600 MPa in the T6 temper, yielding a strength-to-weight ratio of 213 MPa/(g/cm³), comparable to 1500 MPa boron steel (191 MPa/(g/cm³)) (Smith, 1981; Arconic Aerospace and Automotive Products, 2016). Applications for these alloys are primarily aerospace driven; however, interest within the automotive sector has slowly increased to further improve vehicle fuel economy by vehicle light-weighting (Toros et al., 2008). 7000 series-alloys generally have a more complex alloying system than 6000-series alloys (Figure 3), promoting greater strengths in the former alloy series. 7000-series alloys are vulnerable to exfoliation corrosion and stress corrosion cracking (SCC) in the peak aged condition, but this behaviour is improved in the over-aged state (Li et al., 2009; Chen et al., 2012; Sun et al., 2013; Chen et al., 2014; Qi et al., 2016). The over-aged condition is typically designated with a T7x, where 'x' may provide additional information on the over-aged heat-treatment.

2.2. Forming Limit Diagram

The formability of sheet metals is described by the forming limit curve (FLC) or forming limit diagram (FLD). The forming limit curve corresponds to the line fit to a series of “safe” in-plane principal strain pairs (ϵ_1, ϵ_2) obtained under different loading conditions (Keeler & Backofen, 1963). Here, “safe” refers to the strain level below which the sheet material will localize or “neck” for a given strain path. A representation of some strain paths, such as uniaxial tension, plane strain, and biaxial loading, are shown in Figure 5.. The FLD is useful for describing sheet formability as a function of different processing route parameters, such as temperature and strain rate, in which case multiple FLCs will exist depending upon the different processing conditions under consideration. In its most common application, if the strains in a formed part in a numerical simulation lie below the “safe” FLC strains, the part and forming process may be considered ready for initial production.

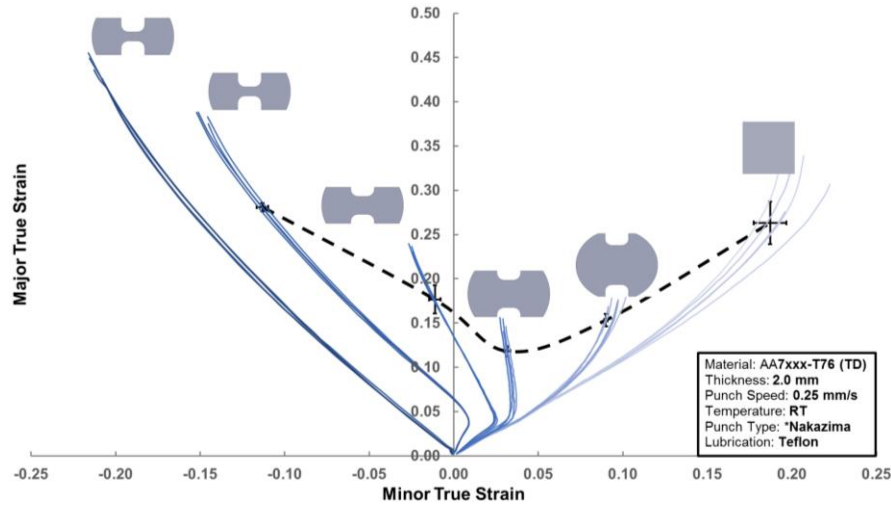


Figure 5: Illustration of a forming limit curve for different limiting dome height (LDH) specimens with corresponding strain paths using a Nakazima hemispherical punch.

A key limitation of the strain-based FLC is that it is only applicable to approximately linear strain-paths for in-plane stretching without bending; thus, the conventional FLC description may not accurately predict localization for the complex strain paths seen in metal forming operations. Stoughton et al. (2000; 2004) first proposed a solution to account for non-linear strain path effects using a principal stress-based representation of the FLC that was assumed to be independent of the load path. Stoughton demonstrates the initial derivation using a Hill-48 (1948) yield surface and shows compelling correlations with experimental FLCs derived under different pre-strain conditions. Seong et al. (2014) later showed that this approach can be used in incremental sheet metal forming operations to suppress necking through optimization of the strain paths on a part. To reduce the considerable time and expense required to generate an FLC, analytical

methods have been developed to predict the FLC such as the MK-technique (Marciniak & Kuczynski, 1967), as will be described in Section 2.2.2.

In the initial development of the FLC, a strain measurement technique known as circle-grid analysis was used to measure the strains on the specimens formed for the development of the FLC (Keeler & Backofen, 1963; Marciniak & Kuczynski, 1967; Danckert, 1979). This type of strain measurement technique consists of etching circle (or sometimes square) grids onto a test specimen and measuring the changes in dimensions of each individual circle after the test was complete. Figure 6 shows a specimen that uses a square grid during a hemispherical dome test. Grid-based analysis techniques are still used in practice for the measurement of strains on formed parts; however, for lab-scale specimens, digital image correlation (DIC) techniques are also used for strain measurement and analysis (Ozturk et al., 2009).

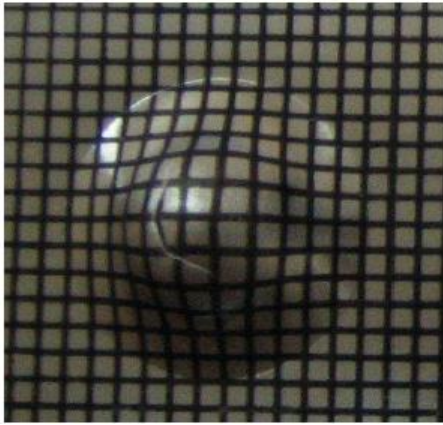


Figure 6: Grid-based strain measurement approach (Ozturk et al., 2009).

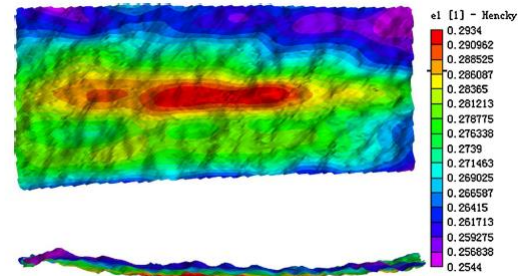


Figure 7: Major strain distribution on a dome specimen acquired using DIC techniques (Wang et al., 2014).

Digital image correlation (Jones & Iadicola, 2018) is a full-field image analysis method that can determine deformation fields on an object in two-dimensional (2D) and three-dimensional (3D) space (Sutton et al., 2009). DIC is a non-intrusive measurement technique that allows for tracking of the strains across a part over the duration of the testing, as long as the measured area has a clear line of sight over the duration of testing. Formability characterization and analysis has seen significant use of DIC techniques over the last decade (Volk & Hora, 2011; Hotz et al., 2013; Li et al., 2013). Figure 7 and Figure 8 show examples of the major strain distributions on formed limiting dome height (LDH) specimens acquired with 3D stereoscopic DIC techniques. The primary advantage of the DIC approach over grid-based analysis is that the entire strain history over the forming operation is available.

2.2.1. Experimental Techniques

The use of FLCs within industry is common practice, but the methods for developing experimental FLCs are quite varied. In the original work of Keeler and Backhofen (1963), coupons were incrementally formed for varying strain states. When a physical localization was observed, the “safe” principal strain pairs at that location were recorded. Here, a physical localization refers to a surface discontinuity (neck), as measured by touch. Historically, the use of circle-grid analysis led to significant variability in the FLC’s determined by different test labs; however, with the advent of DIC with local strain measurement, great strides have been made to standardize limit strain detection methods with the ISO12004-2:2008 standard (2008), herein referred to as the ISO standard.

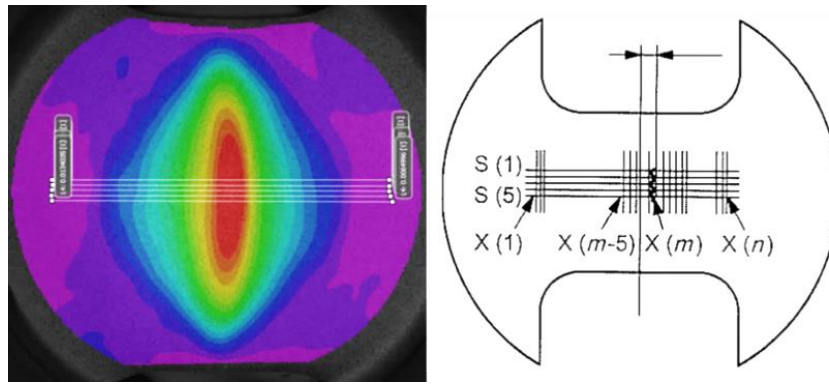


Figure 8: Major strain distribution of an AA6013-T6 plane strain dome before the onset of cracking (left) and the procedure used for extracting the line slice strain data (right) (DiCecco et al., 2016).

The limit strain detection algorithm proposed within ISO12004-2:2008 is an empirical approach based upon the strain distribution. In this approach, forming tests are performed using either a Nakazima hemispherical punch (1968) or a Marciniak (1967) type flat-punch on different sample geometries. After the test is complete, the strain distribution perpendicular to the crack location of a failed specimen is extracted, as shown in Figure 8. This distribution will herein be referred to as “line slice data” and can be obtained with either DIC analysis or circle grid analysis.

To determine the safe forming limit strains using the ISO approach, a parabolic fit to the strain distribution (both major and minor) obtained from the line slice data is first created by omitting the strain measurements directly adjacent to either side of the specimen crack, as shown in Figure 9. The major and minor limit strains are taken as the point on each respective parabolic strain distribution fit coinciding with the location of failure in the experiment; ideally, this occurs at the centre of the specimen. The standard was originally developed for grid-based strain determination techniques (offline); however, it is also applicable for in-situ strain measurement techniques using digital image correlation analysis.

The ISO12004-2:2008 standard played a key role in standardizing FLC generation; however, the approach is not without its limitations. The standard was originally developed for room temperature applications. Furthermore, some evidence suggests the analysis approach may be limited to symmetric strain distributions, i.e. strain distributions corresponding to specimen failure at the centre of the punch (Martinez-Donaire et al., 2014; Vysochinskiy et al., 2016) that are difficult to obtain for some materials due to dynamic strain aging and friction between the punch and blank. Consequently, alternative approaches for FLC determination are actively pursued in the literature, notably with the advent of DIC measurement techniques.

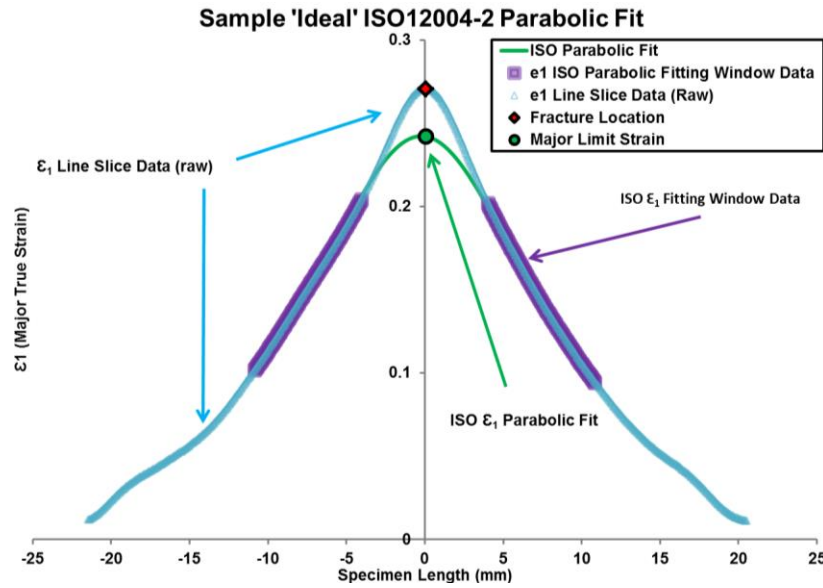


Figure 9: Sample ISO inverse parabolic fitting process (DiCecco et al., 2016). Origin is placed at the location of failure as determined from the experiment. Specimen width corresponds to distance measured along the surface of the specimen.

Volk and Hora (2011) developed a time-dependent method for detection of forming limits that is otherwise referred to as the ‘linear best fit time-dependent method’, or LBF for short. The proposed method is based on the temporal analysis of the thickness strain rate, $\dot{\epsilon}_3$, at a location within the failure region of the test specimen, as illustrated in Figure 10. Two straight lines are computed using regression analysis from the time derivative data. The first line is computed over the “stable” region of the time derivative, which is specified as a number of images from the beginning of the test. The second line is computed over the “unstable” region of the time derivative that corresponds to a number of images prior to fracture. The unstable region (onset of necking) coincides with acceleration of the strain-rate. The authors state that the intersection of the two computed lines coincides with the “safe time” (Figure 10) and the maximum strains from the DIC image just prior to this time are taken as the forming limit strains. The methodology was applied by Volk and Hora (2011) to two steels, HC220YD and C260LAD, and the formability was found

to match closely with measurements previously acquired using circle-grid analysis techniques. The primary benefits of the approach are the simplicity of the methodology and the corresponding ease of implementation. Considerations when implementing the approach are raised with respect to the sensitivity of the method to the number of images from the failure image when computing the time derivative, as well as the number of data points that should be used for regression analysis. The approach has been used in other papers at room temperature (Leotoing et al., 2013; Hotz et al., 2013) and elevated temperatures (Jain, 2016) and presents as a tool for computing formability that is fairly user-independent.

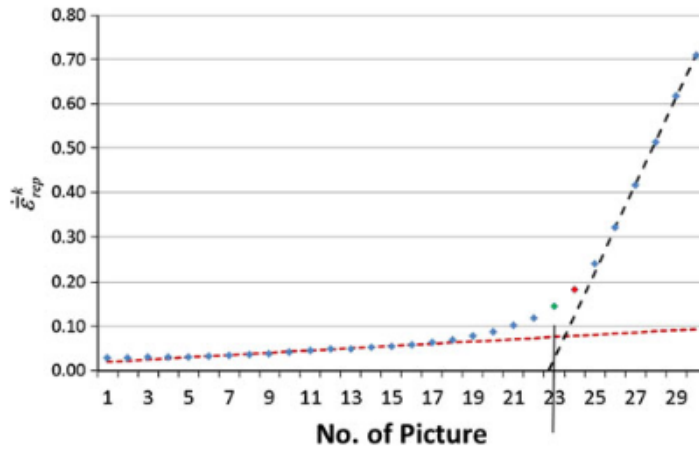


Figure 10: Linear best fit time-dependent FLC methodology (Volk & Hora, 2011). “No. of Picture” refers to a succession of images acquired during the formability experiment and vertical axis plots the negative of the local rate of thickness strain.

The “strain rate unloading” approach was proposed by Martínez-Donaire et al. (2014) (referred to herein as the MD method) in which they consider the time-evolution of the rate of principal strain, $\dot{\epsilon}_1$ (or $\dot{\epsilon}_3$), at locations of interest on the surface of the dome. These rates are calculated based on changes in strain between successive DIC images acquired during testing. The motivation for the MD method is that the local strain rate at the centre of the localization zone should continually increase until fracture. Outside of the localization region, the strain rates will begin to decrease once the material begins to unload as deformation concentrates within the neck. An illustration of this behaviour is shown in Figure 11. In the MD method, the detection of a decreasing strain rate at the boundary of the neck indicates that the forming limit has been reached. This technique was applied to AA7075-O and dual-phase steel alloys for different punch geometries at room temperature. The technique was shown to coincide with safe strains found using the ISO12004-2 standard.

The MD approach is attractive in that it is developed from physical observations of sheet material behaviour during localization and failure. Furthermore, the MD approach takes advantage of the time history available from the recorded DIC data (rather than just the final image prior to fracture as in the ISO

method). The MD approach also requires no additional processing within the digital image correlation software relative to the ISO approach since the same line slice data can be used for calculating strain rates ($\dot{\epsilon}_1$ or $\dot{\epsilon}_3$).

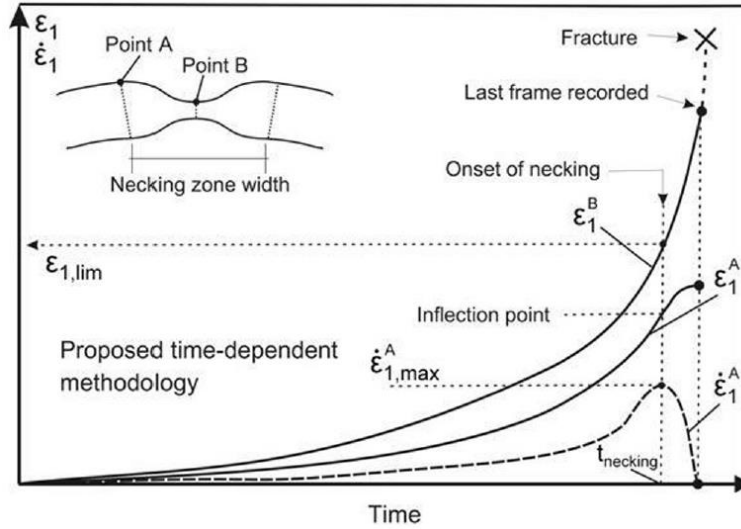


Figure 11: Illustration of MD localization approach (Martinez-Donaire et al., 2014). The ϵ_1^B curve corresponds to the strain within the localized region and the ϵ_1^A curve corresponds to the strain outside of the localized region. Note that the ϵ_1^A curve shows signs of strain rate unloading.

The curvature approach is a forming limit detection scheme in which local changes in the curvature of the sheet metal surface associated with onset of necking are detected from stereoscopic DIC measurements. This method is a focus of the current thesis document and has had different types of implementations within the literature. The curvature approach of Wang et al. (2014) tracked local changes in dome height as a function of time, as seen in Figure 12 and Figure 13. This approach was later expanded on in the work of Min et al. (2016) with a similar methodology proposed by Martinez-Donaire et al. (2014) that was referred to as the ‘flat valley approach’.

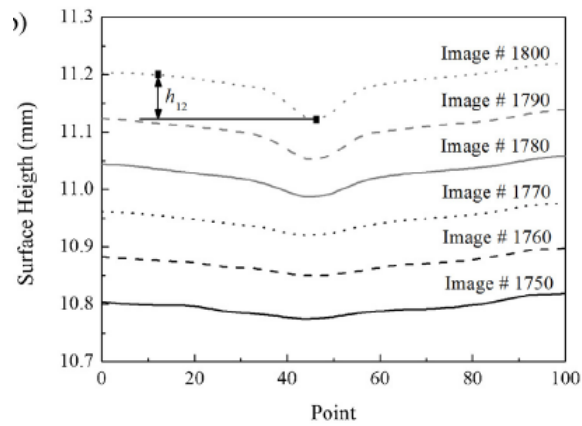


Figure 12: Dome height profile parallel to dome crack (Wang et al., 2014).

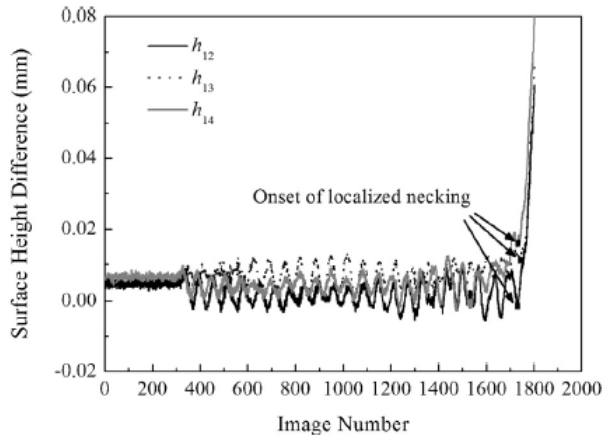
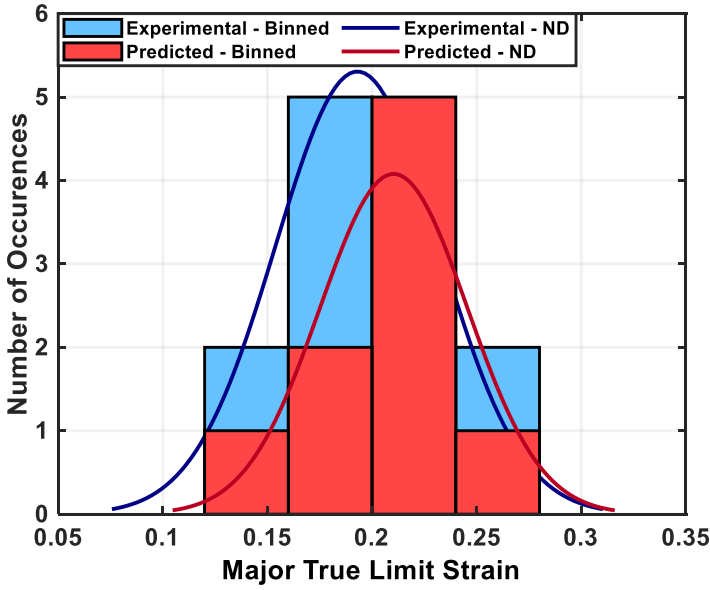


Figure 13: Local dome height differences versus recorded test images.

Despite the range of available limit strain detection techniques, well documented experimental procedures, and almost 50 years of experience in FLC generation within the literature, there are still significant discrepancies in FLC generation between labs. Figure 14 is a distribution of plane strain limit strains reported within the literature for production-grade sheet aluminum alloy AA5182-O. The distribution distinguishes between experimentally and numerically determined limit strains, and, although it is not comprehensive, it does make use of 22 data points from 15 unique sources. The limit strains have been binned in 0.05 true strain intervals and normal distributions have been calculated for both the experimental and numerical limit strains. It can be seen that there is significant disagreement within the literature on the plane strain limit strain value of AA5182-O, despite this alloy being a relatively common production grade sheet material. Furthermore, the variance in the distribution is higher for the experimental data than the numerical predictions; however, both datasets are of similar shape. Some differences in material properties are expected and could contribute to some of the observed variation. Other differences might be attributed to the presence of dynamic strain aging or Portevin-Le Chatelier (PLC) (McCormick, 1971; Robinson & Shaw, 1994; Wang et al., 2014; Tian et al., 2022) effects in AA5182-O which is visualized as serrated flow in the stress-strain response of the alloy. Serrated flow is common in 5000-series aluminum alloys as a result of the transient interaction between solute Mg atoms and dislocation motion.

The observed variation in formability of AA5182-O is meant to emphasize the need for better and more unified formability test methodologies and analysis procedures. Additionally, it should be noted that many numerical limit strain predictions are derived using experimental results; thus, if the latter shows lab-dependent variance, the former can potentially propagate that variance.



Year	Authors
1979	(Ayres & Wenner, 1979)
2003	(Wu et al., 2003)
2003	(Kim et al., 2003)
2004	(Miles et al., 2004)
2005	(Banabic et al., 2005)
2009	(Signorelli et al., 2009)
2010	(Li & Wu, 2010)
2010	(Soare, 2010)
2011	(Khan & Baig, 2011)
2011	(Serenelli et al., 2011)
2013	(Li et al., 2013)
2014	(Rohatgi et al., 2014)
2015	(Ju et al., 2015)
2016	(Maris et al., 2016)
2019	(Noder & Butcher, C., 2019)

Figure 14: Variations in the plane strain limit strains of AA5182-O, as reported within the open literature. Citations used to generate the figure are included to the right. ND for Normal Distribution. Experimental data corresponds to tested conditions and predicted data is developed from analytical and numerical techniques.

2.2.2. Predictive Techniques for FLC Development

The development of experimental forming limit diagrams is paramount in assessing material formability; however, the assessment of formability in a comprehensive way for each new sheet material is often time consuming and cost prohibitive. Numerical and analytical techniques do not have these limitations and can be useful for predicting sheet formability in a more parametric approach. These techniques allow for varying different material parameters that might influence sheet formability such as work hardening rate, anisotropy, friction, elastic modulus, temperature, and strain rate sensitivity (Roll, 2008); however, they are inherently dependent on accurate representation of the material behaviour, as determined from material characterization.

Analytical determination of sheet formability for different loading conditions was first explored in the work of Swift (1952) for diffuse necking. Swift derived an equation for the biaxial formability of a sheet metal using the Considère criterion (1885), $\frac{d\sigma}{d\varepsilon} = \sigma$, in conjunction with a power-law type strain-hardening law, $\sigma = k\varepsilon^n$, where σ is true stress, ε is the equivalent plastic strain, and k and n are hardening parameters identified from a tensile test. The Considère criterion can be directly interpreted to mean that diffuse necking under tensile loading occurs at the peak load. Swift's biaxial formability equation (1), listed below, for the major true strain, ε_1 , is a function of the yield surface, f . In the same year, Hill (1952) also explored

localization in sheet metals and discussed the influence between hardening, anisotropy, and the angle at which localization occurs under specific loading conditions.

$$\varepsilon_1^* = \frac{\sigma_1 \left(\frac{\partial f}{\partial \sigma_1} \right)^2 + \sigma_2 \left(\frac{\partial f}{\partial \sigma_2} \right) \left(\frac{\partial f}{\partial \sigma_1} \right)}{\sigma_1 \left(\frac{\partial f}{\partial \sigma_1} \right)^2 + \sigma_2 \left(\frac{\partial f}{\partial \sigma_2} \right)^2} n \quad (1)$$

Marciniak and Kucynski (1967) have approached prediction of sheet metal formability in biaxial stretching with the assumption that a defect is the cause for the initiation of necking. In this analysis technique, the defect is modelled as a variation in the sheet thickness perpendicular to the specimen length. Two regions of material are defined: (A) the region of uniform thickness and strain; and (B) the defect region where the strain will localize, as shown in Figure 15. Region (A) is loaded under arbitrary conditions, and assuming some predefined hardening function and yield surface; the stresses and strains in region (B) are computed through equilibrium and compatibility. When the ratio of major strains in Regions (B) and (A), $\varepsilon_1^{(B)} / \varepsilon_1^{(A)}$ exceed a prescribed value, often taken as 10, a neck is assumed to have initiated and the strains in region (A) are used for the FLC point. By varying the loading conditions on region (A), different points on the derived FLC can be determined. The initial sheet thickness imperfection should be determined through calibration against an experimental FLC data point. Hutchinson et al. (1978) improved upon this technique by solving the compatibility equations and extending the approach to the left-hand side of the FLC for uniaxial stretching conditions.

