The application of rotary forming to A356 offers a potential improvement in material use, simplified castings and ameliorated fatigue resistance. To investigate the utility of adopting this process industrially, an extensive characterization and modelling effort was undertaken.

The constitutive behaviour of A356 in the as-cast condition was assessed with compression tests performed over a range of deformation temperatures (30-500°C) and strain rates (~0.1-10 s⁻¹). The flow stress as a function of temperature and strain rate was quantified via an extended Ludwik-Hollomon and Kocks-Mecking framework.

The through-process microstructural effects on A356 subjected to rotary forming at elevated temperatures was also investigated. This was conducted on material at 350°C with an industrially-scaled, purpose-built apparatus, inducing varying levels of spinning deformation. This was also conducted on commercially flow formed material with high levels of deformation at the same temperature. Macro and micro-hardness testing was used to track the changes from the as-cast and as-formed states, as well as following a T6 heat treatment. Further EDX analysis indicate that precipitation aspects of heat treatment is not appreciably affected by forming. Forming was found to principally affect the eutectic-Si particle size, resulting in a finer particle post heat treatment.

An explicit finite element rotary forming model reciprocating experimental forming conditions was developed incorporating the Ludwik-Hollomon description. This forming model was found to be computationally expensive; however, demonstrated reasonable agreement with experimental geometry and phenomena.

In evaluating the effect of forming on fatigue, multiaxial testing of A356-T6 was conducted to apprehend the basic fatigue mechanisms. Endurance limits are found to be generally governed by porosity and maximum principal stress for high cycle fatigue. Uniaxial fatigue tests of both experimentally and commercially formed material showed a 30% increase in endurance limits over unformed material, principally through mitigating porosity.
The following journal submissions have been extracted from the body of work presented in this dissertation. My supervisor, Prof. Daan Maijer provided experimental insight, results interpretation and editorial support covering all aspects of my research. Aside from my supervisor, and key secondary contributors, I am the primary contributor to these works:


Prof. Yves Nadot provided experimental insight, supported editorially and in interpretation of results for items 1 and 4. Guillaume Benoit provided technical assistance with the experiments for item 1. Louise Dancoine provide technical assistance in completing the experiments and results interpretation for item 3. Prof. Carole Nadot-Martin and Pierre-Guillaume Bardin provided results interpretation for item 4.

Chapter 2 contains material from items 1, 3 and 4. Chapter 3 is based on material drawn from item 3. Chapter 5 contains material from item 2. Chapter 6 contains material drawn from item 1 and 4. These chapters contain footnotes mirroring the above information.
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# NOMENCLATURE

**Roman Symbols** *(page introduced)*

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$g$ Normalized activation energy (page 69)

$g_0$ Initial normalized activation energy (page 69)

$h$ Heat transfer coefficient (page 111) \[ \text{W/m}^2\text{°C} \]

$H_{HB}$ Brinell hardness (page 12) \[ \text{kg/mm}^2 \]

$H_{RF}$ Rockwell hardness, F scale (page 13)

$H_{V0.01}$ Vickers hardness, 0.01 kg$_f$ (page 55) \[ \text{kg/mm}^2 \]

$H_{V5}$ Vickers hardness, 5 kg$_f$ (page 38) \[ \text{kg/mm}^2 \]

$J_{2,a}$ Second invariant of the deviatoric stress amplitude (page 139) \[ \text{MPa}^2 \]

$k_c$ Thermal conductivity (page 104) \[ \text{W/m}^\circ\text{C} \]

$k_{1-3}$ Extended Ludwik-Hollomon strength parameter function coefficients (page 68)

$k_{LSW}$ LSW temperature-dependent constant (page 89)

$k_B$ Boltzmann constant, $1.3806503 \times 10^{-23}$ (page 69) \[ \text{m}^2\text{kg/s}^2\text{K} \]

$K_{LH}$ Extended Ludwik-Hollomon strength parameter (page 68) \[ \text{MPa} \]

$L$ Element edge length (page 100) \[ \text{mm} \]

$m$ Population mean (page 90)

$m_{1-3}$ Extended Ludwik-Hollomon strain rate function coefficients (page 68)

$N$ Fatigue cycles (page 136)

$n$ Mechanical property to DAS relationship coefficient (page 9)