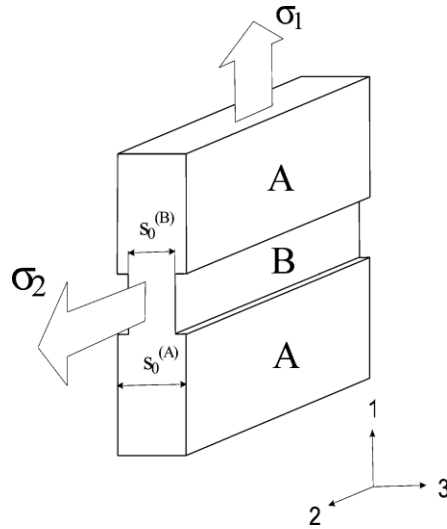


Figure 15: Geometrical model of the MK theory (Banabic, 2010).

As an alternative to MK analysis, the Modified Maximum Force Criterion (MMFC) was proposed by Hora et al. (1996; 2013) for the prediction of necking in sheet metals. The approach utilizes the Swift criterion (Swift, 1952) for onset of diffuse necking and then the strain path naturally evolves towards plane strain since this condition can sustain the maximum load in terms of the principal stress. When plane strain conditions are reached, acute localization occurs and the limit strain is reached. The advantage of this method is that no calibration parameters are assumed, such as a defect or imperfection size as in the MK model. Localization is a function of the yield surface and hardening model and very good agreement has been obtained with experimental FLCs and with MK analysis. The change in the principal strain ratio, defined as β in the figure, and its evolution towards plane strain is illustrated in Figure 16 for an arbitrary yield surface.

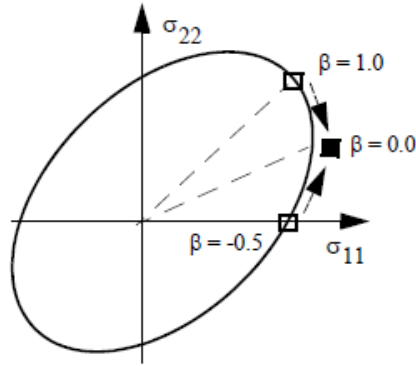


Figure 16: Transition from arbitrary loading conditions to plane strain loading at the onset of necking (Hora et al., 1996).

Formability prediction techniques, like those utilized in the MMFC (Hora et al., 2013) and expanded upon by Noder and Butch (2019), rely on carefully chosen constitutive models and accurate material characterization for accurate predictions. Experimental methods for determining formability, like those described by Fast-Irvine et al. (2021), must account for the influence of process parameters such as tool contact stresses, as highlighted by Noder and Butcher (2023), to be reliable and effective. Direct comparison of these prediction techniques with experimental methods, particularly under complex loading conditions, e.g. combined bending and stretching, is essential to validate and refine formability prediction models.

2.3. Warm Formability of Aluminum Alloys

Warm forming of aluminum alloys is intended to improve formability while also reducing springback (Kim & Koc, 2008; Greze et al., 2010). Studies on warm forming can be found that date back to the mid-1970s. Shehata et al. (1978) explored the influence of magnesium on pure aluminum with respect to strain-rate sensitivity and found that increasing magnesium content increased the elevated temperature strain-rate

sensitivity of the alloy and thus the at-temperature formability. In the work of Ayres and Wenner (1979), the strain and strain-rate hardening effects in punch stretching of AA5182-O aluminum were explored at elevated temperatures. The serrated flow behaviour, also known as dynamic strain aging, of AA5182-O at room temperature was found to decrease with elevated temperatures. Additionally, a larger total elongation in the post-uniform region was observed. These increases were attributed to a shift in the hardening mechanism from work hardening at room temperature to strain-rate hardening at elevated temperatures. Similar observations were made for different alloys over the next 10 years, which were summarized in two review papers by Rao and Doraivelu (1982) and Wilson (1988).

Research on warm forming technology was prevalent in the 2000s, as discussed in the work of Takata et al. (2000) and the publications by Li and Ghosh (2003; 2004; 2004). Li and Ghosh, found that the AA5182 alloy has a good response to warm forming, as observed through visible inspection, total tensile specimen elongation, and dome height observations. Forming limit bands that capture the formability and error in the corresponding measurements are generated in the work of Li and Gosh (Figure 17); however, it is stated that at the time of testing, no standardized formability characterization methodology for warm forming was available and the actual methodology that was implemented is not fully clear. Nevertheless, their work was useful in providing additional details with respect to strain rate effects. Li and Ghosh further extended their testing to AA5754-O and AA6111-T4 in the 200°C - 350°C testing regime. Here, it was found that the AA5754-O alloy showed a similar response to the AA5182 alloy; however, the AA6111-T4 alloy did not show significant improvement in dome height or total elongation with respect to warm forming. It is stated that AA6111-T4 may not be a suitable warm forming candidate. Finally, it was shown that in deep-draw tests, the best draw ratios could be achieved when the punch is set to a lower temperature than the dies, thus indicating that a non-isothermal deep-draw test is necessary for superior deep-drawability. Bagheriasl and Worswick (2015) also observed that non-isothermal conditions promote superior draw-ratios for an AA3003 brazing sheet material under warm forming conditions.

More recent studies on AA5000 series alloys have confirmed the benefits of warm forming for these materials and that these benefits are primarily derived from elevated temperature strain rate sensitivity. In the work of Abedrabbo et al. (2007), the forming limit behaviour of AA5182-O and AA5754-O was determined numerically using an MK analysis coupled with a discrete temperature-dependent flow rule. The formability trends (Figure 18) were shown to match those of previous studies; however, the actual values were found to differ. The mechanism for formability improvement is linked to the increased strain-rate sensitivity at elevated temperature, as shown in Figure 19. In the work of Chu et al. (2014), the formability of AA5086 was investigated using MK analysis with experimentally determined hardening behaviour. It was shown that the formability benefits of AA5086 may decrease with increasing punch

speeds. The same trends are observed experimentally and the strain-rate dependent FLD is computed using the ISO12004 methodology.

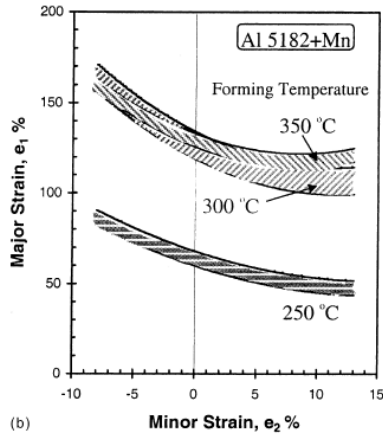


Figure 17: AA5182-O formability bands (Li & Ghosh, 2004).

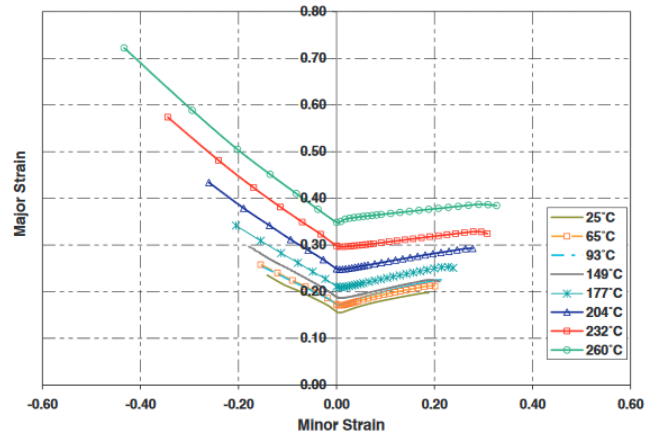


Figure 18: AA5182-O FLD developed using MK analysis (Abedrabbo et al., 2007).

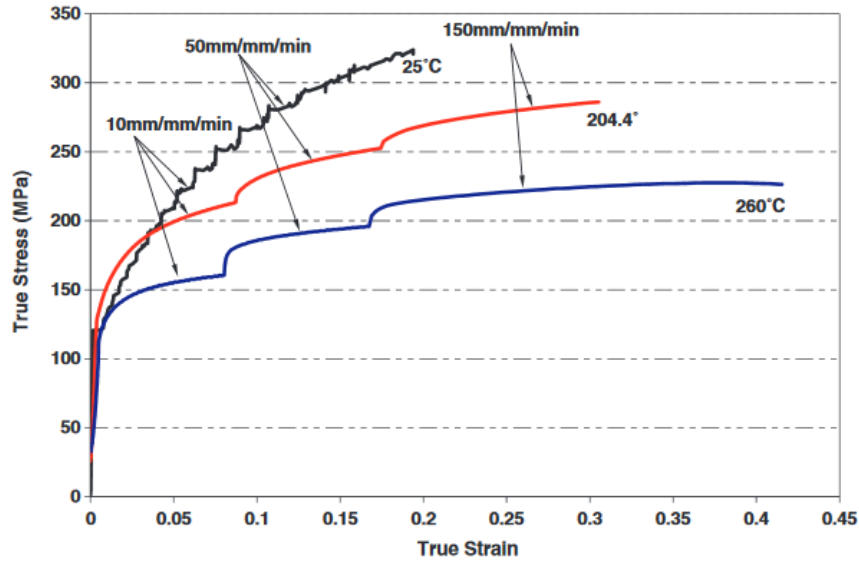


Figure 19: AA5182-O strain-rate sensitivity using 'jump' tensile tests (Abedrabbo et al., 2007).

The interest in warm forming of aluminum alloys has continued to grow with a significant amount of research on the warm formability of non-heat-treatable alloys (Mahabunphachai & Koc, 2010; Bagheriasl et al., 2010; Bagheriasl & Worswick, 2015; Ju et al., 2015). Generally, forming trends remain the same in this class of alloys, in which increased temperatures result in improved formability; however, the absolute values between studies for the same alloys are not always consistent, which accounts for the continued publications on the warm formability of AA5182 over 40 years after the initial work presented in the 1970s (Min et al., 2015; Ju et al., 2015). Two potential reasons for the continued need for research in this area are

as follows: (1) no single standard exists which can be universally applied towards the characterization of warm formability, as observed when reviewing different AA5000 publications and (2) the exact mechanisms for which warm-forming should promote improved formability are not fully understood (Abedrabbo, 2005; Abedrabbo et al., 2007). On the macroscopic level, it is clear from tensile test observations (Ayres & Wenner, 1979; Ghosh & Li, 2003) that the decrease in hardening as a function of temperature is compensated for by an increase in strain-rate hardening (Figure 19); however, it is unclear if this compensation will always be sufficient, or if the one factor may outpace another. Furthermore, the shift towards strain-rate hardening has implications for the type of localization behaviour that might be observed during warm deformation of various aluminum alloys.

2.3.1. Elevated Temperature Forming of Age Hardenable Aluminum Alloys

Of late, there has been an increase in the number of publications with respect to warm forming of higher strength, age-hardenable aluminum alloys in an effort to further reduce vehicle weight (Wang et al., 2012; Harrison & Lucky, 2014) and a limited number of studies on die quenching (Garrett et al., 2005; Kumar & Ross, 2017; Long et al., 2017; Omer et al., 2018). Unlike non-heat-treatable aluminum alloys, the final mechanical properties of heat-treatable aluminum alloys, such as the 6000- and 7000- series alloys, are sensitive to over-aging caused by elevated forming temperatures and long times during pre-heating and warm forming. With increasing forming temperature these alloys experience reductions in the as-processed mechanical properties; however, the exact warm forming conditions to achieve a balance between formability and final properties are largely tied to alloy composition and temper and requires further study. Because of this inherent sensitivity to time at temperature, optimization of the formability window for age-hardenable alloys is much more complex than for non-heat-treatable alloys.

Early work on warm forming of AA6111-T4 by Li and Ghosh (2004) did not result in significant gains in formability. Possible reasons for this lack of benefit may lie in the precipitate (aging) response to temperature that is not seen in 5000-series alloys. The increase in strength due to exposure to relatively high temperatures (200°C to 350°C) can also have had a negative effect on formability. The room temperature formability of AA6013 and other 6000-series alloys is described in the work of Prillhofer et al. (2014). In this work, they show that there is only minor variation in room temperature formability between aluminum sheet alloys AA6013, AA6063, AA6016 and AA6005 for an under-aged temper. Further, they demonstrate that the difference in formability after natural aging for periods of 2-6 months was minimal for all tested alloys. There are other studies for 6000-series alloys with respect to room temperature formability; however, very little data outside of the work of Li and Ghosh (2004) is available to provide guidance on the possible formability benefits for the 6000-series of aluminum alloys in the warm forming

regime. Similarly, little formability work has been conducted with respect to AA6013 within the open literature, outside of publications by DiCecco et al. (2016; 2017) as part of the current research program.

Previous work on warm forming of high strength age-hardenable alloys includes the work of Sotirov et al. (2012), who examined the warm formability response of AA7075-T6 at room temperature, 200°C and 230°C. Kumar et al. (2014) demonstrate that the limiting draw ratio of AW7020-T6 benefits from warm forming up to 250°C, but with significant degradation in the as-formed yield and ultimate tensile strengths after 200°C. They found that formability continuously increases over this range of temperatures. No discussion of the methodology used for determining the FLD is presented, making the results difficult to fully interpret. In the work of Hui et al. (2012), the warm forming response of aluminum alloy AA7075-T6 is indirectly assessed through dome-height measurements, as shown in Figure 20. It can be seen in this figure that the AA7075-T6 alloy may exhibit optimal forming performance at around 200°C. In this work, biaxial specimens were utilized, but it is unclear if the dome heights are measured at fracture or before localization. Furthermore, the time at temperature is not specified, which would be important to know due to the possibility of over-aging at higher temperature forming conditions. Nevertheless, it is stated that when the alloy is exposed to temperatures at or in excess of 260°C, a significant loss of subsequent room temperature strength and elongation occurs, indicative of precipitate coarsening (over-aging) within the AA7075 microstructure.

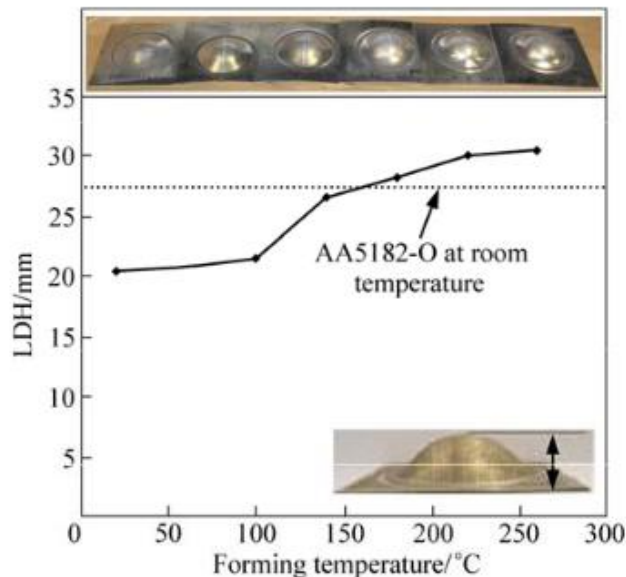


Figure 20: Dome height versus temperature for AA7075-T6 (Wang et al., 2012).

Ivanoff et al. (2015) studied the influence of warm forming and the paint-bake cycle on an AA7075-T6 alloy. It was found that if an AA7075-T6 specimen was rapidly heated to simulate warm forming and

cooled, the specimen would lose strength (hardness) above 200°C due to precipitates entering solution. If the specimen was subsequently cycled through a paint-bake cycle, the strength of the sample would increase through aging. This process is referred to as retrogression and re-aging, as is shown in Figure 21.

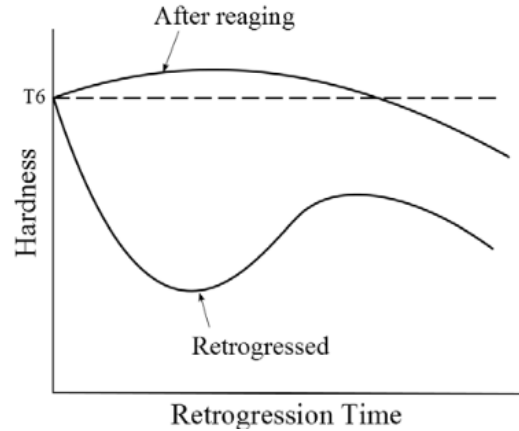


Figure 21: Influence of regression time on retrogressed hardness and re-aged hardness after a paint-bake cycle for AA7075-T6 (Ivanoff et al., 2015).

In the work of Long et al. (2017) the baseline chemical composition of an AA7055 sheet alloy was modified for improved warm forming and corrosion behaviour. Their results show that improved corrosion and forming response is possible with changes to the chemical composition; however, exact alloying element modifications were not provided. Kumar and Ross (2016) investigated the effect of initial temper on the warm forming response of AW7921 sheet alloy. They found that formability was improved up to 230°C and was sensitive to initial temper. The precipitate distribution was found to change from GP zones only at 10 min of aging at 100 °C to GP zones and η' precipitates after subsequent aging at 120 °C for 50 min. Overall, AW7921 in the T4 temper was found to yield the best combination of warm formability and final strength (after exposure to warm forming and paint-bake cycle).

Raja and Kumar (2019) studied an AA6061 alloy at different punch speeds and temperatures, reporting warm forming limit strains on the order of 100%; however, in their study, dome heights were unchanged across all forming conditions, showing a discrepancy between reported limit strains and dome heights.

Hua et al. (2021) studied an Al-Zn-Mg-Cu alloy in a pre-aged condition and formed in a warm deep drawing process. The technique involves pre-aging the alloy prior to warm forming, resulting in a final product with high strength and improved formability. The study found that warm drawability of the pre-aged alloy was improved over that of an O-temper alloy at 200°C using an Erichsen testing apparatus. Chen et al. (2024) explore a similar approach for AA7075 alloy sheet and found that the pre-aged condition can improve warm formability and the warm forming process can still achieve strength within approximately 90% of the T6 condition; however, the study only reports dome heights at failure and not limit strains.

Pishyar (2021) also investigated the warm forming characteristics of the AA7075 aluminum alloy, focusing on the effects of rapid heating and pre-aging of the alloy. The study found that faster heating rates led to slightly enhanced forming limits by reducing the extent of pre-aging during heating, shown in Figure 22 (Pishyar, 2021). Although formability was the primary focus, the research also monitored the effects on mechanical strength and observed that a paint-bake cycle increased the hardness of warm formed specimens. Notably, the T6 temper, which experienced a slight hardness reduction after warm forming, recovered its hardness after the paint-bake cycle, likely due to retrogression and re-aging effects.

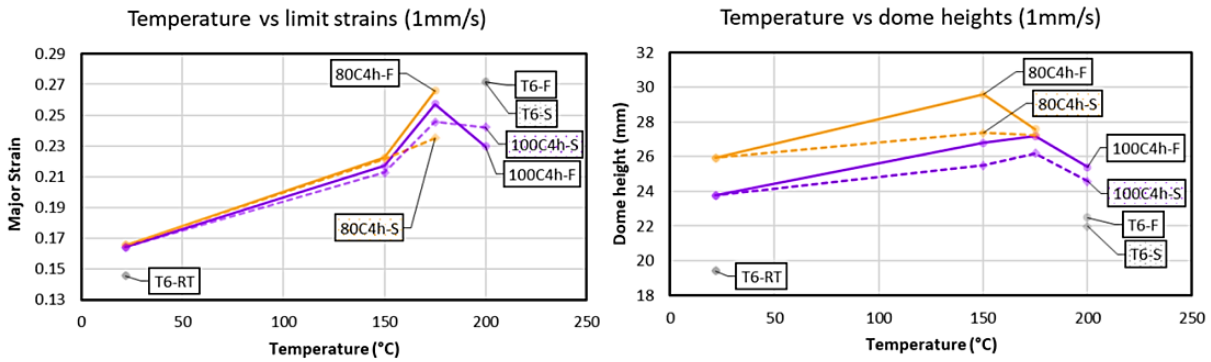


Figure 22: AA7075 major limit strain and dome height measurements are plotted for a nominal Nakazima punch speed of 1 mm/s as a function of forming temperature for pre-aged tempers (80°C, 4 hours; 100°C, 4 hours) and a T6 temper. *Slow* and *fast* heating rates during warm forming are studied and denoted by -S and -F, respectively. Limit strains (left) are observed to increase for all conditions up to 175°C forming temperatures; however, dome heights (right) plateau between 150 - 175°C based on the condition being studied (Pishyar, 2021).

2.3.2. Die Quenching of Age-Hardenable Aluminum Alloys

In instances where high formability is desired, the die quenching (DQ)/hot-stamping process for age-hardenable aluminum alloys has been explored. The process involves heating a blank to a solutionizing temperature and then quenching and forming in cold tooling to prevent precipitation, leaving the part in a supersaturated solid solution (SSSS) (Garrett et al., 2005; Kumar & Ross, 2017; Long et al., 2017; Omer, 2019). Omer et al. (2018) studied the effects of different DQ process parameters, such as solutionizing time, transfer time, and quench rate, on the age hardenability of AA7075 and a developmental 7000-series alloy (AA7xxx). They determined that 8 minutes was a sufficient solutionizing time to achieve the desired level of age hardening for both alloys. They also found that it is possible to reach mechanical properties similar to those of conventionally-aged T6 and T76 tempers using a paint-bake cycle as the final aging step.

Omer et al. (2020) also investigated the constitutive properties of AA7075 and AA7xxx subjected to hot forming. They designed a quenching apparatus to replicate the temperature-time cycle of the DQ process during tensile testing. A technique called the area reduction method (ARM) was used to obtain the stress-strain curves. This method uses stereoscopic digital image correlation (DIC) to approximate the

cross-sectional area of the diffuse neck. They used their findings to calibrate a Barlat Yld2000 yield surface to capture the anisotropy of the alloys, which did not significantly evolve as a function of temperature. Boulis et al. (2019) investigated the effects of the DQ process on AA7050. They found that increasing the solution heat treatment temperature led to increased strength after aging. They also observed an orange-peel defect on AA7050-T6 specimens formed at 150°C.

Other researchers have found that the blank holding force (BHF) can significantly affect the formability of parts during hot stamping. Liu et al. (2021) showed that a lower BHF led to smaller cracks, or even no cracks at all, compared to a higher BHF. They concluded that the increased friction from a high BHF restricted the flow of material in the flange, resulting in the stretching and cracking of the softer material in the sidewall. They also found that a 20-minute solution heat treatment was sufficient to achieve a minimum strength of about 96% of the as-received T6 condition, which exceeds that reported by Omer et al. (2018). The maximum strength achieved slightly exceeded that of the as-received T6 condition. They also found that a short-time aging process of approximately 110°C for 1 hour followed by 180°C for 30 minutes resulted in a strength within 90% of the as-received T6 condition.

2.4. Modelling of Elevated Temperature Forming with Aluminum Alloys

Considerable research has been accomplished concerning simulation of the warm formability of 5000-series aluminum alloys. In the work of Takuda et al. (2002), the cylindrical cup deep-drawing of AA5182-O at warm forming temperatures was numerically investigated and compared with experimental observations. In the experimental work, the work hardening value was found to progressively decrease with temperature, while the normal anisotropy parameter, r , was found to be almost invariant with respect to temperature. When the deep draw models and experiments were compared for 250°C, the maximum thinning locations were found to be in good agreement; however, there is no discussion of how strain-rate sensitivity was addressed in the model and no strain-rate sensitivity measurements were provided.

Abedrabbo et al. (2007) also examined the warm forming response of AA5182-O. In this work, the strain-rate sensitivity is experimentally obtained and used within the developed models. A Barlat YLD-2000-2d (Barlat et al., 2003) yield criterion was implemented to capture the anisotropy of the AA5182 alloy, with anisotropy coefficients that are shown to evolve as a function of temperature using a fourth order polynomial equation, although the trends in anisotropy evolution were minimal. A power-law hardening equation was utilized to capture material flow with parameters that were fit using higher order piecewise polynomials. An MK analysis was performed to numerically determine the warm FLD of the alloy. No experimental FLDs were included for comparison.

Kurukuri et al. (2010) modelled the warm deep-drawing response of AA6061-T6 using a physically-based “Nes” rate-sensitive hardening model and Vegter yield criterion. A phenomenological model and the Bergstrom model were also investigated prior to using the Nes model. A change in earing profile in experimental testing was observed for different temperatures, which was attributed to a change in material anisotropy as a function of temperature. Friction and tool orientation were not found to contribute to this response. In the work of Prillhofer et al. (2014), the AA6013 normal r-value was found to be 0.62 with r-values for the rolling, diagonal, and transverse directions (RD, DD, TD) of the alloy being relatively close to the normal, which may indicate that a simpler yield surface could be implemented to sufficiently capture the material anisotropy. In the thesis of Noder (2017) significant anisotropy was shown for an AA7000 series alloy, which required a Barlat YLD2000-2d (Barlat et al., 2003) yield surface to capture. A stress-saturation type model was used to model material hardening at temperature.

In the work of Sotirov et al. (2012), the warm formability of an AA7075-T6 alloy was modelled for different isothermal forming conditions. A power-law type hardening rule was implemented to capture the flow and strain-rate behaviour at elevated temperatures. A Barlat YLD-89 (Barlat & Lian, 1989) model was utilized to capture the apparent material anisotropy. The predicted and measured final cup heights were found to be similar. The improved formability at elevated temperatures was attributed to recovery and dislocation annihilation at temperature.

The anisotropy of AA7075 and AA7085 sheet aluminum alloys are discussed in the works of Leacock et al. (2013) and Bhattacharyya et al. (2017), respectively. Leacock et al. (2013) show that the AA7075 anisotropy is constant for different natural aging times, with significant variation in r-values as a function of angle from the rolling direction; however, the O-temper anisotropy differs substantially from the naturally aged anisotropy, as shown in Figure 23. The implication of this result may be that so long as the precipitates remain stable, the material anisotropy should not change significantly. For AA7085, the T711 temper shows strong variation in r-values for different angles measured from the rolling direction, as shown in Figure 24.

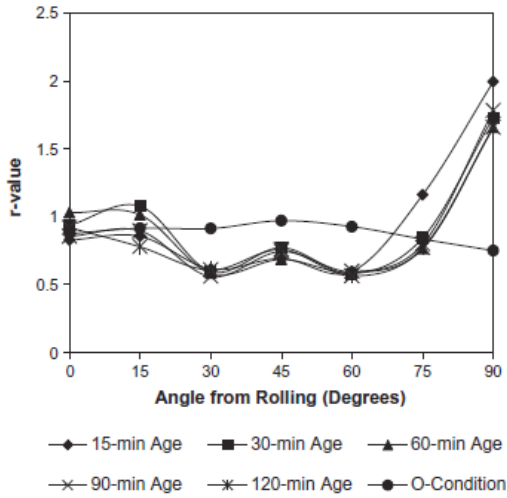


Figure 23: AA7075 r-value variation (Leacock et al., 2013).

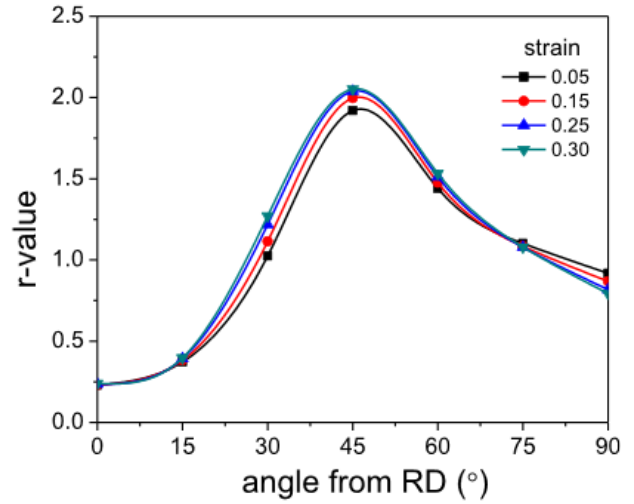


Figure 24: AA7085-T711 r-value variation (Bhattacharyya et al., 2017).

2.5. Summary of Previous Research

The foregoing review of the literature concerning warm forming has revealed that while considerable work has addressed the warm formability of 3000- and 5000-series alloys, only limited work has been done on age-hardenable 6000- and 7000-series alloys.

The basic mechanism that enables a positive warm forming response for non-heat-treatable aluminum alloys consists of an increase in strain-rate sensitivity with temperature that compensates for the drop in material hardening at elevated temperature. For age-hardenable alloys, this topic has only recently begun to be investigated in greater detail; notably with respect to the effect of time at temperature on precipitate evolution and the corresponding effect on at-temperature formability and as-processed mechanical properties. The warm formability of AA7075-T6 has received some attention in the published literature, with study of pre-aging effects; however, in some such studies, details on the formability methodologies are not well documented. Conversely, no information is currently available regarding the warm formability response of AA6013 and AA7xxx, the alloys of interest in the current thesis. On the elevated temperature constitutive response, the AA7xxx alloy has been studied at elevated temperatures in a solutionized condition (Omer et al., 2020).

The lack of comprehensive data on the elevated temperature formability of age-hardenable aluminum alloys must be addressed to support further use of these materials within the transportation sector. Notably, the effect of the initial temper on the room-temperature and elevated temperature forming response, as well as the final strength properties following the warm-forming thermal cycle and paint-bake are all points of interest required to implement precipitation-hardenable alloys in automotive applications. Furthermore, an

approach to experimentally characterize the warm-formability of aluminum alloys must be developed in response to current deficiencies within the literature regarding formability characterization. Predictive models of the alloy kinetics on the constitutive behaviour would greatly benefit the precipitation-hardenable alloy use in temperature-dependent applications, which is currently not well documented. Finally, on the topic of die-quench formability assessment, little work has been done to experimentally characterize the die-quench formability of aluminum alloys, primarily due to limitations in experimental techniques to assess formability under die-quench conditions at temperatures in excess of 400°C under isothermal and non-isothermal conditions.

3. OBJECTIVES

To address these shortfalls in the published literature, the goal of this thesis research is to quantify the formability of two precipitation-hardenable aluminum alloys under elevated temperature forming conditions to enable further use of aluminum alloys within the automotive sector for light-weighting applications. To support this goal, a new technique to characterize sheet metal formability has been developed, termed the Enhance Curvature Method (ECM). The ECM and other techniques available in the scientific literature have been applied to characterize the formability of AA6013 and a developmental 7000-series aluminum alloy, herein referred to as AA7xxx, as functions of both temperature and initial temper. AA6013 was selected for study as it offers relatively high strength within the 6000-series alloy class in a peak aged temper. The AA7xxx developmental alloy has a reduced quench sensitivity relative to more common 7000-series alloys, such as AA7075, making it a particularly suitable candidate for hot stamping.

The following objectives were established for this research and are addressed within this thesis:

1. Develop an objective methodology for formability assessment of sheet aluminum alloys suitable for elevated temperature forming conditions;
2. Characterize the warm formability of AA6013-T6 through the development of forming limit diagrams applying the methodology developed in Objective 1;
3. Quantify the warm formability of AA6013 as a function of temperature and temper using the methodology of Objective 1 and apply a calibrated aging-kinetics model of a pre-aged temper to suggest optimal forming process parameters balancing formability and final strength properties;
4. Experimentally characterize the warm formability of AA7xxx as a function of temperature and starting temperature using the methodology of Objective 1 and quantify the effects of the thermal processing history on the strength and ductility response of the alloy;
5. Assess the formability of AA7xxx under isothermal and non-isothermal die quenching conditions to understand the mechanisms associated with the die quenching process and corresponding limitations.

The remainder of this thesis serves to describe work done in support of the aforementioned objectives, and comprises experimental procedures, results, discussion, conclusions, and recommendations for future work in this area of research.

4. EXPERIMENTAL METHODOLOGY, RESULTS, AND DISCUSSION

This section presents the experiments developed as part of this research and their application to characterize the formability of several aluminum sheet alloys under room and elevated temperature conditions. Section 4.1 details the ECM developed as part of this work for formability characterization which is based on the detection of the onset of necking using changes in surface curvature during deformation. This method is applied to room temperature forming of an Al-Mg alloy sheet, AA5182-O. In Section 4.2, the ECM and other formability detection methods are applied to warm forming of a precipitation hardening 6000-series alloy, AA6013-T6. The effect of aging during warm forming, as well as the initial starting temper, on formability of this alloy is examined in Section 4.3. In Section 4.4, the warm formability of a 7000-series alloy, AA7xxx, in various initial tempers, is characterized using the ECM. The potential benefits of hot forming of this alloy using a die quenching processing route are explored in Section 4.5.

This thesis has been written in a manuscript format in which this synopsis provides an overview of the most relevant methods and results, while several appendices accompany the synopsis and provide greater detail on each aspect of the research. Each appendix comprises a journal or conference publication. The five appendices are:

- A. Formability Characterization Using Curvature and Strain-Rate Based Limit Strain Detection Methods Applied to Marciniak, Nakazima, and Stretch-Bend Tests (DiCecco et al., 2024)
- B. Numerical and experimental investigation of the formability of AA6013-T6 (DiCecco et al., 2017)
- C. Warm Aging of Pre-aged AA6013 Sheet and its Relevance to Room Temperature and Warm Forming Applications – Experimental and Modeling Analyses (DiCecco et al., 2023)
- D. Effect of Initial Temper on the Warm Forming Characteristics of a High Strength 7000-series Al-Zn-Mg-Cu Alloy (DiCecco et al., 2022)
- E. Limit Strain Characterization in an Aluminum Die Quenching Process (DiCecco et al., 2018)

Note that all five articles directly associated with this work are published and available under open access.

4.1. Formability Characterization Using Curvature and Strain-Rate Based Limit Strain Detection Methods Applied to Marciniak, Nakazima, and Stretch-Bend Tests

A methodology entitled the Enhanced Curvature Method (ECM) was developed to determine safe limit strains in sheet metals under arbitrary loading conditions for the development of forming limit curves. The ECM detects the physical development of a neck on the sheet surface by evaluating the sheet topology during deformation, leveraging stereoscopic digital image correlation techniques. The approach is applied to 1.1 mm thick AA5182-O sheet under Marciniak, Nakazima, and Stretch-Bend test conditions and benchmarked against two other limit strain detection schemes. A key aspect of this work is the development of a threshold/necking tolerance metric for the ECM in an effort to reduce the degree of subjectivity in neck detection. Results using the ECM are compared to other published formability data for AA5182-O found in the literature, as well as a theoretically calculated plane strain limit strain value. The ECM algorithm and relevant FLC results are presented here, as well as some constitutive material properties, with the complete documentation of this work in Appendix A. The ECM is used in elevated temperature limit strain characterization throughout the remainder of this thesis.

4.1.1. Tensile Properties

A sheet aluminum-magnesium alloy, AA5182-O, 1.1 mm, was studied, with a nominal composition of 4.5 weight percent magnesium and 0.35 weight percent manganese with balance being aluminum. The engineering stress-strain curves from all tensile tests (ASTM International, 2016) are presented in Figure 25a) in which the presence of the dynamic strain aging or Portevin-Le Chatelier (PLC) effect (McCormick, 1971; Robinson & Shaw, 1994; Tian et al., 2022) is readily observed as serrations in the stress-strain response. Serrated-flow is common for the AA5182-O alloy, as observed in the work of Wang et al. (2014). The major true strain rate contours of RD tensile tests are presented in Figure 25b) over an applied tensile strain range of 0.025 to 0.149 in which PLC effects are evident. Average Lankford coefficients (R-values) in RD, DD, and TD were 0.67, 0.70, and 0.71, which corresponds to approximately normal plastic anisotropy ($\bar{r} = 0.7$).

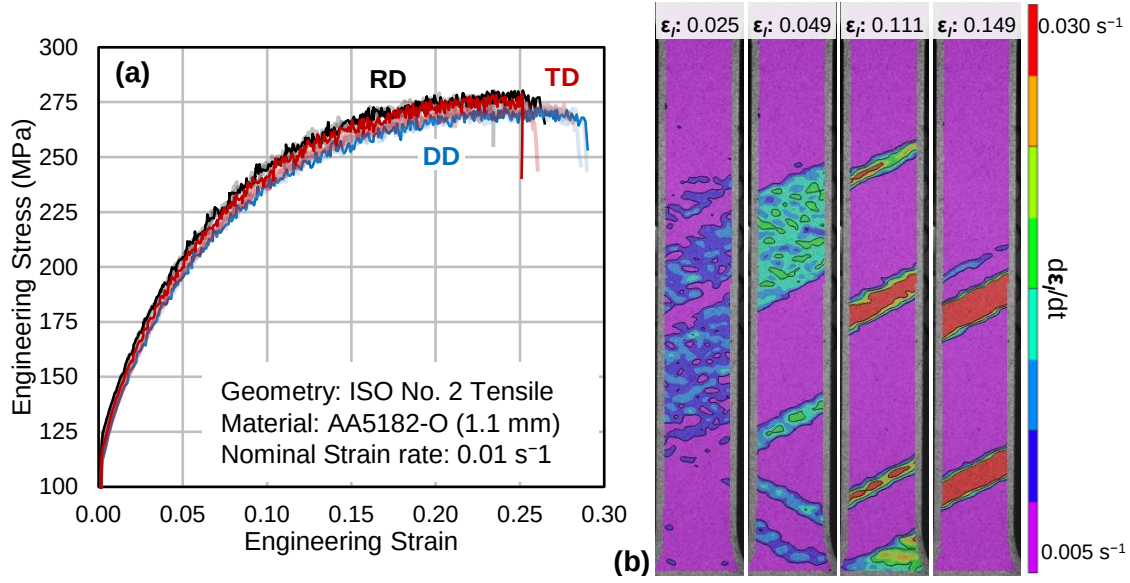


Figure 25: a) Repeated engineering stress-strain curves of AA5182-O (1.1 mm) in the rolling (RD), diagonal (DD) and transverse (TD) sheet directions. Major true strain rate contours from the gauge region of an RD tensile sample are presented in b) for applied tensile strains varying from 0.025 to 0.149 true strain. The strain rate contours highlight the PLC effect in the material and result in the serrated flow observed in a). At an applied tensile strain of 0.111, PLC bands are observed in the tensile shoulder, outside of the uniform gauge region.

The average uniform elongation in the rolling direction was determined according to the Considère criterion (equation (2)) and was found to be 23.7% (engineering strain), which corresponded to 0.208 plastic strain. For a plane stress assumption, the diffuse applied necking strain for the limiting formability condition of plane strain tension (FLC0) was determined as 0.194 using the Constrained Logan-Hosford (Hosford, 2009) yield function from Fast-Irvine et al. (2021) (equation (3)) in conjunction with the RD uniaxial flow curves and equation (2). An exponent of $m = 8$ for FCC materials and $\bar{r} = 0.7$ was employed.

$$kH - \sigma_{flow} = 0, H = \frac{d\sigma_{eq}}{d\varepsilon_p} \quad (2)$$

$$k = \left(\frac{\sigma_1}{\sigma_{eq,0}} \right)^{PST} = \left(\frac{1 + R_0}{1 + 2R_0(1/2)^m} \right)^{1/m} \quad (3)$$

4.1.2. Forming Limit Test Procedures

Formability was characterized using Marciniak and Nakazima tests, respectively. The effect of through thickness strain gradients was also investigated using cylindrical stretch-bend punches. All formability testing was conducted using dog-bone (DB) type specimens in accordance with the ISO12004-2:2008 standard, with a 25.4 mm long gauge section and a 20 mm transition radius, and various widths, as shown

in Figure 26 and summarized in Table 1. Three repeated tests were completed per test configuration under low friction conditions using three to four 0.125 mm sheets of polytetrafluoroethylene (Dupont, 1999) lightly coated in white petroleum jelly. The full details of die and punch geometries and test parameters can be found in Appendix A.

Table 1: Specimen widths used for Marciniak, Nakazima and Stretch-Bend testing. Widths used for FLC0 data collection in Nakazima and Marciniak tests are noted by the bolding in green.

Test	Specimen Width (mm)							
	25.4	50.8	76.2	101.6	114.3	127.0	152.4	177.8
Nakazima	x	x	x	x	x	x	x	x
Marciniak	x	x	x	x	x	x	x	x
Stretch-Bend			x					

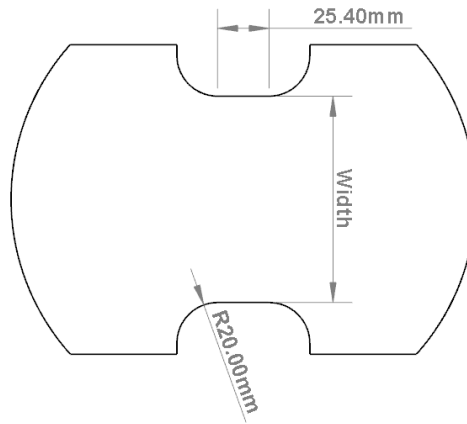


Figure 26: Specimen geometry used in all formability testing. Widths are given in Table 1.

Two four-megapixel Gazelle Camera Link cameras were utilized for stereoscopic image acquisition in all forming limit tests, providing an image scale of approximately 25 pixels/mm at a minimum frame rate of 10 fps. The software used for DIC analysis was Vic-3d 8® by Correlated Solutions Inc. Relevant image processing parameters in the software package include the subset size, strain filter size, and step size, all measured in pixels. The product of the step size, filter size, and image resolution is often termed the virtual strain gage length virtual strain gage length (VSGL) (Reu, 2015, Rossi et al., 2015). In this work, the respective step and strain filter sizes were 4 and 5 pixels, with a subset of 29 to 35 pixels. These parameters yielded an approximate VSGL of 0.8 mm. A brief study on VSGL sensitivity was conducted and it was determined that the limit strain results under plane strain loading were relatively insensitive to the VSGL below ~1.1 mm. Similar DIC software processing parameters were used in all experimental testing unless otherwise noted. Further details are available in Appendix A.

It is important to note that the term *thinning strain* is calculated using the measured strains on the exposed (convex) surface of the sample, using a volume conservation assumption, $\varepsilon_3 = -(\varepsilon_1 + \varepsilon_2)$, and neglecting elastic strains. A positive sign convention for ε_3 is adopted for convenience.

Displacement and strain data were extracted using so-called ‘line inspector tools’ or *line slices* in Vic-3d 8®, an example of which is shown in Figure 27a). Constrained smoothing splines (Reinsch, 1967) were used to smooth the strain and strain rate histories for the time-dependent and curvature-based forming limit detection methods to address PLC effects and facilitate derivative calculations, with an example of this approach shown in Figure 27b). The constrained splines (Reinsch, 1967) were fit using the open-source toolbox *SLM*, of D’Errico (2009) in Matlab®, with further discussion included in Appendix A.

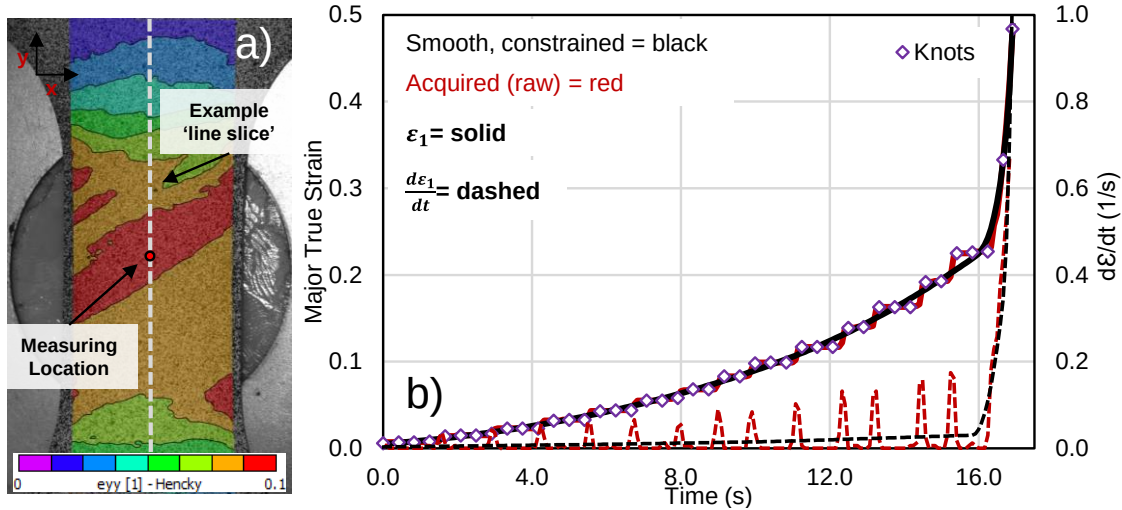


Figure 27: a) Lengthwise true strain contour of a 25.4 mm AA5182-O Marciniak dog-bone sample demonstrating the typical line slice location. In b), symbols are measured major strain (from “Measuring location” in a) versus time, solid line is a fit to the measured data using a smooth and constrained spline, while dashed curves are time derivatives of the measured (red) and smoothed (black) major true strain using a central finite difference scheme. The periodic spikes in the strain rate data in b) are attributed to PLC effects.

4.1.3. Limit Strain Detection: The Enhanced Curvature Method

In the curvature approach to limit strain detection, stereoscopic DIC is applied to optically detect the development and growth of a localized neck in terms of a surface discontinuity that results in a local change in surface curvature. This method is analogous to the so-called “finger test” of Keeler (1965) who defined a neck as being present only if it could be felt by tracing one’s fingernail across the surface of the blank. Figure 28a)-c) shows a conceptual evolution of the necking profile on the surface of a Marciniak sample and the detection of localization by changes in curvature on the sheet. Prior to necking, the curvature is relatively constant and close to zero, corresponding to uniform stretching of a nominally flat sheet. At the onset of localized necking the curvature signal rapidly changes as a physical groove, *i.e.* acute neck, develops. The limit strains correspond to the sudden change in curvature relative to the baseline signal.

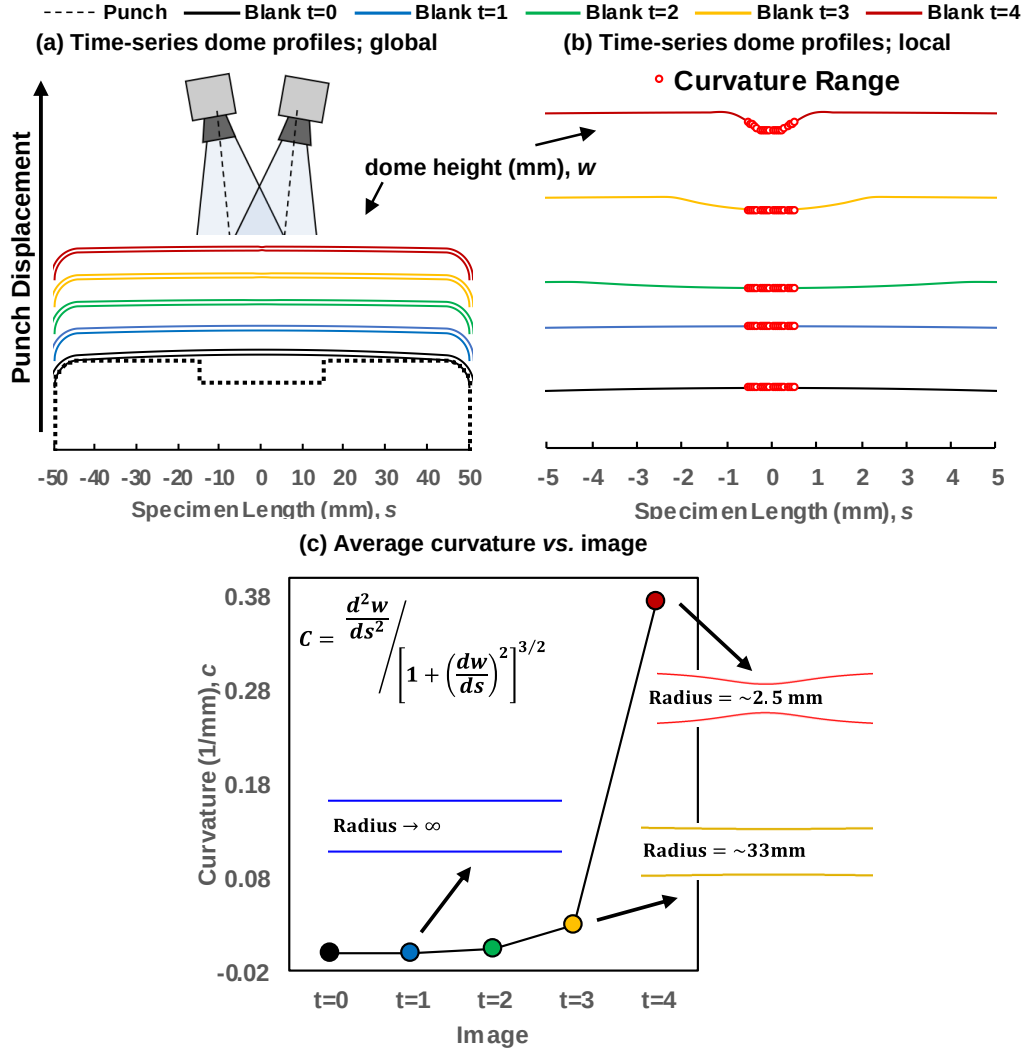


Figure 28: Schematic to illustrate idealized development of a neck in a sheet metal specimen and detection of the neck based on local changes in the sheet curvature. a) Includes a macro-view of the punch and sheet profile while b) shows the exaggerated close up profile of the top surface of the shear near the punch centre. c) Reflects the local curvature evolution for each time step with sample necked profiles.

The detection of necking by tracking curvature changes in the sheet is an extension of the original curvature methodology developed by DiCecco et al. (2016) and is thus termed the Enhanced Curvature Method (ECM). The ECM incorporates improvements in (i) the calculation of curvature and subsequent signal processing and (ii) thresholds for neck detection in terms of a thickness imperfection factor. In the curvature calculation and signal processing, spline formulations and a regression scheme are applied to differentiate signal from noise, as shown in Figures 29 and 30, respectively, for an AA5182-O 127.0 mm Marciniak plane-strain test. A quintic spline is applied to the DIC surface profile to compute the curvature using equation (4), where w is the displacement profile of the blank, and s is the length of the sample over the area of interest, with $s = 0$ mm centered about the failure location. The corresponding time-curvature

signal is processed using constrained smoothing splines to facilitate necking-detection as a function of signal threshold. The full details of the ECM implementation are discussed at length in Appendix A.

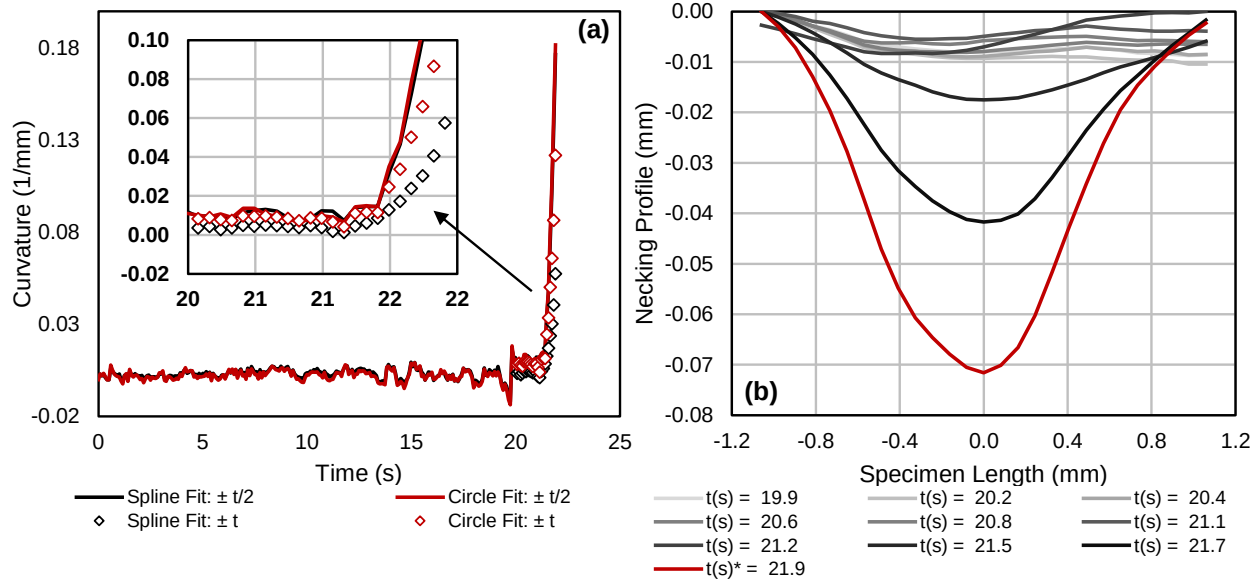


Figure 29: a) Comparison of calculated curvature evolution in a plane strain Marciniak test sample, as calculated using either a quintic smoothing spline or using a circle fit over ranges of 1.1 mm and 2.2 mm. The necking profiles at different instances in time at and after the onset of necking are included in b), with the red curve corresponding to the profile from the image just prior to fracture.

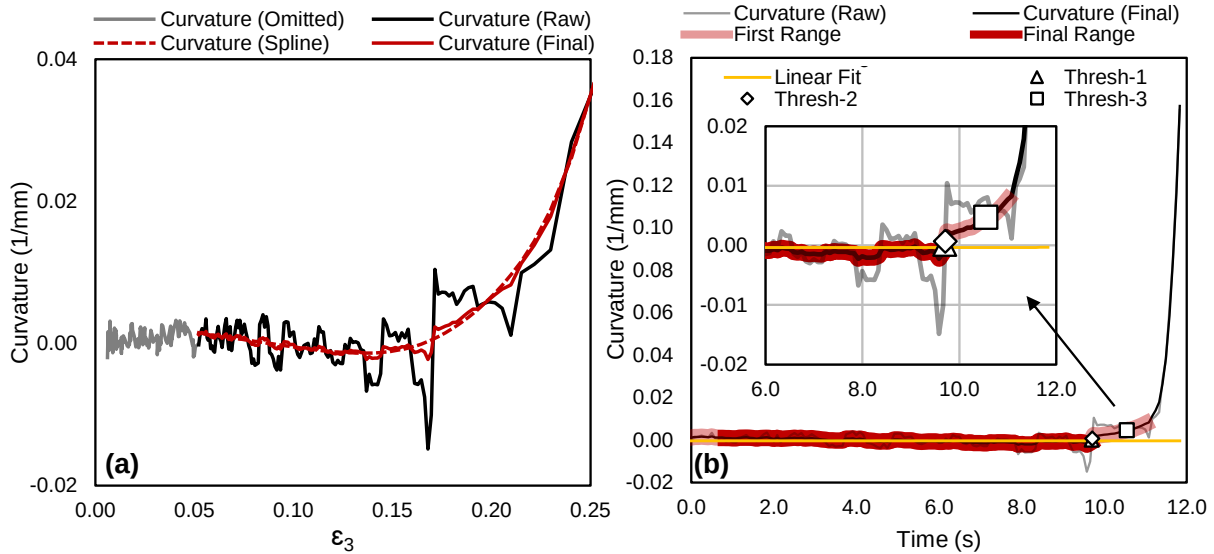


Figure 30: a) The measured curvature (raw) signal versus thinning strain (ϵ_3), portion of curvature data omitted, spline fit (C_s) using a constrained smoothing spline, and weighted average of the two curves is taken to produce the final curvature signal (C_f) for necking analysis. b) Shows the same data with the vertical axis range reduced for clarity. In b), deviation from the linear region in curvature vs. time is used to identify necking for different threshold values.

$$C = \frac{1}{n} \sum_{i=1}^n \left(\frac{d^2 w_i}{ds_i^2} \right) \left(1 + \left(\frac{dw_i}{ds_i} \right)^2 \right)^{-\frac{3}{2}} \quad (4)$$

A primary development of the ECM is the implementation of necking sensitivity thresholds which probe the perceived evolution of a neck – at what point a neck/imperfection might be observable by touch, e.g. As the curvature signal deviates from the baseline curvature, represented by ΔC_{f-lin} , necking progresses and is compared to different curvature thresholds. In the application of the ECM to AA5182-O (1.1 mm), three curvature thresholds of 0.0005 mm^{-1} (lower bound, Thresh-1), 0.001 mm^{-1} (Thresh-2), and upper bound of 0.005 mm^{-1} (Thresh-3) were initially considered as shown in Figure 30. These thresholds were applied to the plane strain Marciniak and Nakazima geometries over the 101.6 mm to 127.0 mm sample range. Corresponding limit strains are included in Figure 31. The 0.005 mm^{-1} threshold yielded average Marciniak limit strains of 0.198, which compared well to the theoretical limit of 0.194 determined using equation (2). This value also agrees with an average of reported plane strain limits strains of 0.193 (0.038) obtained from a review of the literature survey outlined in Appendix A. Thus, this threshold was adopted for subsequent FLC development using the ECM. A lower adopted threshold would decrease the resulting ECM FLCs. This might be leveraged in the future based on further study; however, for this work the threshold was selected to more closely align with the analytical FLC_0 solution.

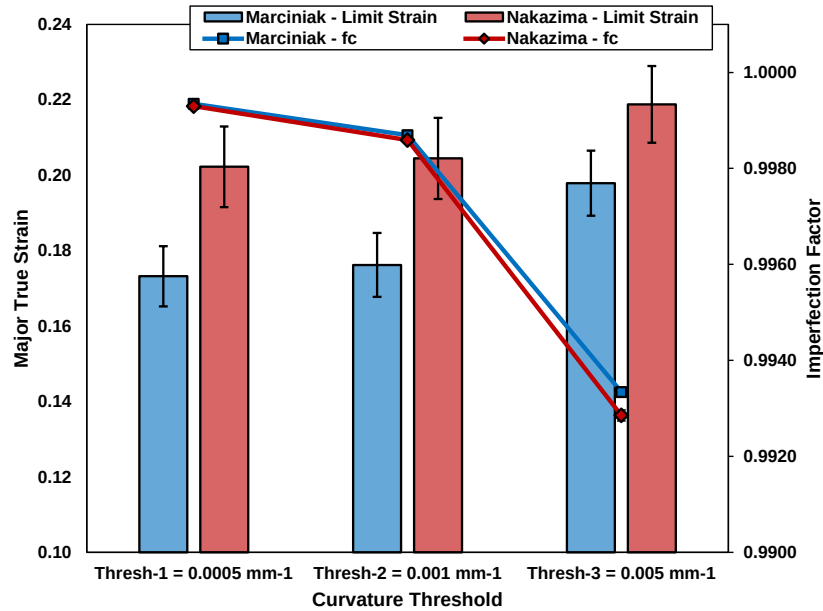


Figure 31: Average curvature limit strains and corresponding imperfection factors under approximate plane strain loading using samples with widths varying from 101.6 mm to 127.0 mm for Marciniak and Nakazima punches. Note that the scatter in the ECM limit strains (95% confidence interval) is relatively consistent across all thresholds.

An imperfection metric, f_c , is proposed as a more intuitive interpretation of ΔC_{f-lin} and the curvature thresholds. The imperfection metric corresponds to the ratio of the sheet thickness in the neck region to the sheet thickness outside of the neck region, with a visual representation included in Appendix A. Under certain circumstances, the calculation of f_c may be represented by equation (5), where R may be considered the reciprocal of ΔC_{f-lin} and t is the sheet thickness. The proposed imperfection metric holds the same form as that used in MK forming limit predictions (Marciniak & Kuczynski, 1967); however, in the current context, it acts as a tolerance on sheet thickness change for necking detection, as opposed to characterizing an initial thickness defect that triggers necking (in MK analysis).

$$f_c = 1 - 2 \left(\frac{R}{t} \right) \left(1 - \sqrt{1 - \left(\frac{t}{R} \right)^2} \right) \quad (5)$$

The evolution of f_c as a function of major strain is examined in Figure 32a-b) for the Marciniak and Nakazima plane strain tests, results of which are also included in Figure 31. Three of nine test curves are included as well as an average curve and corresponding 95% confidence intervals for both Marciniak and Nakazima tests. A large band in the confidence intervals exists, suggesting significant variation in the evolution of f_c with each test that is possibly attributed to PLC effects and material variations. To account for variance in onset of acute necking, a correction was applied to the individual f_c curves in which the curves were shifted by the difference between the mean true strain and the individual true strains at $f_c = 0.9933$, *i.e.* the average f_c value in Marciniak testing for the 0.005 mm^{-1} curvature threshold. The shifted curves are plotted in Figure 32c-d), with the horizontal axis strains arbitrarily zeroed at 0.15 relative to Figure 32a-b) to highlight the strain evolution in the vicinity of acute necking. It can be seen that following the mean-curve shift, the corresponding error bands are significantly diminished. This result suggests that the variability in the onset of necking is real, and, if accounted for, the geometric evolution of the necking region once necking initiates is quite similar between tests, although PLC effects might still cause some additional variations. This study is repeated in Section 4.2 on AA6013-T6 with no PLC effects where the same trends are more readily observed.

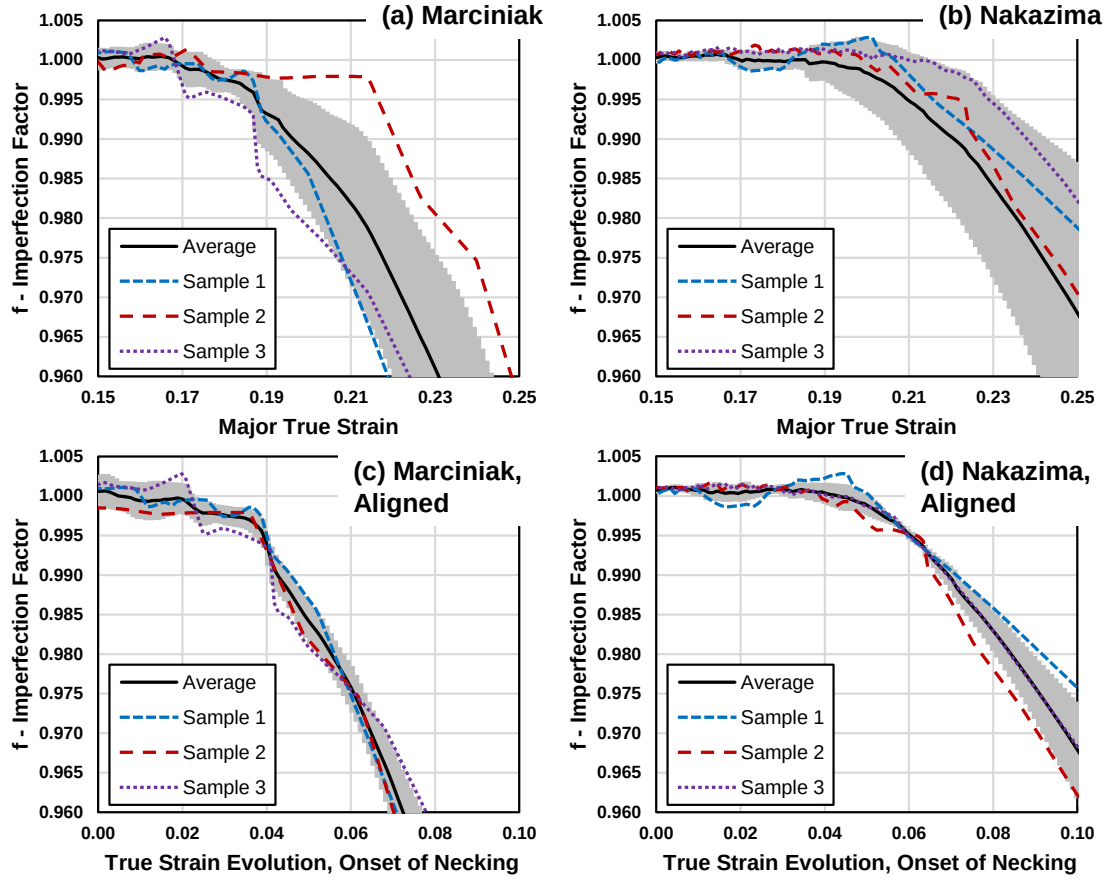


Figure 32: Plot of imperfection factor as a function of applied strain. Results from three (of nine) repeat tests are plotted for each case, along with the average response (black curve) and 95% confidence intervals (shaded) for a) AA5182-O tested with a Marciniak punch for plane strain samples (101.6 mm – 127.0 mm), b) AA5182-O tested with a Nakazima punch for plane strain samples (101.6 mm – 127.0 mm). In c) and d), the individual and mean f_c curves have been corrected to account for variability at the onset of necking by shifting all curves in the strain domain such that the individual curves overlap with the mean curves at strains corresponding to $f_c = 0.9933$, the 0.005 mm^{-1} curvature threshold. The resulting curves are plotted (zeroed) as onset of necking strains.

4.1.4. Formability Results using the Enhanced Curvature Method

Formability results from the ECM applied to AA5182-O are presented in this section for the Marciniak, Nakazima, and tight-radius stretch-bend test configurations. Comparisons of the ECM results are made with those obtained using the ISO12004-2:2008 standard and the Linear Best-Fit method of Volk & Hora (2011), with detail on the implementations of the two latter approaches given in Appendix A.

Marciniak limit strains identified using the ECM, LBF, and ISO approaches are plotted in Figure 33 alongside the average strain paths for each blank geometry. Generally, there was good strain path repeatability for all geometries. The averages of the limit strains for each strain path (specimen geometry) are summarized in Table 2. Under plane strain loading with sample widths from 101.6 to 127 mm, the limit strains are all within 0.02, with ECM, ISO, and LBF values of 0.198, 0.188, and 0.208, respectively. The

average values for the FLC0 using each methodology and the respective variation of about 0.01 are roughly in accordance with the literature survey for AA5182; however, the 0.208 value of the LBF exceeds the theoretical upper bound of 0.194 by more than the reported error bands. The opposite trend held for uniaxial tests, where the LBF method was generally more conservative, but also showed high variation. In the biaxial regime, the ECM limit strains were intermediate to those of the ISO and LBF values. The LBF approach was challenging to apply to the biaxial Marciniak geometries for which the extent of necking was relatively limited prior to onset of fracture.

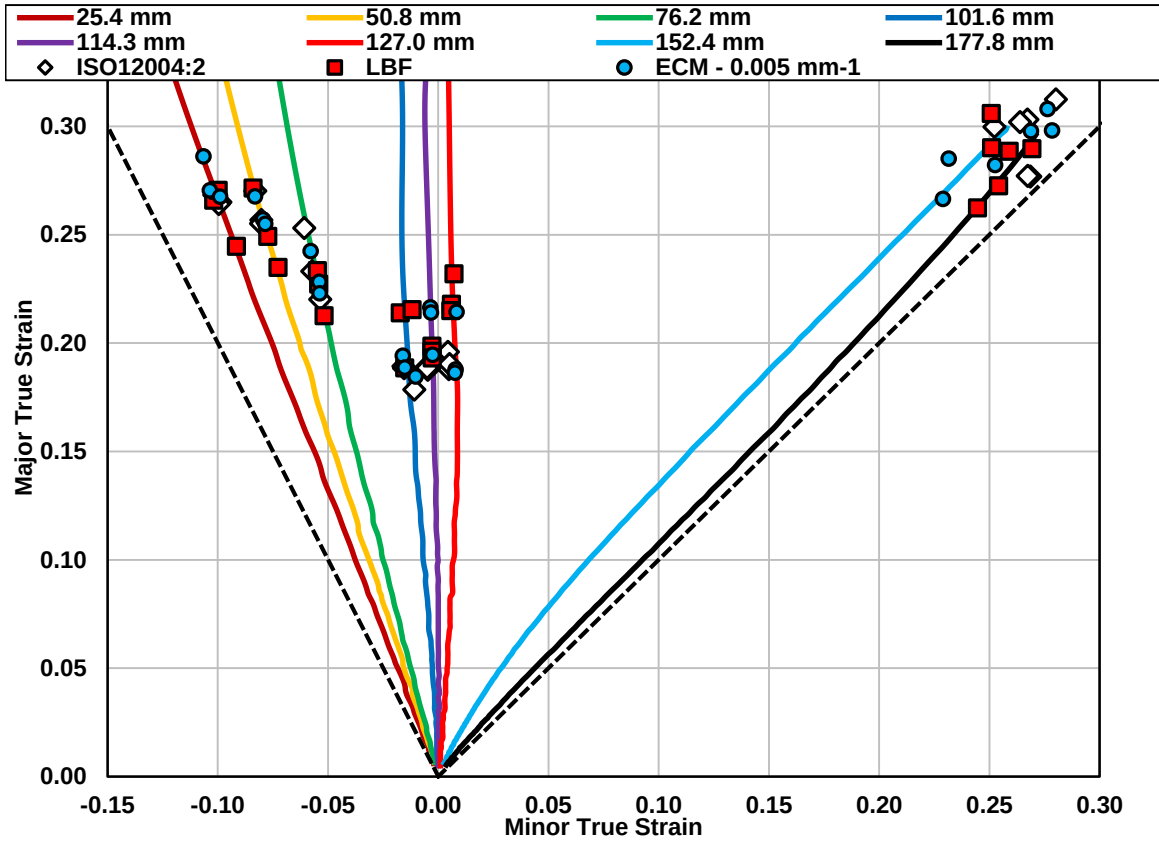


Figure 33: Individual limit strain points for each average strain path (specimen width) under Marciniak testing, as computed using the ISO12004-2:2008 standard, the LBF method of Volk and Hora (2011), and the upper threshold of the Enhanced Curvature Method.

Imperfection metric values corresponding to a curvature of 0.0005 mm^{-1} under Marciniak testing were all in the range of 0.9933-0.9935. f_c evolves based upon thinning which is largest under biaxial loading, thus the corresponding value of f_c is lowest for a given ECM threshold.

Table 2: Average limit strains from Marciniak testing as a function of specimen width determined with the ISO12004-2 standard, the LBF method, and the Enhanced Curvature Method. 95% confidence intervals are included in brackets to represent the error in each measurement.

Geometry	ISO12004-2:2008		LBF		ECM - 0.005 mm ⁻¹			ECM - 0.0005 mm ⁻¹		
	ϵ_2	ϵ_1	ϵ_2	ϵ_1	ϵ_2	ϵ_1	f_c	ϵ_2	ϵ_1	f_c
25.4 mm	-0.1 [0]	0.266 [0]	-0.098 [0.01]	0.26 [0.02]	-0.103 [0]	0.275 [0.01]	0.9935 [0]	-0.093 [0]	0.248 [0]	0.9994 [0]
50.8 mm	-0.081 [0]	0.261 [0.01]	-0.078 [0.01]	0.252 [0.02]	-0.08 [0]	0.26 [0.01]	0.9934 [0]	-0.077 [0]	0.247 [0.01]	0.9993 [0]
76.2 mm	-0.057 [0]	0.235 [0.02]	-0.054 [0]	0.224 [0.01]	-0.055 [0]	0.231 [0.01]	0.9934 [0]	-0.051 [0]	0.214 [0.02]	0.9994 [0]
101.6 mm	-0.014 [0]	0.185 [0.01]	-0.015 [0]	0.206 [0.02]	-0.014 [0]	0.189 [0.01]	0.9934 [0]	-0.013 [0]	0.176 [0.01]	0.9994 [0]
114.3 mm	-0.005 [0]	0.188 [0]	-0.003 [0]	0.196 [0]	-0.003 [0]	0.208 [0.01]	0.9933 [0]	-0.002 [0]	0.171 [0.02]	0.9994 [0]
127.0 mm	0.005 [0]	0.191 [0]	0.006 [0]	0.222 [0.01]	0.008 [0]	0.196 [0.02]	0.9933 [0]	0.009 [0]	0.172 [0.01]	0.9993 [0]
152.4 mm	0.261 [0.01]	0.302 [0]	0.254 [0.01]	0.295 [0.01]	0.238 [0.01]	0.278 [0.01]	0.9908 [0]	0.2 [0.01]	0.239 [0.01]	0.9992 [0]
177.8 mm	0.272 [0.01]	0.289 [0.02]	0.256 [0.01]	0.275 [0.02]	0.275 [0.01]	0.301 [0.01]	0.9928 [0]	0.248 [0]	0.267 [0.01]	0.9991 [0]

Limit strains from Nakazima testing are plotted in Figure 34 with corresponding average strain paths. The FLC values are also tabulated in Table 3 with corresponding f_c values. For near plane strain geometries with widths ranging from 101.6 mm to 127.0 mm, the major limit strains were 0.219, 0.204, and 0.222 for the ECM, ISO, and LBF methods, respectively. The ECM imperfection factors are similar to the Marciniak tests for these geometries with $f_c = 0.9933$ for Marciniak and $f_c = 0.9928$ for Nakazima tests, on average.

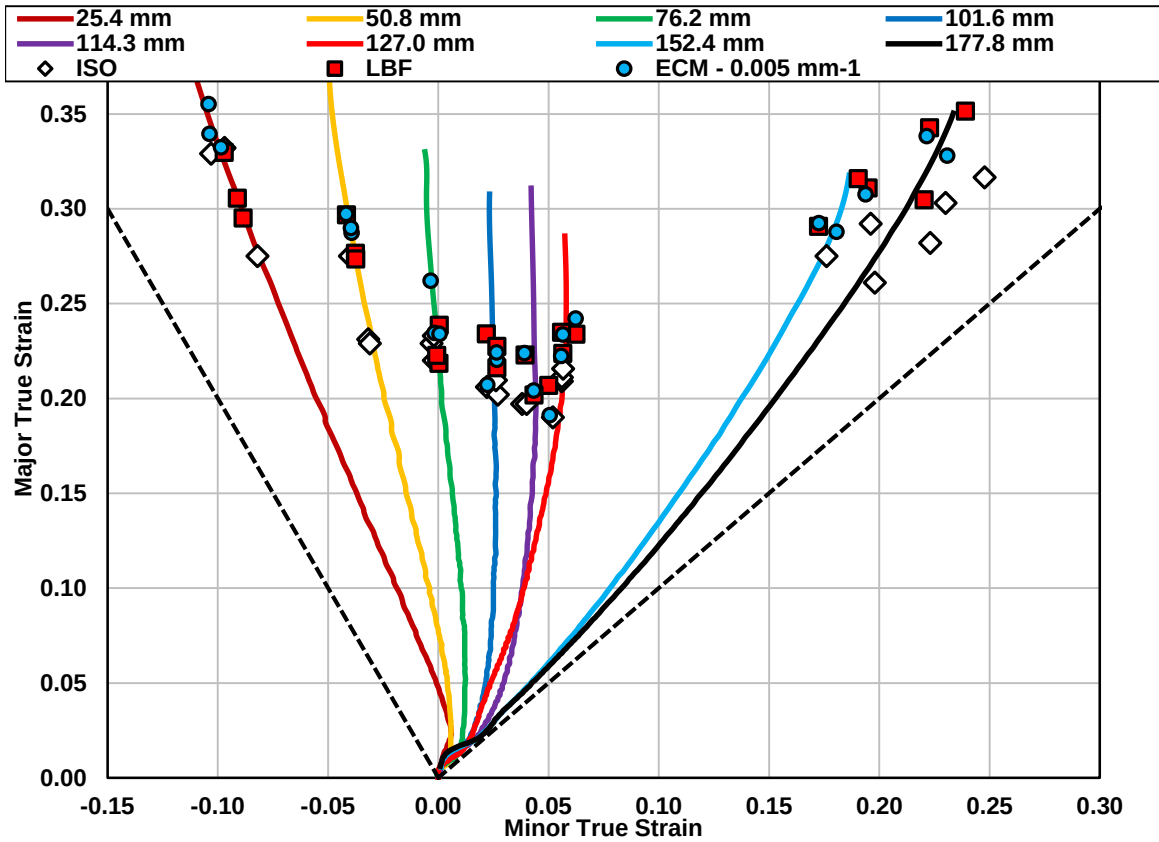


Figure 34: Individual limit strain points for each average strain path under Nakazima testing, as calculated using the ISO12004-2:2008 standard, the LBF method of Volk and Hora (2011), and the upper threshold of the Enhanced Curvature Method.

Table 3: Average limit strains from Nakazima testing as a function of specimen width determined with the ISO12004-2 standard, the LBF method, and the Enhanced Curvature Method. 95% confidence intervals are included in brackets to represent the error in each measurement.

Geometry	ISO12004-2:2008		LBF		ECM - 0.005 mm ⁻¹			ECM - 0.0005 mm ⁻¹		
	ε_2	ε_1	ε_2	ε_1	ε_2	ε_1	f_c	ε_2	ε_1	f_c
25.4 mm	-0.094 [0.01]	0.312 [0.04]	-0.092 [0]	0.31 [0.02]	-0.105 [0]	0.35 [0.01]	0.993 [0]	-0.1 [0]	0.334 [0.01]	0.9993 [0]
25.4 mm*	-0.094 [0.01]	0.312 [0.04]	-0.092 [0]	0.31 [0.02]	-0.102 [0]	0.342 [0.01]	0.9929 [0]	-0.091 [0]	0.307 [0.01]	0.9993 [0]
50.8 mm	-0.034 [0.01]	0.245 [0.03]	-0.039 [0]	0.282 [0.01]	-0.04 [0]	0.291 [0.01]	0.9929 [0]	-0.037 [0]	0.27 [0.01]	0.9993 [0]
76.2 mm	-0.002 [0]	0.227 [0.01]	0 [0]	0.227 [0.01]	-0.002 [0]	0.243 [0.02]	0.993 [0]	0.001 [0]	0.211 [0.03]	0.9993 [0]
101.6 mm	0.025 [0]	0.206 [0]	0.025 [0]	0.226 [0.01]	0.025 [0]	0.217 [0.01]	0.993 [0]	0.025 [0]	0.201 [0.02]	0.9993 [0]
114.3 mm	0.043 [0.01]	0.195 [0]	0.044 [0.01]	0.211 [0.01]	0.044 [0.01]	0.206 [0.02]	0.9929 [0]	0.044 [0.01]	0.191 [0.02]	0.9993 [0]
127.0 mm	0.056 [0]	0.212 [0]	0.058 [0]	0.231 [0.01]	0.058 [0]	0.233 [0.01]	0.9926 [0]	0.057 [0.01]	0.215 [0.02]	0.9993 [0]
152.4 mm	0.19 [0.01]	0.276 [0.02]	0.186 [0.01]	0.306 [0.02]	0.182 [0.01]	0.296 [0.01]	0.9911 [0]	0.165 [0]	0.254 [0.01]	0.9992 [0]
177.8 mm	0.234 [0.01]	0.301 [0.02]	0.227 [0.01]	0.333 [0.03]	0.234 [0.02]	0.348 [0.03]	0.9902 [0]	0.214 [0.01]	0.303 [0.02]	0.9991 [0]

In uniaxial loading (25.4 mm width), a minor sensitivity to the line slice orientation was observed where the ECM was less conservative when extracting DIC data parallel to the sample width (ISO standard) vs. parallel to the neck width. Both results are included in Table 3, but only the lower result, parallel to the neck width, is recommended for FLC determination. Other sample widths did not exhibit this sensitivity.

In biaxial loading, the ECM upper bound shows the highest limit strains of the three detection schemes. The f_c values decrease from 0.9929 for the 114.3 mm width geometry to 0.9902 – 0.9902 for the biaxial geometries. A similar but muted trend was observed in Marciniak testing where lower f_c measurements in biaxial loading were attributed to a higher level of thinning. Note, also, under biaxial loading that the ISO limit strains diverge from the average strain paths which is a result of the process used in the ISO method to compute ε_2 limit strains from the fitted strain distributions of ε_3 and ε_1 .

Tight radius stretch-bending was completed on a 76.2 mm width dog-bone for punch radii of 15 mm and 2 mm. The ECM and LBF limit strain approaches were applied to the tests with results summarized in Figure 35 (the ISO approach was not appropriate for this test). Both punch radii produced approximately plane strain loading as seen by the strain paths in Figure 35 which are also compared against the Marciniak and Nakazima paths for the 127.0 mm geometry. For both punch radii, the LBF limit strains were marginally higher than those from the ECM upper threshold and this observation is consistent with FLC0 observations in the Marciniak and Nakazima tests. In stretch-bend testing, it can be seen that the limit strains increase as tooling radius is decreased, with corresponding increases in superimposed bending strain and tooling contact pressure, as discussed by Min et al. (2016).

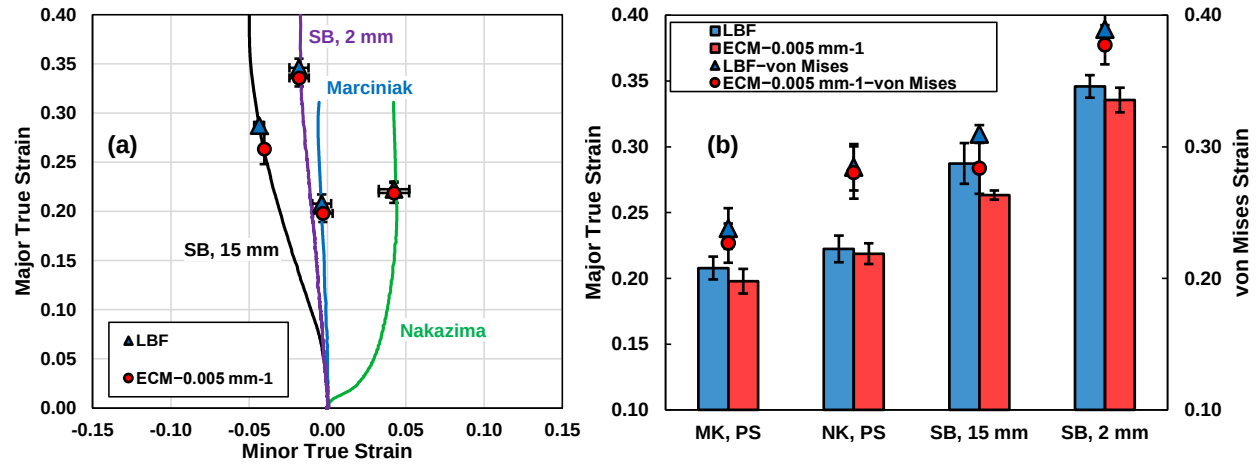


Figure 35: a) Average strain paths obtained from stretch-bend, Nakazima and Marciniak (plane strain) testing with corresponding limit strains using the LBF and ECM approaches. b) Corresponding limit strains (major and von-Mises equivalent true strains) plotted in order of tooling bend severity from lowest to highest.

4.2. Numerical and Experimental Investigation of the Formability of AA6013-T6

Warm forming experiments were performed on AA6013-T6 (2.0 mm) sheet, with the composition shown in Table 4. The experiments were performed to understand the effect of forming temperature on the formability of a 6000-series alloy, as well as to evaluate the use of curvature-based necking strain detection at elevated temperature. The experiments considered a Nakazima hemispherical punch at room temperature and warm temperatures of 150°C, 200°C, and 250°C. An earlier version of the curvature-based limit strain detection scheme was utilized that was the predecessor to the ECM of Section 4.1. Constitutive testing was also performed for the corresponding temperature conditions. Numerical simulations of the warm forming experiments were developed in order to understand the effect of elevated temperature and rate sensitivity on formability.

Table 4: AA6013-T6 nominal chemical composition (wt. %)

Alloy	Mg	Si	Cu	Mn	Fe	Others	Al
AA6013-T6	0.8-1.2	0.6-1.0	0.6-1.1	0.2-0.8	0.5	0.15	Bal.

4.2.1. Experiments

Digital image correlation techniques were used in all tensile and Nakazima testing, with full details of the procedures listed in Appendix B. Tensile testing in the rolling (RD), diagonal (DD), and transverse (TD) directions was completed at room temperature (RT), 150°C, 200°C, and 250°C for a nominal strain rate of 0.01 /s. At 200°C, tests were also performed at strain rates of 0.001/s and 0.1/s. All tensile testing was completed using ASTM E8 conformant geometries. Median Lankford coefficients and stress ratios were obtained from the tensile results and are summarized in Table 5. Stress ratios were computed relative to the rolling direction. No significant dependence of the Lankford coefficients or stress ratios on temperature was observed.

Table 5: AA6013-T6 Median R-Values and Stress Ratios

R-Values				Stress Ratios			
RD	DD	TD	Biaxial	RD	DD	TD	Biaxial
0.654	0.729	0.623	0.684	1.000	0.992	0.996	1.000

Room temperature forming limit testing followed a similar process to that employed in Section 4.1.1, with the specimen major axis aligned with the sheet RD. For elevated temperature testing, heated tooling was used for the Nakazima punch and dies and the samples were heated in-situ. Limit strain analysis used a curvature-based neck detection scheme. In this work, the onset of necking was assumed to correspond to a local curvature of zero which corresponds to the local loss of the initial curvature (flattening) of the Nakazima sample (the sample takes on the punch curvature of 0.02 mm^{-1} early-on in the forming process).

This criterion, herein referred to as zero-curvature, corresponds to a sharper neck than that detected using the 0.005 mm^{-1} upper bound curvature threshold considered in Section 4.1.3, and would be less conservative in nature. Three dog-bone geometries (Figure 26) were tested for FLC development: 25.4 mm (uniaxial), 76.2 mm (approximately plane strain), and 177.8 mm (biaxial).

4.2.2. Temperature Dependent Forming Limit Diagram

The temperature dependent forming limit diagram, determined using the zero-curvature neck detection method, is shown in Figure 36. The AA6013-T6 exhibits a mild increase in formability between RT and 200°C for all tested geometries. The distinction between the forming limit curves at elevated temperature is small and largely within the measurement variability, particularly for biaxial loading conditions. Beyond 200°C, a decrease in formability is seen for plane strain conditions. For the 76.2 mm geometry, major applied limit strains of 0.197, 0.213, 0.246, and 0.236 were observed for RT, 150°C, 200°C, and 250°C, respectively.

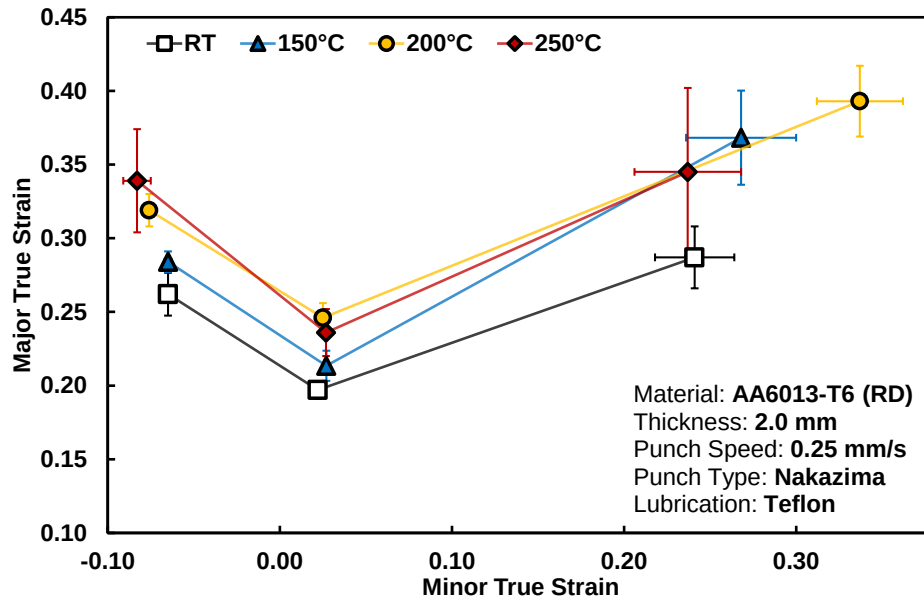


Figure 36: AA6013-T6 Temperature Dependent FLD, with scatter bars presented using the median absolute deviation.

The effect of the difference in curvature threshold used to assess the current AA6013-T6 data versus that used for the AA5182-O data in Section 4.1 can be seen by examining Figure 37 which shows the evolution of the imperfection factor versus the major limit strain for the 76.2 mm AA6013-T6 geometry at room temperature. Figure 37 was generated from eight repeated tests and using the same methodology as that used in the development of Figure 32. Note that after shifting the curves to the mean strain for $f_c = 0.9933$ (upper threshold value from Section 4.1.3), the error band in Figure 32a) almost completely collapses in Figure 32b). This was also observed for AA5182-O, but because AA6013-T6 has no PLC

effects, the trend is even more readily seen, further demonstrating that the geometric evolution of a neck is consistent, but the initiation of the neck may have a stochastic component.

For the established limit strain of 0.197, the corresponding imperfection factor is approximately 0.972. If one adopts the 0.9933 imperfection factor from the threshold work completed for AA5182-O in Section 4.1.3, a limit strain of 0.167 might be expected from Figure 37a) which would be in-line with a calculated ISO limit strain of 0.166, as reported in DiCecco et al. (2016). Thus, the RT FLC might be shifted downwards by 0.03 major strain at FLC0 for AA6013-T6. The RT formability of the AA6013-T6 alloy is modest and lower than that seen for the AA5182-O (Figure 34). It is expected that formability at other temperatures would be see a similar or greater decrease when considering the 0.9933 imperfection factor. Assuming a fixed 0.03 offset, corrected limit strains for the 76.2 mm geometry would be 0.183, 0.216, and 0.206 for 150°C, 200°C, and 250°C, respectively. This is further discussed in 4.3.

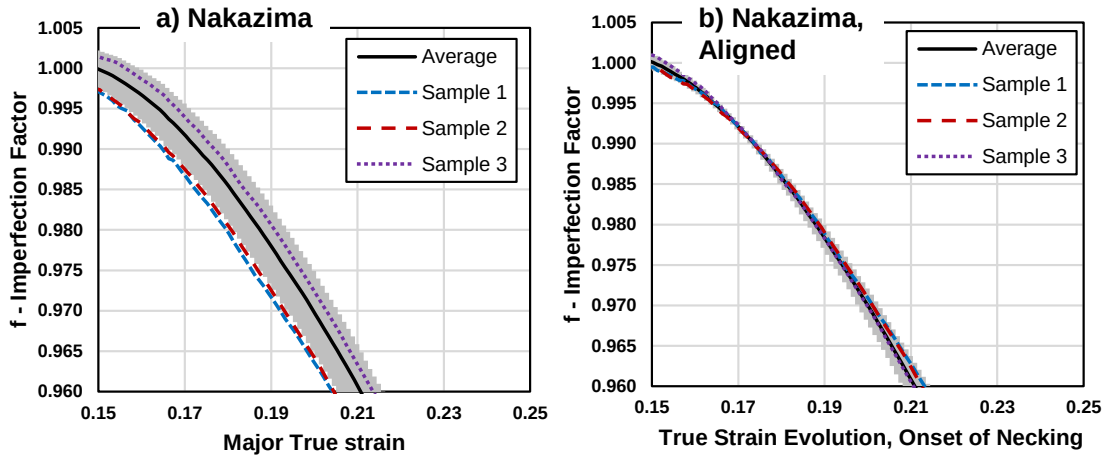


Figure 37: Evolution of the imperfection factor in AA6013-T6 sheet, 2.0 mm, for a 76.2 mm dog-bone geometry tested using a Nakazima hemispherical punch at room temperature.

4.2.3. Constitutive Model Development and Simulation of Elevated Temperature Formability

Temperature dependent flow curves are shown in Figure 38, with a Hockett-Sherby formulation used for extrapolation beyond necking, with more details in Appendix B. The data at 0.01 s^{-1} exhibits a clear decrease in flow stress with increased temperature. Furthermore, increases in strain rate (at 200 °C) reveal reasonably strong rate sensitivity at elevated temperature.

In order to delve into the increase in formability exhibited by this alloy at 200°C versus RT, finite element simulations were performed, the details of which are given in Appendix B. The constitutive model utilized the flow stress data in Figure 37, both with and without the rate sensitivity effect. A Barlat YLD2000 yield surface was also calibrated using the values from Table 5 and a yield exponent of 8 (for

FCC materials) was assumed. The normalized calibrated yield surface is shown in Figure 39 and compared to a *von Mises* yield surface for reference. The coefficients of the calibrated yield surface are found in Appendix B.

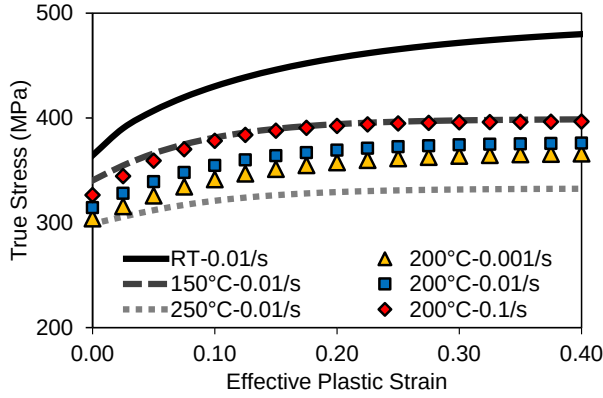


Figure 38: AA6013-T6 temperature dependent flow curves

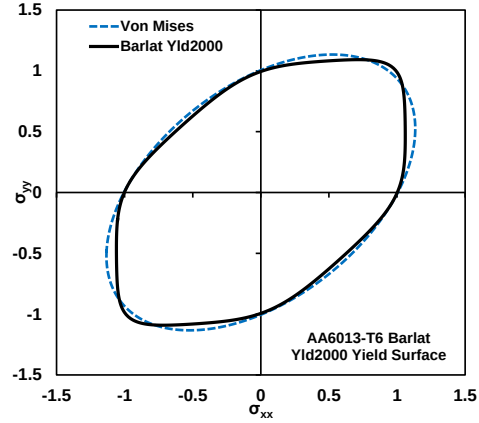


Figure 39: AA6013-T6 temperature independent Barlat YLD-2000 yield surface

A model of the 76.2 mm width Nakazima experiments is developed and simulated in LS-DYNA at RT and 200°C. The predicted punch load-displacement response was in good agreement with the test results at room temperature, as shown in Figure 40. At elevated temperature, the model was run with and without strain-rate sensitivity. The elevated temperature load-displacement response was found to be effectively insensitive to strain-rate effects, with good agreement between the test and model. The predicted and measured thinning-strain distribution at 200°C is shown in Figure 41 for a time corresponding to the onset of localization from a representative sample. Predictions are shown for the material model with and without strain rate sensitivity. The effect of strain-rate is quite significant insofar as the model without rate effects has already significantly localized, but the model with rate effects is relatively stable and in good agreement with the test strain distribution. Here the strong rate sensitivity at elevated temperature acts to resist the onset of necking leading to increased warm formability.

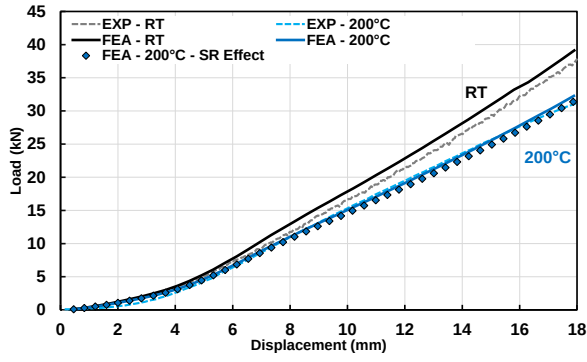


Figure 40: Experimental and model load-displacement curves for AA6013-T6 76.2 mm Nakazima tests.

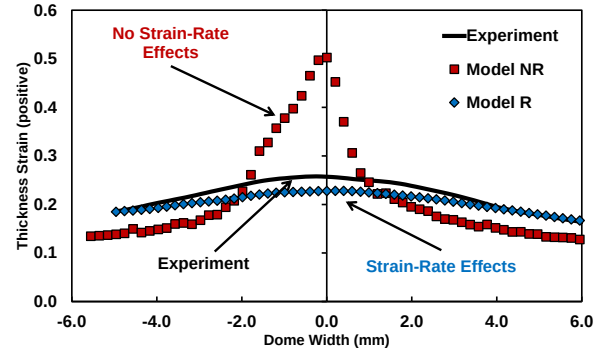


Figure 41: Strain-rate effects on thickness strain distribution at 200°C (thinning taken as positive).

4.3. Warm Aging of Pre-aged AA6013 Sheet and its Relevance to Room Temperature and Warm Forming Applications – Experimental and Modeling Analyses

As shown in Section 4.2.2, warm forming of AA6013-T6 only resulted in a mild enhancement in formability vs. room temperature. Experiments were undertaken to examine whether different starting tempers (other than peak aged) would result in a stronger warm forming response. Elevated temperature tensile and formability tests in the temperature range of 220°C to 280°C were conducted on under aged (referred to herein as pre-aged) AA6013 (designated AA6013-PA). The pre-aging heat treatment consisted of a 4-hour, 100°C pre-aging cycle following solutionization. To establish the conditions that would potentially optimize both the elevated temperature formability and the final mechanical properties (tensile and bendability) of the AA6013-PA alloy, aging studies were conducted to determine the precipitation kinetics for this alloy and initial temper. Subsequent warm forming tests were conducted for selected thermal histories at various warm forming temperatures, as well as in-situ warm constitutive characterization of the warm forming process using a thermo-mechanical simulator.

4.3.1. Experiments

Industrially produced 2.0 mm thick AA6013-T6 sheet, obtained from Arconic, with typical microstructure and chemical composition given in Appendix C, were subject to the thermo-mechanical processing histories shown in Figure 42. To obtain a pre-aged (PA) temper of the alloy, samples were solutionized at 560°C for 10 minutes in a fluidized sand furnace (FB-08C Fluidized Bath Furnace, Techne®), water quenched, and then aged at 100°C for 4 hours and water quenched a second time.

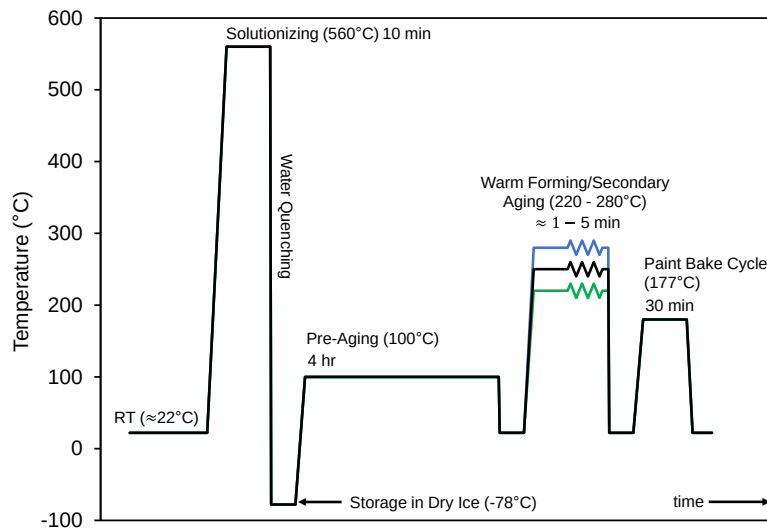


Figure 42: Processing routes used for warm forming of pre-aged aluminum alloys. Note that dry ice storage was used in the current work to limit aging between the quenching and pre-aging cycles.

After pre-aging, samples were either subject to warm formability testing, warm tensile testing, aging heat treatments (un-deformed) followed by mechanical testing, or warm pre-straining tests followed by calorimetry (The calorimetry results formed a small portion of the entire work scope, so these results are omitted herein and the reader is referred to Appendix C).

For the aging studies, ASTM-E8 sub-sized tensile coupons were used. In order to simulate aging occurring during warm forming, the AA6013-PA samples were subjected to a secondary aging process in the fluidized sand furnace and quenched in room temperature water. Isothermal secondary aging treatments were conducted on the tensile coupons over the temperature range of 220°C to 280°C. The exposure times for these isothermal heat treatment times varied from 30 s to 3000 s in duration. A subset of these tensile coupons was further treated with an additional laboratory aging process of heating at 177°C for 30 minutes to simulate a paint-bake cycle (PBC). AA6013-PA specimens (60 mm by 60 mm by 2 mm) were also subjected to bending tests after each aging step following the VDA 238-100 standard (VDA, 2020).

Formability characterization of the PA and T6 tempers was undertaken isothermally at RT, 150°C, and 230°C. A full description of the testing process is included in Appendix C; however, the formability testing procedures were consistent with those described in Section 4.2.1. A 76.2 mm dog-bone type geometry (Figure 26) was utilized which produced approximately plane-strain loading for the AA6013 alloy. Testing was completed in the rolling direction. The Enhanced Curvature Method described in Section 4.1.3 was used for limit strain determination.

In order to examine the potential coupling between aging and deformation during warm forming, warm tensile testing of AA6013-PA tensile samples was performed using a thermo-mechanical simulator (Gleeble-3500, Dynamic Systems Inc.). The purpose of this testing was to understand the effects of time at temperature and strain-rate at 230°C, adopted here as a representative warm forming temperature, on the in-situ flow response of the alloy in the PA condition and to examine the effect of in-situ aging on formability. Samples were first heated (while subject to a small pre-load of 200 N) to the target temperature of 230°C in 10 s, held at temperature for a predetermined hold time, and then pulled to failure at a constant stroke rate. The *hold time* parameter was intended to characterize the effect of secondary aging at 230°C on the subsequent at-temperature stress-strain response of the PA temper. From a practical standpoint, different hold times of 5, 70, and 180 s were considered to reflect the range of pre-heating durations within the context of industrial warm forming processes incorporating different heating methods. Nominal strain rates of 0.01/s and 0.1/s were adopted to assess the strain rate sensitivity of the -PA temper at temperature.

4.3.2. Aging Characteristics

The evolution of the room temperature yield stress (YS), ultimate tensile strength (UTS), and toughness (integral of the tensile stress-strain response) of AA6013-PA after secondary aging at temperatures ranging from 220°C to 280°C is shown in Figure 43. For material peak aged at 220°C, the YS is slightly lower than the measured as-received (T6) YS, while the UTS is slightly higher. For short exposure times, the YS of the material does not decrease when aged at 250°C and below, compared to the PA starting condition, indicating the PA treatment sufficiently stabilizes precipitates such that dissolution does not occur upon higher temperature exposure. Aging at 265°C and 280°C results in a small decrease in the measured YS at the shortest exposure time of 30 s, but then recovers. The UTS follows a similar trend as the YS at these short exposure times; however, the time required to recover the UTS is longer. Scatter in the tensile toughness data is larger than the YS and UTS data; however, for aging at 265°C and 280°C, the tensile toughness drops below that of the AA6013-T6 condition. For exposure at 250°C, the tensile toughness after 100 to 400 s exposure remains higher than the T6 condition, although it does appear that at long exposure times the final toughness is equivalent to that of the T6 condition. For lower temperature exposures, the final tensile toughness remains greater than the AA6013-T6.

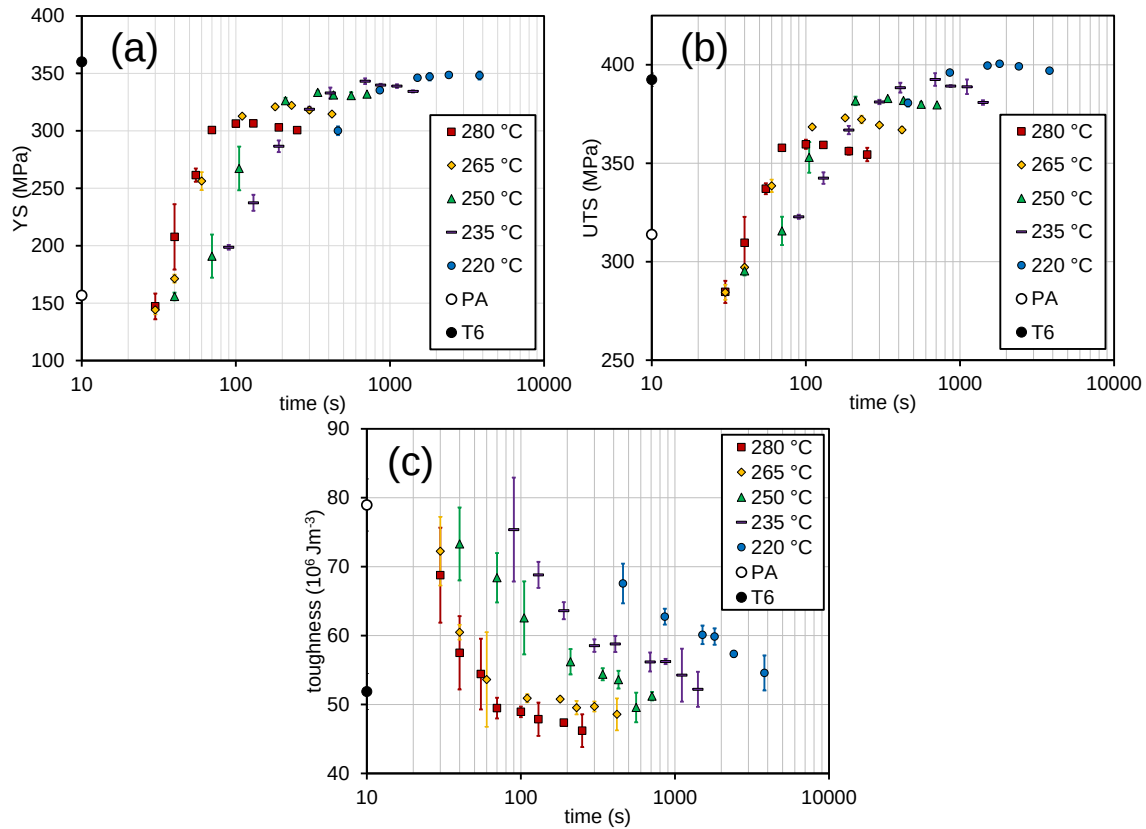


Figure 43: Isothermal aging (between 220 °C and 280 °C) response of AA6013-PA: (a) yield stress (b) ultimate tensile strength, and (c) tensile toughness.

The effect of the PBC following secondary aging on the flow behavior of AA6013-PA is examined in Figure 44. The aging temperature of 235°C was selected since it offered a high level of toughness (see Figure 43) and a strong difference in yield strength between short (90 s) and long (410 s) secondary aging durations. The UTS following the PBC and short aging times is very close to the UTS for AA6013 in the T6 condition. Interestingly, the corresponding YS is relatively low compared to the T6 condition, resulting in a stronger work hardening response and higher elongation. As the aging duration increases, the observed differences in the pre and post paint-baked engineering stress-strain curves reduces. For all secondary aging times considered, the strain hardening rates and elongations of the pre and post paint-baked specimens remain higher than the strain hardening rate and elongation in the T6 condition. The lower yield strength levels (relative to the T6 condition) are likely a consequence of the difference in aging temperatures between the as-received T6 material, which would have been aged at temperatures lower than 180°C, and the aging temperature used in this series of tests, *i.e.*, 235°C. From Figure 44, the aging temperature strongly influences the peak YS that can be obtained, which is a direct manifestation of the effect of temperature on the precipitate population (*i.e.* dependence on the metastable phase diagram (Esmaeili, 2002), as well as solute diffusivity.

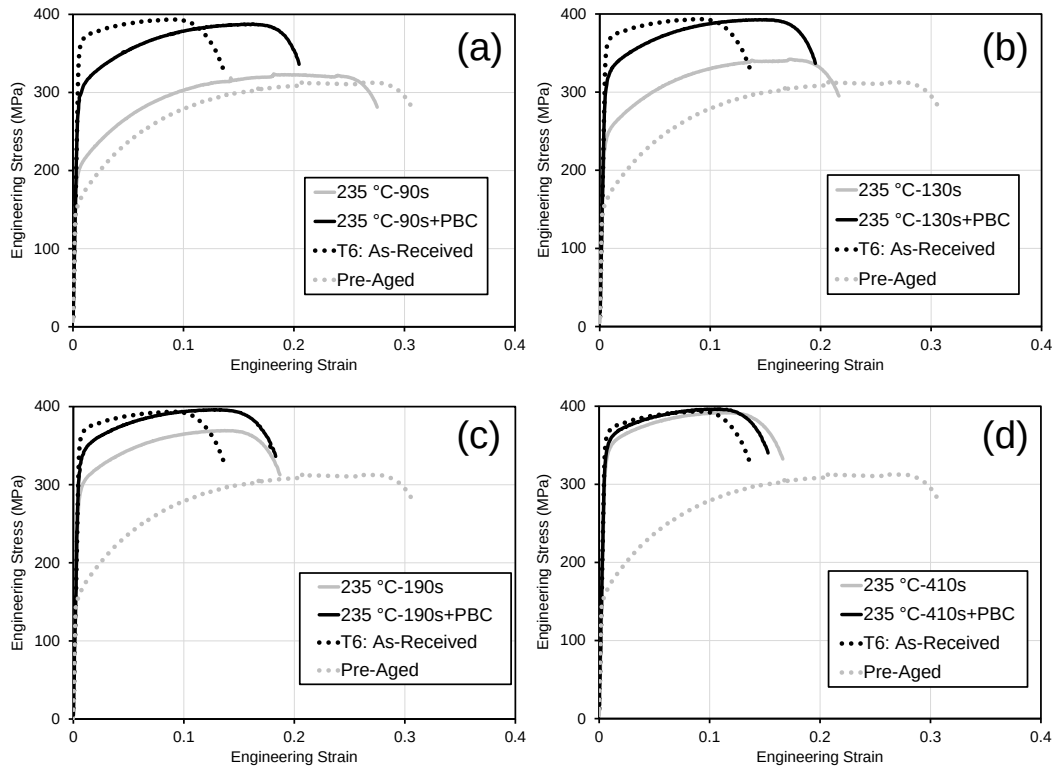


Figure 44: Engineering stress vs. engineering strain of AA6013-PA before and after PBC, for specimens exposed to aging at 235 °C for (a) 90 s, (b) 130 s, (c) 190 s, and (d) 410 s (25 mm gauge length). Engineering stress vs. engineering strain corresponding to PA and T6 conditions area also included for comparison purposes.

The effect of aging duration and paint-bake cycle on subsequent sheet bendability (fracture resistance) was also examined. These samples also considered an aging temperature of 235°C and durations of 90 to 410 s. The bendability results for AA6013-PA aged at 235°C are shown in Figure 45. To achieve a bend angle greater than the AA6013-T6 condition, the secondary aging (warm forming) exposure time should be limited to less than 300 s. For exposure times less than or equal to 300 s there is a clear distinction between the pre and post PBC AA6013-PA conditions. Although the post PBC AA6013 still has a final bend angle greater than the AA6013-T6, in order to obtain a final bend angle greater than 80-100°, the minimum range shown by Kurz et al. (2016) to suppress fracture during automotive side impact, the secondary aging (warm forming) time must be kept less than 90 seconds. In Figure 45b, the bend angle has been plotted against the strain hardening potential, defined here as the difference in UTS and YS. The resulting linear trend shows a strong correlation, indicating that fracture resistance during crash can be improved by maintaining a strong work hardening potential.

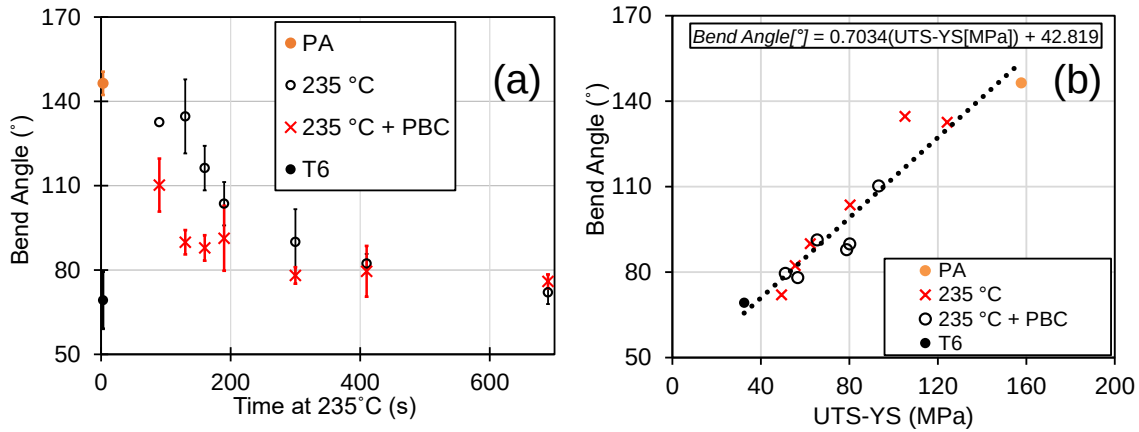


Figure 45: (a) Bend angle as a function of duration of aging at 235 °C with and without PBC. Also shown are bending angles corresponding to PA and T6 conditions; and (b) hardening potential (UTS-YS) vs. bend angle for AA6013.

The current results have potential practical implications related to the warm forming of 6000-series aluminum alloys using a pre-aged starting temper. First, depending on the temperature, time, and strain imposed during warm forming and subsequent paint curing, UTS levels and, to a lesser extent, YS levels very close to T6 values can be readily achieved if a PBC is utilized. In a manufacturing setting, the time at temperature prior to forming, warm forming, and subsequent PBC treatments can be utilized to obtain parts with a high strength and bendability (bend angle greater than 110° as in Figure 45).

4.3.3. Elevated Temperature Constitutive Behaviour

Despite the potential mechanical benefits associated with using AA6013-PA as a starting material, the practical formability implications associated with its warm forming require assessment to determine its overall usefulness of this initial temper vs. a T6 starting condition. As a first assessment, elevated

temperature tensile tests were performed primarily to understand the effect of aging during pre-heating and warm forming on the mechanical response of the alloy during warm forming. The engineering stress-strain response of AA6013 in the pre-aged condition is presented in Figure 46a for a range of heating profiles and strain rates, while Figure 46b shows the corresponding true stress versus effective plastic strain curves for each condition. The different hold times prior to tensile testing serve to characterize the effect of secondary aging during heating on the “at-temperature” tensile response of the PA temper.

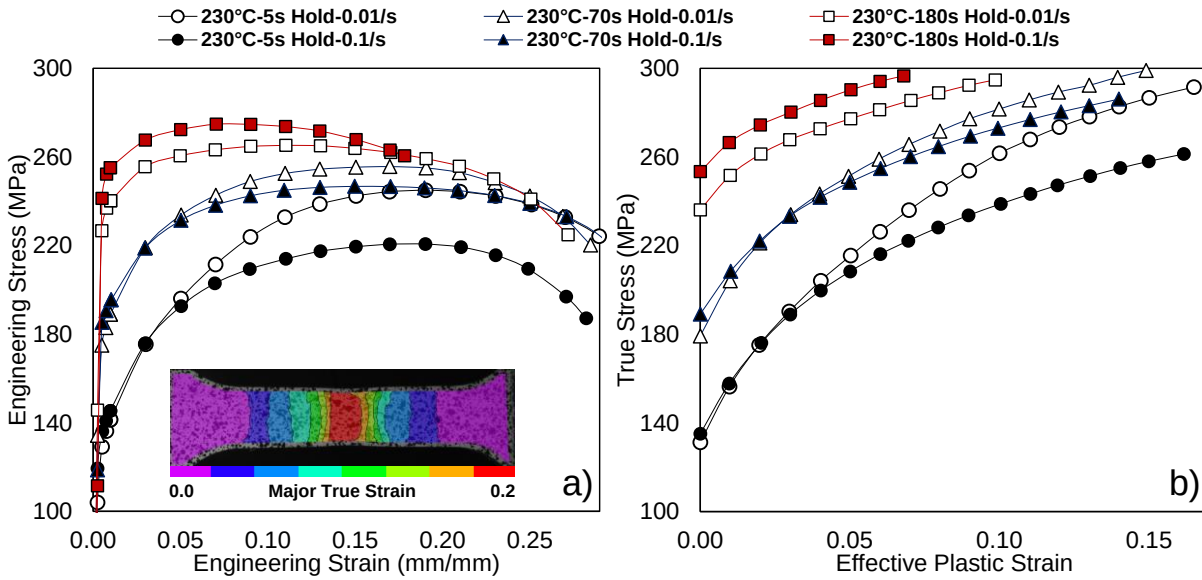


Figure 46: Tensile response of AA6013-PA at 230 °C as a function of hold time following heating to temperature and strain rate. Samples were heated rapidly using a thermomechanical simulator and held for various times at 230°C (5 s, 70 s, and 180 s) prior to tensile testing. Note that representative curves are plotted for each test condition and that the scatter is relatively low, as captured in Table 3.

The yield strength at 230°C is plotted in Figure 47 (tabulated data given in Appendix C) for both strain rates as a function of hold time where the yield strength significantly increased with the hold time that is attributed to secondary aging during the heating and hold period. For comparison purposes, the predicted increase in room temperature yield strength, obtained using the kinetics model developed as part of this study (for details see Appendix C), is also plotted in Figure 47. While these predictions are for room temperature, the aging predictions scale well with the measured increase in elevated temperature yield strength as the hold time increases.

For all hold times, the higher strain rate led to an increase in YS relative to the lower strain rate that evolved from 2 MPa to 14 MPa for the shortest to the longest hold time. A positive difference in YS for the imposed increase in strain rate indicates a positive strain-rate sensitivity at yielding at 230°C that was most pronounced for the longest hold time.

The hardening behavior after yielding for the longest hold time of 180 s also exhibits positive strain rate sensitivity, as seen in Figure 46b, which is consistent with other stable (solution strengthened or peak aged) aluminum alloys at elevated temperature (DiCecco et al., 2017; Ghosh et al., 2014). Of particular interest in the current results, the strain hardening response at strains beyond yielding for the shorter 5 s and 70 s hold time experiments exhibits a seemingly negative strain rate sensitivity. As shown in Figure 46b, the lower rate curves at 5 s and 70 s exhibit a higher degree of hardening, overtaking the strength of the higher strain rate curves after approximately 3-4% tensile strain. This was further exemplified by calculating the hardening exponent (n -value) for effective plastic strains in the range of 0.04 – 0.06, which is plotted in Figure 47. For a hold time of 5 s, the average n -values for the 0.01/s and 0.1/s rates were 0.27 and 0.20, respectively, which again reflects a negative rate sensitivity of the hardening response. For the 70 s hold time, the n -values were somewhat lower and still exhibited a mild negative rate sensitivity, with n -values for the lower and higher strain rates being 0.16 and 0.14, respectively. For the longest hold time of 180 s, the calculated n -values were both effectively 0.08 for the lower and higher strain rates.

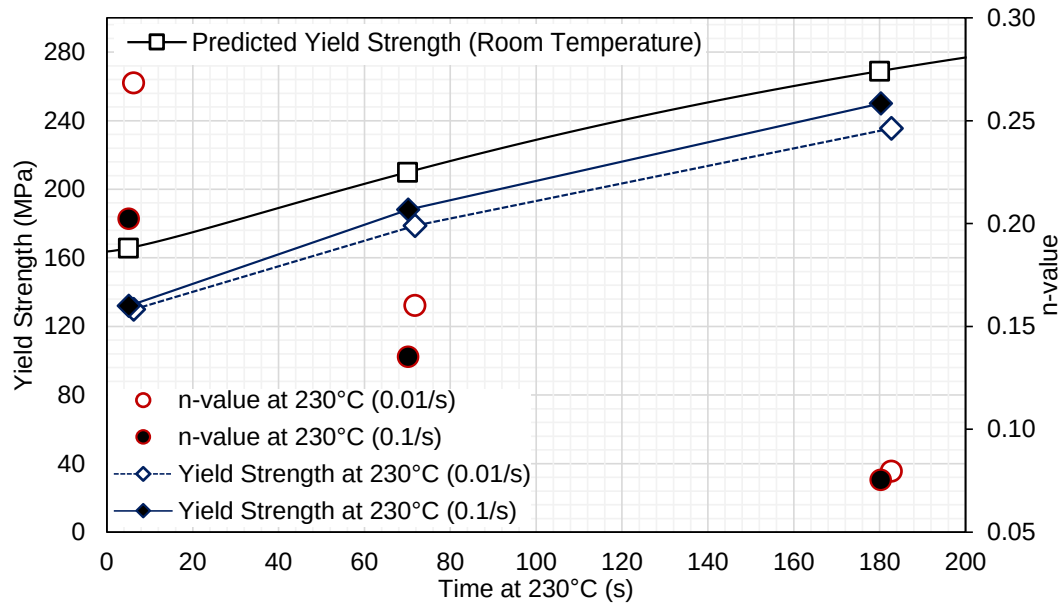


Figure 47: Yield strength and strain-hardening exponent of AA6013-PA at 230°C as a function of strain rate and time at temperature (hold time) prior to testing. The predicted room-temperature yield strength as a function of aging time at 230°C is also plotted.

The decrease in the degree of hardening observed at higher rates of strain for shorter hold times is somewhat surprising since aluminum alloys typically exhibit an increase in work hardening with increases in strain rate at elevated temperatures. The overall hardening response captured by the calculated n -values in Figure 47 has several components, the first being conventional work hardening due to plastic strain accumulation. There will also be an age hardening contribution during the tensile test that will be larger at

lower strain rates due to the associated increase in the test duration leading to longer times at elevated temperatures. This effect is most pronounced for the PA temper and shorter hold times for which the extent of secondary aging prior to the start of the tensile test is lower, as reflected by the YS data in Figure 47.

4.3.4. Warm Forming Observations and Discussion

The formability of AA6013 was characterized in the PA and T6 tempers under isothermal conditions at room temperature, 150°C, and 230°C, at a stroke rate of 1.0 mm/s, using a plane strain loading condition. Representative strain paths taken at room temperature for each temper are plotted in Figure 48a (note that the strain paths shown extend to the onset of fracture). Both strain paths demonstrate typical Nakazima characteristics, starting with biaxial bending as the blank wraps around the hemispherical punch geometry, followed by near-plane strain stretching controlled by the blank width and lubrication condition.

The measured plane strain limit strains using the ECM threshold of 0.005 mm^{-1} from Section 4.1, as a function of forming temperature and initial temper, are plotted in Figure 48b. Also plotted are the dome heights at onset of fracture for each case. At room temperature, the major true limit strain of the PA temper was 0.28 ± 0.01 and 0.18 ± 0.00 for the T6 temper, marking, on average, a 55% improvement in formability of the PA temper relative to the T6 baseline. This significant formability improvement associated with the PA temper is attributed to the much stronger work hardening (n-value) relative to the T6 temper, as is evident in Figure 47. In contrast to the trends of Section 4.2, the T6 temper shows no meaningful difference in formability (limit strain and dome height) with temperature for the range of test temperatures. This might be the result of comparison between the ECM and the prior method for analysis used before.

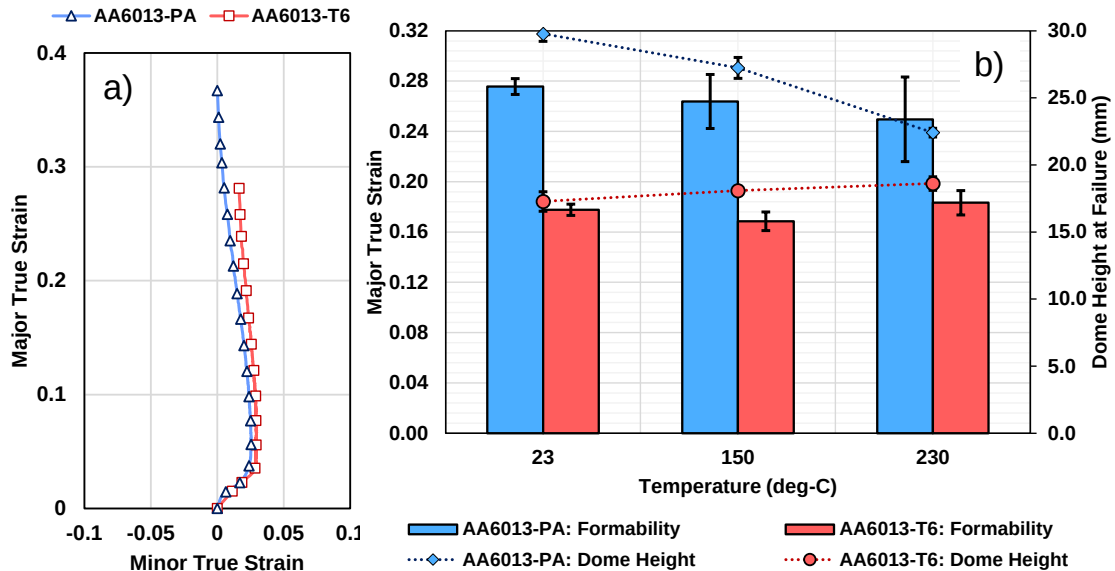


Figure 48: (a) Room temperature strain paths for AA6013-PA and -T6 tempers and (b) major limit strains and average dome heights at failure for the two tempers as a function of temperature.

In general, the elevated temperature formability of the PA temper was found to decrease relative to RT while the repeatability also decreased, as evidenced by the increase in the scatter of the measurements. The final dome heights prior to fracture further confirm this trend, with a decrease from 30 mm to approximately 23 mm from RT to 230°C. These plane strain formability measurements are in alignment with the warm tensile test observations shown in Figure 46. The hardening rate of the PA temper at 230°C drops quickly with time at temperature due to in-situ aging which accounts for the loss in formability during warm forming. This outcome begs the question as to whether formability could be increased by using fast heating and processing times; however, the limitations on heating rate for the warm formability testing conducted in the present study resulted in no appreciable gains in formability for the AA6013-PA material. In these experiments, heating rates were limited by the contact heating rates of the tooling. In total, 3 minutes was required to heat the specimens to the target temperatures, with 60-90 seconds of asymptotic heating. Thus, the current 230°C warm formability experiments on the PA temper reflect a condition that likely falls somewhere between the 70 s and 180 s hold times in the warm tensile tests for which n-values of approximately 0.16 and 0.08 were measured, respectively, as opposed to the 5 s hold time for which an effective n-value of approximately 0.27 was measured.

Warm forming was not found to be beneficial for the PA temper using the current specimen heating approach (duration); however, the room-temperature formability of the PA temper was excellent and showed a significant improvement over the peak aged formability. From Figure 44 it can be seen that the PA temper, after exposure to a thermal cycle of 410s (6.8 minutes) at 235°C, will have a stress-strain response in-line with the baseline T6 temper (independent of a subsequent paint-bake.) Thus, the PA starting temper, when combined with a relatively short secondary aging cycle after forming, offers a T6 final strength level with a 55% improvement in formability. This cycle could be further optimized using the calibrated kinetics model described in Appendix C, based on the desired final strength and ductility characteristics or for different heating cycles based on furnace capabilities, such as in the work of Pishyar (2021), as discussed in Section 2.3.1.

4.4. Effect of Initial Temper on the Warm Forming Characteristics of a High Strength 7000-series Al-Zn-Mg-Cu Alloy

The effect of the starting temper on the warm forming response and subsequent aging response was also studied for a developmental 7000-series alloy, designated as AA7xxx for the purposes of this work. The nominal composition of the alloy is given in Table 6. The alloy was produced by Arconic and supplied in the T76 over-aged condition for improved stress corrosion cracking (SCC) resistance (Sun et al., 2013; Chen et al., 2014). The tempers studied in this work were pre-aged (PA), peak-aged (T6) and over-aged (T76). Warm forming conditions consisted of isothermal Nakazima hemispherical testing at RT, 150°C, 175°C, and 200°C in the transverse direction of the material. The aging response to warm forming and a paint-bake cycle was studied, as well as the warm constitutive response at 175°C. The primary purpose of this study was to identify optimal forming and aging conditions that balance formability and final strength for commercially attainable processing conditions. Full details of the work and results may be found in Appendix D, with a synopsis of the experimental methods and results in the remainder of this section.

Table 6: Nominal composition of AA7xxx, in wt %.

Alloy	Cr	Cu	Fe	Mg	Mn	Si	Ti	Zn	Zr	Other	Al
AA7xxx	<0.04	1.3-2.0	<0.08	1.2-1.8	<=0.4	<=0.06	<=0.06	7.0-8.0	0.08-0.15	0.15	Bal.

4.4.1. Experimental Method

The PA and T6 AA7xxx tempers were produced in two steps: (i) the sheets were first solutionized in a fluidized sand bath and quenched and (ii) subjected to aging. Dry ice was used to store samples between the quenching and aging steps to facilitate bulk processing. Details on the selected aging processes may be found in Appendix D, with additional details regarding the PA temper given by Abolhasani et al. (2017) and Omer et al. (2018). Representative tensile stress-strain curves from the PA, T6, and T76 sheets are shown in Figure 49. The PA temper exhibits the highest work hardening and the T76 the least. Additionally, there is a distinctive flat region after yielding not observed in the PA and T6 tempers. Fribourg et al. (2011) explored this behaviour in detail for AA7449 and showed that this flat region after yielding emerges and becomes more prominent as the extent of overaging increases.

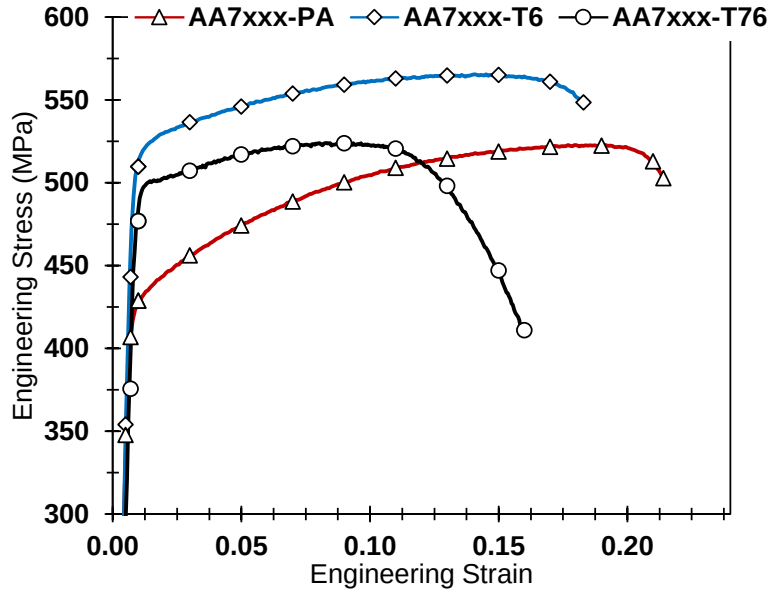


Figure 49: Representative stress-strain curves of AA7xxx in the PA, T6 and T76 tempers, tested in the rolling direction.

Formability testing was completed using a 101.6 mm Nakazima hemispherical punch and corresponding tooling. The testing was completed under isothermal conditions, where all tooling was heated with heater cartridges and controlled in a closed-loop PID system. Full details on the testing may be found in Appendix D; however, the general methodology was consistent with the procedures detailed for Nakazima warm formability testing in Sections 4.2 and 4.3 and Appendices B and C. Formability was evaluated using a 76.2 mm dog-bone type geometry (Figure 26) which corresponds to an approximately plane strain condition. Limit strains were determined using the ECM (using a 0.005 mm^{-1} threshold) as well as the ISO12004-2:2008 standard. Stereoscopic DIC was employed for all formability and constitutive characterization.

The aging response of the PA and T6 tempers was characterized as a function of the warm forming process with and without a simulated paint-bake of 177°C for 30 minutes. Subsize ASTM E8 samples were extracted from the flanges of the dome samples exposed to the warm thermal profile(s), but undeformed. Samples were extracted in the rolling direction, consistent with the results in Figure 49.

The warm constitutive response of the PA and T76 tempers was measured at 175°C for nominal strain rates of 0.01/s and 0.1/s as a function of different pre-heat times of 5 s and 180 s, respectively, to simulate “fast” and “slow” heating conditions. Note that the 180 s slow heating condition aligns more closely to the time taken to heat the samples in Nakazima testing. A closed-loop thermo-mechanical simulator (Gleeble-3500, Dynamic Systems Incorporated) was used for this work, the same system used for the AA6013 warm constitutive work in Section 4.3.

4.4.2. Formability Results

Major true limit strains are plotted in Figure 50 for all temper and temperature conditions. For all tests, the minor limit strains were on the order of 0.016 – 0.030 which is close to a plane strain loading condition. At room temperature, the PA temper had a limit strain of approximately 0.21, while T6 and T76 tempers were 0.18 and 0.12, respectively. For comparison, the limit strains for AA6013 -PA and -T6 for the same sample geometry and testing procedures were 0.28 and 0.18, respectively. These trends were consistent with the work hardening trends in Figure 49. At 150°C, the limit strains are improved for all three tempers, with the greatest improvement observed for the T76 temper, albeit still at levels below PA and T6 room temperature conditions. Beyond 150°C, it is difficult to see further improvements in formability, given the increase in scatter and the divergence between the ECM and ISO methods. Additionally, for the PA and T6 tempers, the recorded dome heights are observed to decrease as a function of temperature, which is opposite to the trends in the limit strains.

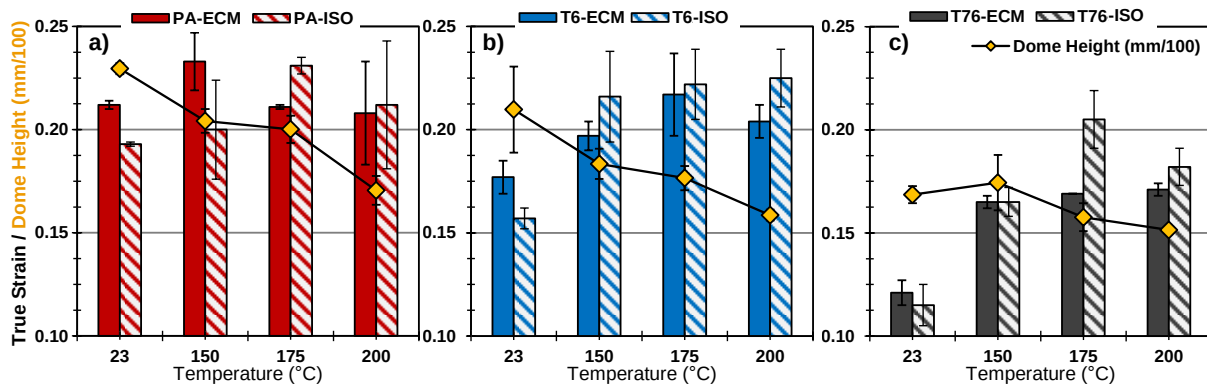


Figure 50: Major true limit strain (plane strain sample geometry) as a function of temperature for the different AA7xxx tempers: a) PA, b) T6, and c) T76. Solid bars indicate limit strains based on ECM, while dashed bars indicate values from ISO method. Dome heights corresponding to failure are plotted on the same axis in units of mm/100.

4.4.3. Aging Response of PA and T6 Tempers Following Warm Forming and Paint-Bake

The effect of the warm forming and paint-bake thermal cycles on samples starting with either a PA or T6 initial temper can be seen in Figure 51. The triangular symbols correspond to the engineering stress-strain response in the initial temper conditions. The stress-strain curves after the warm forming thermal cycle (at either 150, 175 or 200 °C) are indicated using open symbols for each warm forming temperature. The closed symbols indicate the stress-strain curves following a subsequent paint-bake cycle. For the PA condition, a decrease of 16 MPa in YS relative to RT is observed in the stress-strain response after exposure to the lowest WF temperature of 150°C (Figure 51a), for which the flanges were at the target temperature for under 120 seconds. After the paint-bake, a significant aging response is observed and the strength of the warm formed PA samples approach that of the RT T6 condition (seen in Figure 51 d-f). The observed drop

and subsequent recovery in strength may be indicative of retrogression and re-aging (Ward et al., 2011) for the 150°C WF+PB cycle. After WF at the higher temperatures of 175°C and 200°C, the PA temper shows a positive aging response to both WF and WF+PB, in that the strength increases relative to the RT PA condition, while the ductility decreases. The highest tensile strength of 546 MPa $\pm$ 2 MPa is achieved after the 175°C WF+PB cycle, while the strength following the 200°C WF+PB is lower which may indicate overaging during the PA+200°C+PB cycle.

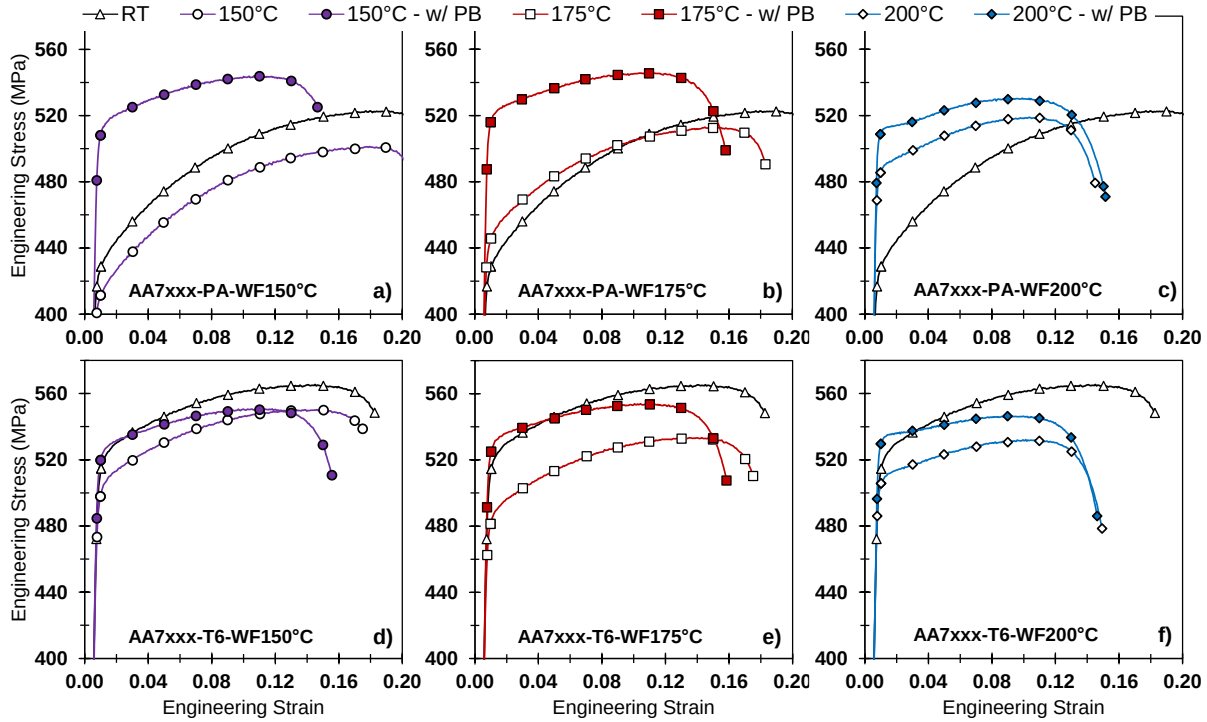


Figure 51: Tensile response of a) – c) the PA temper and d) – f) the T6 temper prior to forming (RT - triangular symbols), after warm forming at the temperatures indicated (open symbols), and after a subsequent paint-bake of 177°C for 30 minutes solid symbols. The curves plotted are the average of three tests.

The trends in strength and ductility are summarized across Figure 52. After the 200°C forming cycle and paint-bake, the ductility, in terms of the UE and TE, of the PA and T6 tempers approximate the as-received T76 temper. From a SCC susceptibility standpoint, the trends towards overaging in the PA temper following the 200°C+PB route may improve the SCC response (Sun et al., 2013; Chen et al., 2014) while offering high strength, albeit with no formability improvements relative to lower temperatures. By contrast, the RT and 150°C forming routes offer excellent formability, but may require additional investigation to achieve good SCC resistance (note that SCC behaviour is outside the scope of this thesis).

A softening in the stress-strain response of the T6 temper relative to the RT baseline is observed after warm forming (Figure 51 d-f) which is attributed to a retrogression response (Fribourg et al., 2011). This effect is most significant at 175°C, with a drop in YS and UTS of 31 MPa (6%) and 32 MPa (6%),

respectively. A positive re-aging response is noted following PB of the T6 samples for all WF cycles, with the most significant re-aging occurring again at 175°C.

Overall, as demonstrated in Figure 52, both the PA and T6 tempers tend to converge towards the baseline T76 strength and ductility properties for the 200°C forming cycle and paint-bake combinations, although there sometimes still remains a gap in measured properties. It is conceivable that through further study and temperature optimization, a combination of short-duration temperature exposure and paint-bake could allow both the PA and T6 tempers to return to at or the as-received T76 properties.

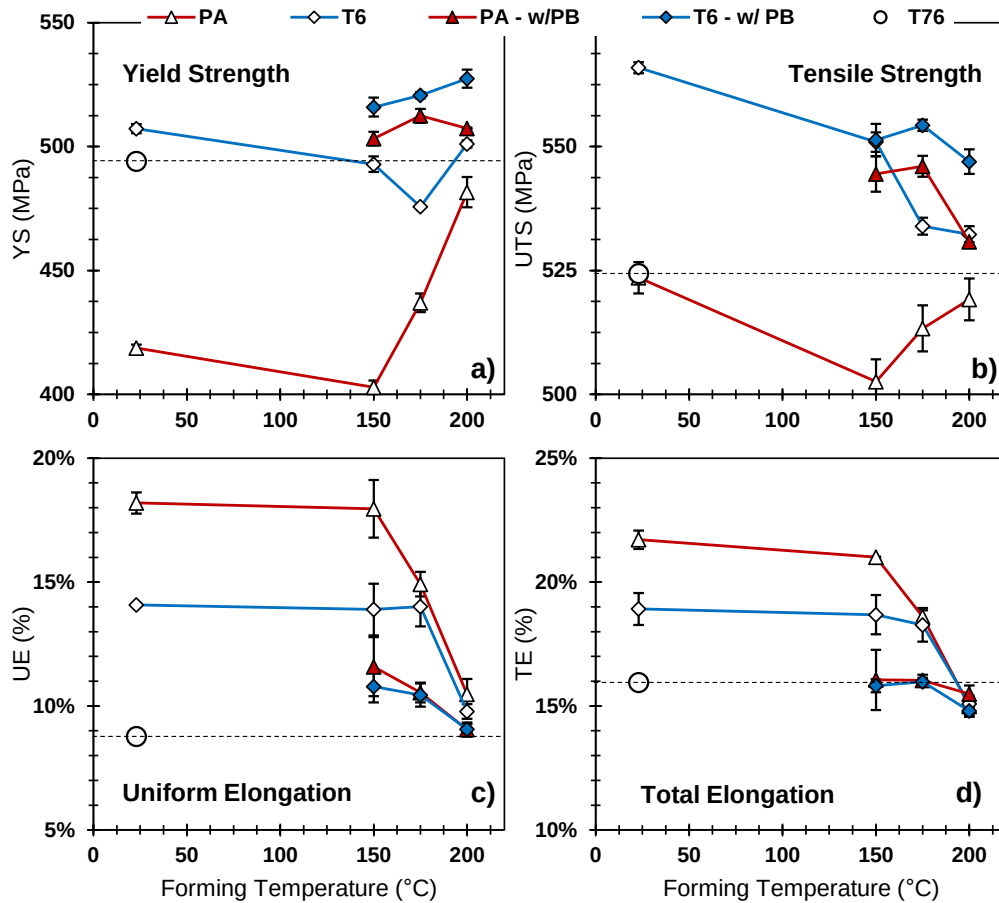


Figure 52: Effect of the warm forming cycle temperature on the tensile properties of the PA and T6 tempers with (solid symbols) and without (open symbols) a paint-bake. a) and b) show trends in yield and tensile strength, respectively. Uniform (c)) and total (d)) elongation capture trends in ductility.

4.4.4. Warm Constitutive Behaviour of the PA and T76 Tempers

The isothermal constitutive response of the PA and T76 tempers at 175°C as a function of hold time at temperature and strain rates are shown in Figure 53. The constitutive response between the two tempers is markedly different, with the PA temper exhibiting conventional work hardening while the T76 specimens begin to diffusely neck almost immediately after yielding. Given the lack of work hardening in the elevated

temperature T76 data, there is effectively no plastic uniform elongation, and yet the T76 temper positively responded to warm forming (Figure 50). The strain rate sensitivity of the T76 temper results in a 7% increase in flow stress for an increase in strain rate from 0.01/s to 0.1/s. It appears that this positive rate sensitivity is the driving force for promoting stability during diffuse necking, leading to the formability improvements in the T76 temper with warm forming. A similar, albeit muted, trend was also observed for the AA6013-T6 in Section 4.2.3.

For the PA condition, a positive rate sensitivity of approximately 6% between the 0.01/s and 0.1/s conditions was observed, independent of hold time and the shape of all tensile curves were relatively consistent, per Figure 53a). At 150°C, the warm formability (ECM-based data) saw a maximum improvement of only 10%. At 175°C, there was no meaningful difference in formability compared to room temperature. In contrast to the T76 temper, the PA temper had strong RT work-hardenability and showed correspondingly good formability. In addition, the PA temper exhibited consistently higher formability than the T76 temper for all temperatures, which is largely attributed to the prominent work hardening in the PA condition. Although not investigated in this work, it is possible that rapid heating of the PA temper for warm forming may enable improvements in formability by preventing a loss in work hardening occurring during pre-heating prior to forming, as explored in the work of Pishyar (2021) for AA7075. This potential improvement was also speculated for the AA6013-PA temper in Section 4.3.

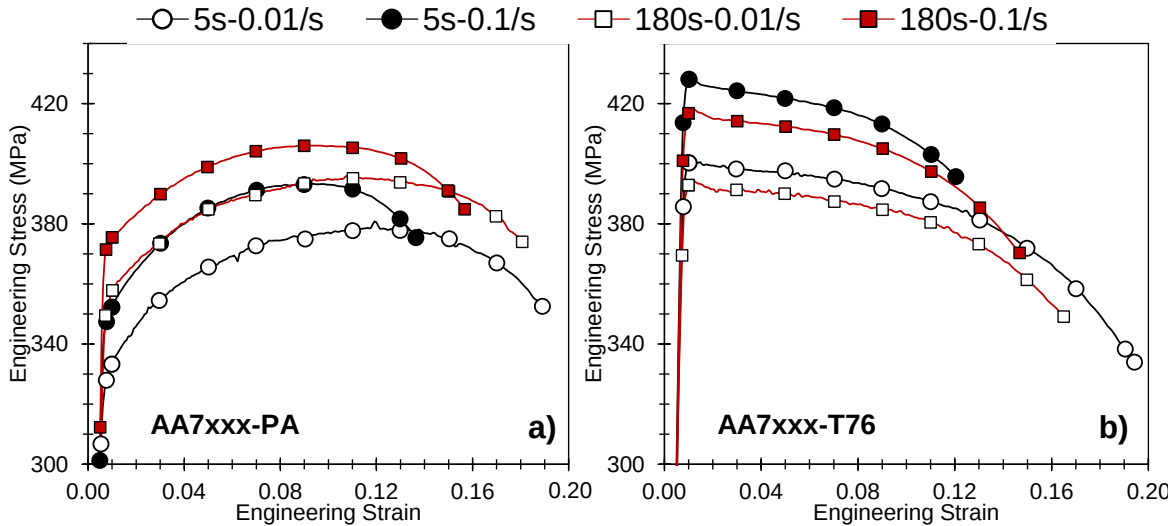


Figure 53: Engineering stress-strain response of the PA (a)) and T76 (b)) tempers at 175°C. Note the T76 temper begins to diffusely neck immediately after yielding in figure b). Conversely, the PA temper shows conventional positive work hardening.

4.5. Limit Strain Characterization in an Aluminum Die Quenching Process

Given the limited gain in formability offered by warm forming of the AA6013 and AA7xxx alloys, the die quenching processing route (Garrett et al., 2005) was considered as an alternative to warm forming. Forming limit experiments using plane strain Nakazima specimens were performed under die quenching conditions on the AA7xxx alloy. This process entailed first solutionizing the samples, after which they were rapidly transferred to the Nakazima tooling and formed immediately. Note that the starting temper prior to solutionizing was T76, which corresponds to the as-received condition for this work. The AA7xxx alloy was selected for this work due to a lower quench sensitivity relative to other common 7000-series alloys, such as AA7075. A conventional die-quench (DQ) processing route was considered in which the tooling was initially at room temperature. An intermediate quench and form processing (IQF) route was also considered that utilized a heated punch. The intention of the IQF work was to characterize the plane strain formability of the alloy in a solutionized state under various quasi-isothermal forming conditions. Full details of this work are given in Appendix E, with selected results and testing procedures presented below.

Following die-quenching, it is expected that with an appropriate heat treatment, the as-received T76 properties could be recovered, as described in the work of Omer et al. (2018).

4.5.1. Experimental Limiting Dome Height Testing Details

Nakazima testing was completed on 76.2 mm wide blanks (Figure 26) in the transverse direction. The testing apparatus was consistent with that used in Sections 4.3, and 4.4; however, the punch and tooling were either RT (DQ) or heated (IDQ). The DQ and IQF thermal processing routes are illustrated in Figure 54. For the DQ processing route, the punch speed was 12.5 mm/s in an effort to replicate a practical die quenching operation. The punch was operated at 2.5 mm/s for the IQF routes and for room temperature testing of the as-received sheet. In the IQF testing, the tooling was maintained at the desired temperature using a closed-loop PID system with in-die heater cartridges.

Strains were measured in-situ using stereoscopic DIC techniques. Given the high temperatures studied, a novel approach to speckling was implemented to allow for DIC analysis without significant degradation in the speckle pattern when exposed to high temperatures and large strains. Typically, temperature resistant paint is used for the background and random speckle pattern in elevated temperature LDH testing. During testing, it was found that temperature resistant paint could not remain adhered at 400°C under large strain, as shown in Figure 55a). To overcome this restriction, a grit-blasting technique was used to generate a sufficiently dull background to provide contrast to the black speckle pattern in place of white background paint. This speckling technique was found to perform sufficiently well at large strains compared to the conventional approach, as shown in Figure 55a). A similar benefit to grit-blasting was reported by Boba et

al. (2015). Other details of the DIC procedures are included in Appendix E, but the methods adopted were generally consistent with other techniques employed throughout this work. The one exception was the use of higher frame rates, 140 fps and 40 fps, for the DQ and IFQ processes, respectively, to account for the higher punch speeds relative to the warm forming experiments.

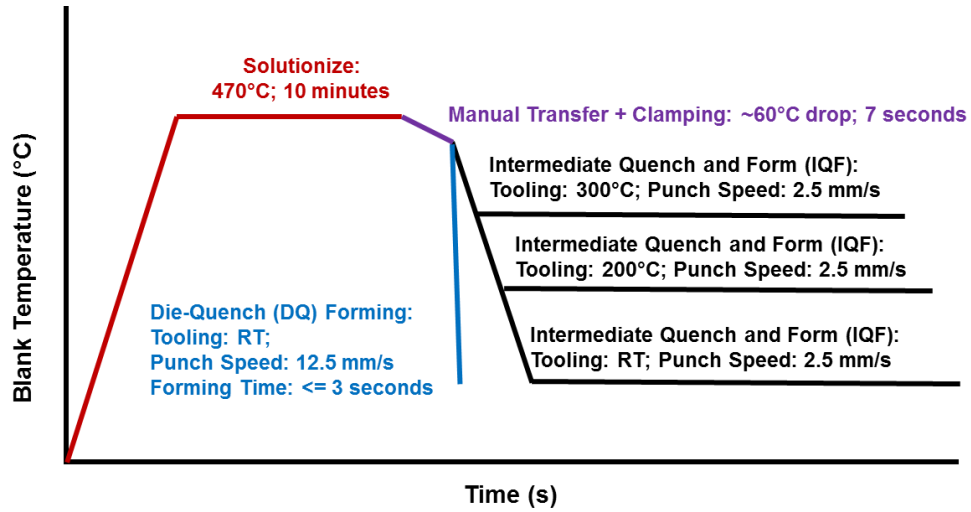


Figure 54: DQ and IQF processing routes for formability assessment.

Despite the improvements in the speckle pattern, some noise was present in the DIC signal at high levels of strain; consequently, the ECM approach could not be used for limit strain detection. Instead, the ISO12004:2-2008 standard method was employed. A time-dependent approach inspired by Martinez-Donaire et al. (2014) was also employed to quantify some necking characteristics, such as the progressive transition from diffuse to localized necking. The approach centers on the observation that different locations along the sample experience peak strain rates (PSR) and strain-rate deceleration at different time intervals, corresponding to diffuse necking, with the exception of locations near the final neck, where the peak strain rate occurs at the sample failure time – consistent with acute necking (Figure 55b)). The full treatment of this approach and results are included in Appendix E.

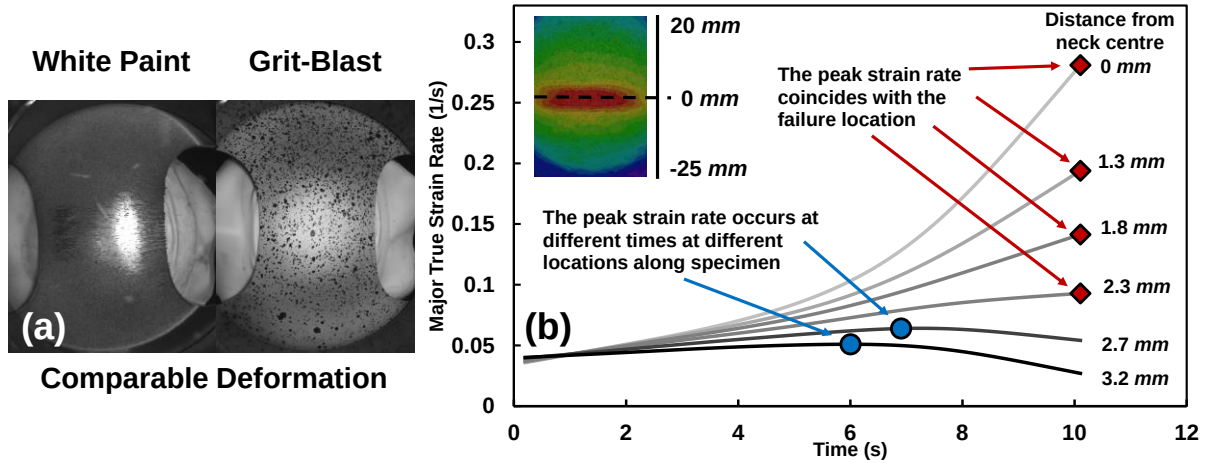


Figure 55: (a) traditional (showing loss of paint) and grit-blast speckle patterns for the DQ process with similar deformation and (b) the time-dependent approach applied to A7xxx showing different PSR times.

4.5.2. Formability Results

The measured limit strains for the DQ and IQF processes are shown Figure 56a), while representative strain-paths are plotted in Figure 56b). Based on the strain paths, it can be seen that the final strain state is near-plane strain, but marginally shifted to the draw-side. The as-received alloy in the T76 temper tested at room temperature demonstrated the lowest formability of all the processing conditions ($\epsilon_1 = 0.12 \pm 0.01$). The AA7xxx alloy tested under the IQF processing route at room temperature had formability of $\epsilon_1 = 0.19 \pm 0.01$ and was approximately 58% better than the alloy in the as-received condition. The IQF process at RT demonstrated significant serrated flow-type deformation (Hertzberg, 1989). This type of deformation is normally attributed to Portevin-Le-Chatelier (PLC) band propagation, as discussed in Section 2.2.1 and Section 4.1, and may have negatively influenced formability. Figure 56c) shows that the PLC effect is initially small but increases with deformation. Omer et al. (2018) reported the same behaviour in tensile tests of the AA7xxx in W-temper, with the severity of PLC effects increasing with deformation.

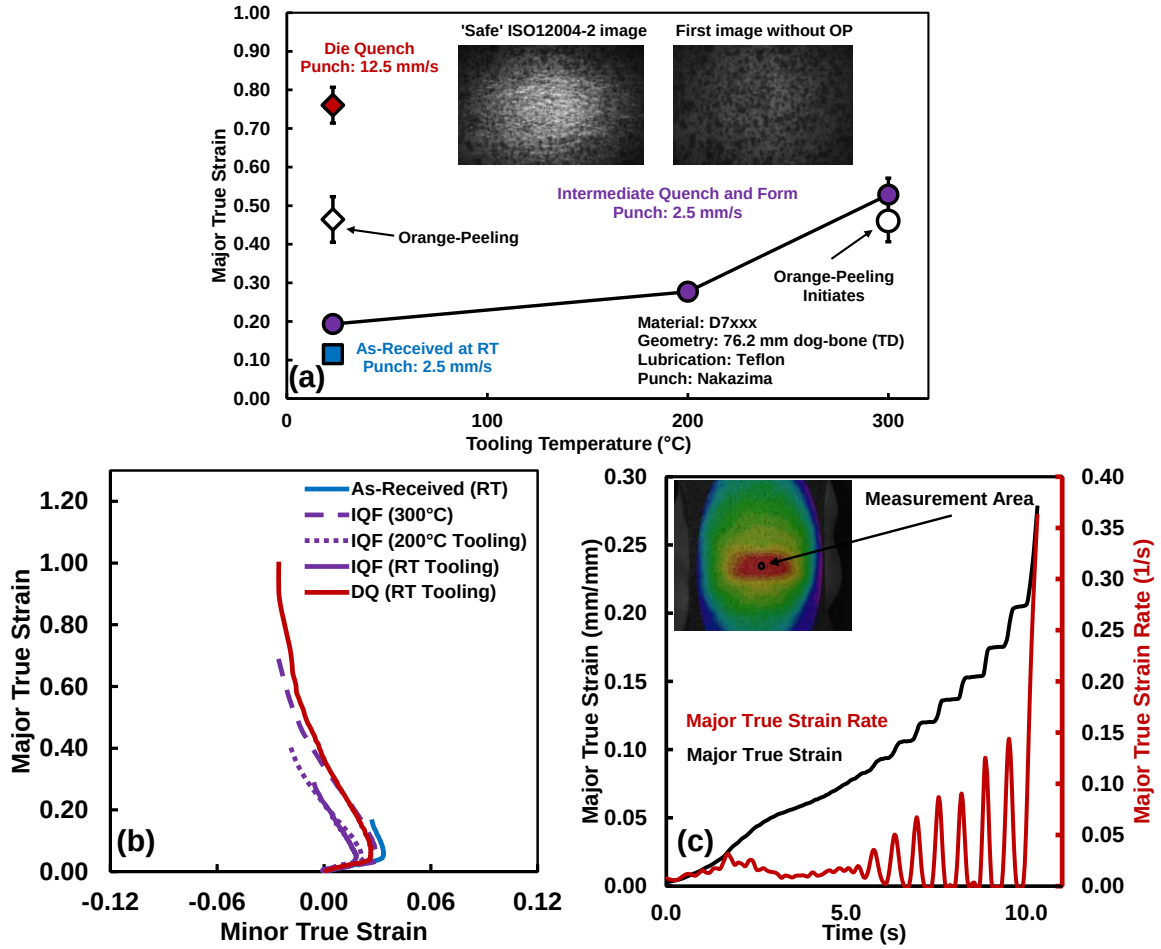


Figure 56: (a) major true limit strains for all processing routes, (b) representative strain paths for all processing routes, and (c) the presence of strain-wave propagation in an IQF specimen tested at room temperature, attributed to PLC effects.

At temperatures of 200°C and 300°C, the major limit strains were 0.28 and 0.53, respectively. The increased formability at 300°C may be attributed to the expected increase in strain-rate sensitivity as a function of temperature. At 300°C, some orange peeling was encountered and likely represents a practical lower bound for formability at that IQF condition. Under DQ processing conditions, the major limit strain was extremely high, with $\varepsilon_1 = 0.76 \pm 0.05$. Under DQ processing, the temperature-history during the forming operation was not measured. Factors that might have contributed to the improved formability in the DQ process may include the non-isothermal forming conditions and the initially high temperatures at which deformation was initiated. Orange peeling was also observed in DQ processing, and, in order to avoid this roughened surface condition, the strain levels should be kept below those of the IQF 300°C tests. Orange peeling is an undesirable defect and may serve as a better practical limit for formability for the DQ and IQF processing routes. Nonetheless, the limit strains in DQ forming are impressive and represent a considerable increase over that achievable in warm forming as characterized in Section 4.4.

5. SUMMARY, CONCLUSIONS, AND RECOMMENDATIONS

This research effort assessed the formability of two precipitation-hardenable aluminum alloys, AA6013 and the developmental 7000-series alloy, AA7xxx. The work spanned a spectrum of elevated temperature forming conditions and starting tempers, with special attention given to the post-forming mechanical properties. A combination of formability and strength would ultimately enable the appropriate selection of the right alloy and process for use in automotive lightweighting applications. A key achievement of this work was the development, implementation, and validation of the Enhanced Curvature Method (ECM), a new formability characterization technique that was applied to support the elevated temperature forming characterization of the alloys and processes studied in this work. A discussion of the major results and findings from Section 4 is presented below, followed by key conclusions and recommendations for future research on the studied topics.

5.1. Summary

The ECM was developed as a physically motivated approach to detect necking by evaluating changes in surface curvature using stereoscopic digital image correlation techniques. The technique was validated at room temperature using AA5182-O sheet, which was selected because it has a wealth of historical forming data for benchmarking purposes and typically presents challenges in formability characterization due to Portevin-Le Chatelier effects. The average Marciniak plane strain limit using the ECM (upper bound) was 0.198, in-line with the theoretical limit of 0.194 and with the historical literature (0.193 ± 0.038). The method was extended to Nakazima and stretch-bend testing where it showed similar trends to the ISO12004-2:2008 standard (2008) and the linear best fit method (Volk & Hora, 2011), despite using fundamentally different necking detection algorithms. Due to the nature of the ECM, validation at room temperature gave confidence in the subsequent extension of the technique to elevated temperature characterization of the AA6013 and AA7xxx alloys for which the approach was initially developed.

Formability characterization of the AA6013 alloy was carried out in two phases. In the first phase, a warm forming limit diagram of AA6013-T6 was developed using a precursor of the ECM approach. The AA6013-T6 alloy was found to only modestly improve in formability, with the maximum improvement at 200°C. A brief numerical model was developed using constitutive data to show that any formability benefits observed at 200°C were from the result of the increased strain-rate sensitivity of the alloy with temperature, offset by a reduction in uniform elongation at elevated temperatures. This trade-off in rate sensitivity and uniform elongation was further observed in the study of warm forming of AA6013 in the pre-aged condition. Furthermore, when this topic was studied further using the ECM applied in Section 4.3, no meaningful benefit to formability was observed in the T6 temper under warm forming.

The warm formability of AA6013-PA was studied in an effort to better understand the effects of warm forming on the alloy after the relatively minor improvements observed for the T6 temper. The pre-aged temper showed a marked improvement in formability at room temperature relative to T6 for a 76.2 mm Nakazima geometry, with average major limit strains of 0.28 and 0.18 observed between the two respective tempers. When the PA temper was studied under warm forming conditions, the forming advantages diminished, and the PA temper was observed to decrease in formability and dome heights at the studied temperatures of 150°C and 230°C. The decrease in formability of the PA temper was severe relative to the T6 temper, which saw a marginal improvement with temperature. The PA temper response to warm forming was attributed to in-situ aging of the alloy. It was found that the temper ages rapidly during the 180 second heating time required for the warm forming experiments; thus, the PA temper was losing work-hardened ability from in-situ aging that could not be compensated by the corresponding increase in rate sensitivity. The importance of controlling in-situ aging effects when warm forming pre-aged tempers to balance formability with final strength properties cannot be overstated from this observation. Aging studies of the PA temper further supported this observation and were used in the development of an aging kinematics strength model for AA6013-PA as a function of aging time at temperature.

A significant outcome of the research on AA6013 was the observation that the strength of AA6013-PA could be restored to near T6 levels after exposure to thermal cycling of less than 410 seconds at 235°C. Although initially conceptualized to explore the effects of warm forming and subsequent aging, the processing route of room temperature forming in the PA condition with a short post-forming aging treatment offers a viable pathway to achieve T6 strength levels with a significant improvement in formability relative to AA6013-T6. This finding addresses the need to identify optimal forming process parameters that balance formability and final strength properties for AA6013. This conclusion is particularly relevant for industrial applications where room temperature forming is more common and cost-effective than warm forming.

The formability of the AA7xxx alloy, much like the AA6013, was characterized as a function of temperature and starting temper using a 76.2 mm dome sample and a Nakazima punch. The strength and ductility of AA7xxx were also measured, with particular attention to the effects of different starting tempers (PA, T6, and T76) and post-forming aging treatments. The AA7xxx alloy in the as-received condition was in the over-aged T76 temper. This temper offers improvements in stress corrosion cracking and was used for baseline warm forming characterization. The over-aged condition, when warm formed at 150°C, showed an improvement in formability from 0.12 to 0.17 major strain as measured with the ECM. Temperatures beyond 150°C did not show meaningful formability improvements. Furthermore, the warm formability of

the T76 temper was less than the measured room temperature limit strains of the PA and T6 tempers, which were 0.21 and 0.18, respectively.

Warm forming at 150°C enhanced the formability of AA7xxx-PA and AA7xxx-T6; however, there were some inconsistencies between measured limit strains and measured dome heights. Additionally, there were differences in the limit strains between the ECM and ISO method, with the latter predicting higher limit strains, despite decreasing dome heights. These trends were similarly observed at 175°C and 200°C. In-situ warm tensile testing of AA7xxx-PA and -T76 was completed at 175°C to understand the warm forming observations in greater detail. It was found that the T76 temper shows higher rate sensitivity at temperature relative to the PA temper but lower work hardening potential. Since the T76 has low formability at room temperature, the strain-rate sensitivity improvement tended to improve the warm forming response. The opposite trend was observed for the PA temper. The loss in hardening at elevated temperatures from in-situ aging was not offset by increases in rate sensitivity for the conditions studied.

Similar to the trends observed for AA6013-PA, exposure to a warm forming cycle at 200°C and a subsequent paint-bake allowed the PA and T6 tempers of AA7xxx to achieve strength and ductility comparable to the as-received T76 temper. Additionally, with sufficient time and temperature exposure post-forming, some anecdotal evidence of over-aging in the PA temper was observed which might improve the stress corrosion cracking response. This result indicates that post-forming aging treatments can be effectively used to tailor the final properties of warm formed AA7xxx components. The T76 temper, while benefiting from warm forming, exhibited lower overall formability than the PA temper across all tested temperatures. This finding suggests that components initially designed for room temperature forming of the T76 temper might experience formability benefits from warm forming, but the PA temper remains the superior choice when high formability is critical.

Comparing the warm forming response of AA6013 and AA7xxx reveals that the alloys saw no benefit and modest benefit, respectively, in formability improvements in their as-received conditions (T6 and T76) during warm forming. The PA temper provided significantly higher formability at room temperature for both alloys, supporting the conclusion that room temperature forming of the PA temper followed by a short thermal cycle to improve final strength is a viable processing route for both AA6013 and AA7xxx. Notably, AA6013-PA experienced a reduction in formability at elevated temperatures due to in-situ aging, while AA7xxx-PA showed some benefits in formability with temperature exposure. This contrast highlights the distinct influence of in-situ aging on different alloy systems and underscores the importance of considering alloy-specific aging behaviour when designing warm forming processes.

Given the limited formability improvements through warm forming (except under certain conditions), the die quenching (DQ) process was investigated as an alternative for AA7xxx using a 76.2 mm blank and Nakazima punch. The AA7xxx die quenching formability results were exceptional, and showed that orange peeling (a surface defect), not necking, was the primary factor limiting formability. The die quenching process resulted in a major limit strain of 0.76, but orange peeling occurred closer to 0.45 major strain. Under an interrupted, isothermal (300°C) hot-stamping operation, a major limit strain of 0.53 was observed, whereas at 200°C it was 0.28. These limit strains were significantly higher than the room temperature formability of 0.12 measured for the as-received T76 temper and that of 0.17 achieved using warm forming.

The ECM, while effective for warm forming, encountered challenges in characterizing formability during the DQ process due to high temperatures affecting local DIC measurements as well as the presence of orange-peeling. This limitation necessitated the use of the ISO 12004:2-2008 standard, which is more resilient to local DIC and surface quality issues. This finding emphasizes the need for robust measurement techniques capable of handling extreme conditions inherent in die quenching, a key area for future research.

Figure 57 summarizes key major limit strain trends observed over the course of this research for select forming technologies, alloys, and tempers. All results are for the 76.2 mm dog-bone geometry using a Nakazima punch and were calculated with the ECM, with the exception of the DQ result. Tensile strength for each alloy is also shown. These strengths correspond to the measured as-received tensile strength of each studied alloy which are expected to be achievable for the highlighted forming processes based on the results of discussed in this research and Omer et al. (2020). The AA6013-PA temper formed at room temperature compares favourably to the formability of AA5182-O while achieving superior strength. AA7xxx-PA has formability in-line with that measured of AA5182-O, albeit less than AA6013-PA; however, it offers a significantly higher as-received strength. The AA7xxx alloy formability can be further improved through the die-quench process and can more than double the measured formability of AA5182-O and was only limited by orange-peeling defects in the current experiments.

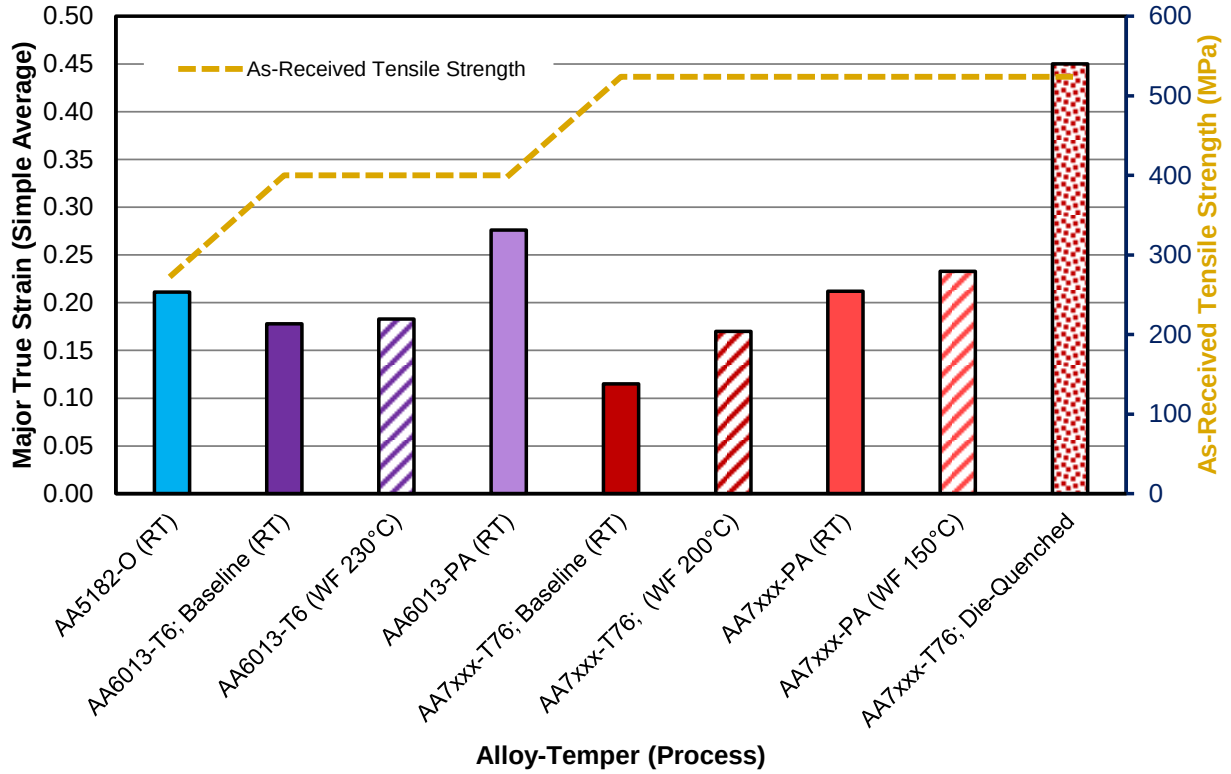


Figure 57: Select forming limit and corresponding as-received tensile strengths of AA5182-O, AA6013, and AA7xxx as a function of temper and starting temperature. All results correspond to testing using a Nakazima punch, a 76.2 mm dog-bone sample, and ECM analysis. Tensile strength is plotted on the secondary axis. Strength data from this research and Omer et al. (2020) suggest appropriate processing parameters can achieve comparable final strength to that observed in the as-received condition. Formability results suggest that AA6013-PA at room temperature offers excellent formability and strength with a short duration post-forming aging cycle and warm forming did not improve AA6013 formability in this research. The AA7xxx-PA alloy shows significant improvement over the baseline T76 formability and can achieve T76 strength. AA7xxx formability can be further improved via die-quenching, up to approximately 0.45 major strain before surface defects may be observed.

5.2. Conclusions

The primary conclusions that stem from the present research are summarized below.

The Enhanced Curvature Method (ECM) is a reliable and physically motivated technique for characterizing formability leveraging the detection and growth of defects observed during necking. Its development was crucial for quantifying formability under elevated temperature forming conditions. One significant contribution of the current research is in the highly descriptive documentation regarding implementation of the ECM process such that any user should be able to implement the ECM, as described in Appendix A. Of note, employing the ECM for large-deformation problems underscores the necessity for robust Digital Image Correlation conditions, as observed during die-quench formability characterization.

Initial temper significantly influences formability, with the pre-aged (PA) temper demonstrating superior formability at room temperature for both AA6013 and AA7xxx alloys. For AA6013, the PA temper showed a 55% improvement in limit strain compared to the T6 temper at room temperature, with average measured limit strains of 0.28 and 0.18, respectively. AA7xxx-PA had a room temperature limit strain of approximately 0.21, compared to 0.18 for AA7xxx-T6. Balancing in-situ aging effects with desired final strength and formability requires careful process control during warm forming of pre-aged tempers. The current research found that, for the warm forming conditions studied, the benefits of warm forming were modest and mostly limited to the AA7xxx alloy in the overaged condition, with no benefit to warm forming observed for AA6013-T6 when using the ECM for analysis.

Room temperature forming of the PA temper, followed by a short post-stamping thermal aging cycle, is a viable and industrially relevant processing route for achieving high strength and formability in AA6013 and AA7xxx. For AA6013, a relatively short secondary aging cycle (less than 410 s) at 235°C after forming, combined with a pre-aged initial temper, could result in a stress-strain response similar to the baseline T6 temper, regardless of a subsequent paint-bake. For AA7xxx, data suggests a similar strength to the baseline T76 overaged condition could be achieved leveraging cold forming of PA with a secondary aging cycle; however, further work is required to establish those processing parameters.

Die quenching (DQ) offers a promising method for achieving substantially higher formability in AA7xxx alloys. Major limit strains of $\epsilon_1 = 0.76 \pm 0.05$ were achieved under DQ processing conditions. However, further research is necessary to mitigate surface defects such as orange peel and optimize processing parameters. For example, orange peel was observed for strain levels in excess of approximately 0.45 major true strain. This points to future research directions where extreme formability requirements may be required.

Overall, the current findings contribute valuable insights into the formability behavior of AA6013 and AA7xxx. The development and validation of the ECM, coupled with the comprehensive characterization of warm formability and die quenching responses, provide a foundation for developing optimized forming processes for these alloys. These advancements will enable the expanded use of AA6013 and AA7xxx in lightweighting applications in the automotive and other industries.

5.3. Recommendations for Future Work

Based on the conclusions of this work, recommendations for future research are given.

ECM Standardization: Future work should further standardize the ECM and encourage its use within the formability community with benchmarking to a host of sheet aluminums as well as advanced high strength steels to demonstrate its applicability.

This research developed a robust framework for the ECM, including procedures for implementation and a threshold/necking tolerance metric based on an imperfection factor, f_c . Including imperfection factor plots (e.g. Figure 31) for various strain paths in research reports would empower end-users to have better control over their FLC margin and make more informed decisions regarding acceptable strain levels for specific forming operations. This insight would enhance the practical utility of the ECM and provide a more nuanced understanding of material formability. A formalized standard for the ECM, similar to the ISO12004-2:2008 standard, would promote its widespread use and contribute to more reliable and consistent formability characterization.

Rapid Heating of Pre-Aged Tempers: This research showed that pre-aged (PA) tempers can improve room temperature formability; however, in-situ aging during warm forming can decrease formability, particularly for AA6013-PA. Future research should investigate rapid heating to mitigate the negative impacts of in-situ aging by limiting precipitate growth during heating. Rapid heating could maximize the formability benefits of PA tempers in warm forming, as shown in the work of Pishyar (2021).

Optimization of Die Quenching: This research showed that die quenching (DQ) substantially improved the formability of AA7xxx, but may increase production costs and extend cycle times. Future research on DQ should explore the sensitivity of various stamping parameters to identify optimal processing windows for achieving desired strength and formability while minimizing defects. Developing a more industry-friendly DQ process is crucial for realizing its full potential in high-volume manufacturing settings.

Post-Stamping Process Optimization: In both the warm forming and die quenching processes, the effect of time and temperature on final strength and ductility properties was explored. Future work should entail further optimization of these properties in conjunction with the recommendations of rapid heating and die quench optimization. Notably, the development of more extensive processing maps is critical for adoption of the aforementioned tempers and forming technologies in an industrial environment. Additional focus should be given to the final stress corrosion resistance of the AA7xxx alloy, a topic that was not within the scope of the current research.

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APPENDIX A: FORMABILITY CHARACTERIZATION USING CURVATURE AND STRAIN-RATE BASED LIMIT STRAIN DETECTION METHODS APPLIED TO MARCINIAK, NAKAZIMA, AND STRETCH-BEND TESTS

DiCecco, S. *et al.* (2024) 'Formability characterization using curvature and Strain-Rate-Based limit strain detection methods applied to Marciniak, Nakazima, and Stretch-Bend tests,' *Metals*, 14(10), p. 1164. <https://doi.org/10.3390/met14101164>.

APPENDIX B: NUMERICAL AND EXPERIMENTAL INVESTIGATION OF THE FORMABILITY OF AA6013-T6

DiCecco, S. *et al.* (2017) 'Numerical and experimental investigation of the formability of AA6013-T6,' *Journal of Physics Conference Series*, 896, p. 012114. <https://doi.org/10.1088/1742-6596/896/1/012114>.

APPENDIX C: WARM AGING OF PRE-AGED AA6013 SHEET AND ITS RELEVANCE TO ROOM TEMPERATURE AND WARM FORMING APPLICATIONS – EXPERIMENTAL AND MODELING ANALYSES

DiCecco, S. *et al.* (2023) 'Warm aging of pre-aged AA6013 sheet and its relevance to room temperature and warm forming Applications—Experimental and Modeling Analyses,' *Journal of Materials Engineering and Performance*, 32(21), pp. 9797–9813. <https://doi.org/10.1007/s11665-023-08216-6>.

APPENDIX D: EFFECT OF INITIAL TEMPER ON THE WARM FORMING CHARACTERISTICS OF A HIGH STRENGTH 7000-SERIES AL-ZN-MG-CU ALLOY

DiCecco, S. *et al.* (2022) 'Effect of initial temper on the warm forming characteristics of a high strength 7000-series AL-ZN-MG-CU alloy,' *IOP Conference Series Materials Science and Engineering*, 1238(1), p. 012087. <https://doi.org/10.1088/1757-899x/1238/1/012087>.

APPENDIX E: LIMIT STRAIN CHARACTERIZATION IN AN ALUMINUM DIE QUENCHING PROCESS

DiCecco, S. *et al.* (2018) 'Limit strain characterization in an aluminum Die-Quenching process,' *IOP Conference Series Materials Science and Engineering*, 418, p. 012035. <https://doi.org/10.1088/1757-899x/418/1/012035>.