$N_f$ Number of cycles to failure (page 16)

$n_{1-3}$ Extended Ludwik-Hollomon strain hardening function coefficients (page 67)

$P$ Probability (page 90)

$P$ Radial penetration of roller into workpiece employed in model (page 50) \[ \text{mm} \]

$p$ Dislocation interaction parameter (page 69)

$q$ Dislocation interaction parameter (page 69)

$Q_c$ Cooling rate (page 4) \[ \text{°C/s} \]

$q_H$ Heat flux (page 111) \[ \text{W/m}^2 \]

$q_{c,f}$ Heat fluxes applied to model workpiece cooling (page 108) \[ \text{W/m}^2 \]

$R$ Radial direction of forming (page 20)

$R_L$ Load ratio for cyclic loading, $\sigma_{max}/\sigma_{min}$ (page 16)
Experimental strain rate ranges (page 61)

Standard deviation (page 90)

Temperature (page 15) °C

Time (page 86) s

Reference temperature for thermal expansion (page 104) °C

Workpiece temperature employed in model (page 106) °C

Mandrel temperature employed in model (page 106) °C

Simulated process time (page 108) s

Extended Ludwik-Hollomon strain rate transition temperature (page 67) °C

Sink temperature (page 111) °C

Liquidus temperature (page 63) K

Axial/radial roller nose positions employed in model (page 108) mm

Axial direction of forming (page 20)

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<tr>
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<td>Low Pressure Die Casting</td>
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<td>Equivalent Circle Diameter</td>
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<tr>
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<td>Probability Density Function</td>
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“Was der Vater schwieg, das kommt im Sohne zum Reden; und oft fand ich den Sohn als des Vaters entblößtes Geheimniss.”
—F. W. Nietzsche, Also Sprach Zarathustra
CHAPTER 1

INTRODUCTION

Particularly in the transportation sector, modern manufacturing is constantly evolving to meet increasing demands for higher performance, lighter component weights and reduced ecological impact, all with lower costs. Classical near net-shape manufacturing processes for metallic components such as casting and forging have remained central in addressing these demands. Forging often improves the material properties by imparting work hardening and leaving compressive residual stress on the surface of a component, however, it is cost prohibitive for production of small numbers of parts. The advent of numerically controlled machining centres has diminished the cost of these types of components. However, subtractive manufacturing processes, such as machining, can be very inefficient in the use of material. Currently, the cost of material waste in machining is offset by the lack of specialized, component-specific tooling required for short production runs.

Rotary forming techniques, such as spinning, shear forming and flow forming have the ability to produce near net-shape components that optimize material use. These techniques use open dies, or tools, to generate highly localized plasticity to achieve the final shape of the part. By mounting a workpiece on a rotating mandrel, incremental deformation is applied by the tool in the same way a potter shapes clay. This produces dimensionally accurate, axially symmetric components. The benefit of these rotary forming techniques is that they may impart the same benefits as forging, albeit with decreased forming loads and more adaptive tooling. As a result, rotary forming has the potential to be either a stand-alone, disruptive technology or to compliment standard metal turning technology.

One of the novel applications of rotary forming is aluminum alloy wheel fabrication. The most cost effective manufacturing technique for these wheels is through Low Pressure Die Casting (LPDC) employing an Al-Si-Mg casting alloy such as A356. These wheel castings are then heat
treated and then machined to the final size. The marriage of flow forming to standard casting processes permits a wider range of possibilities than by casting alone. This includes casting the front face of the wheel including the spokes and hub with LPDC, and then flow forming the remainder of the rim [2]. Wheels manufactured in such a manner have the potential to be lighter due to improved properties. Further weight savings, up to 20% over a conventionally cast wheel, can be realized by harnessing a flow forming operation to close hollow spokes [3]. However, rotary forming of aluminum alloy wheels remains a somewhat expensive operation owed to high initial costs as operational parameters are arrived at by trial and error. As such, it has been limited in application and successful operating parameters are that are arrived at by trial and error are closely guarded trade secrets. The costs associated with the implementation of rotary forming are further exacerbated by the necessity to carry out rotary forming of aluminum castings at elevated temperatures.

Compared to other material processing technologies such as rolling or extrusion, there has been limited academic study of rotary forming processes. This is especially true regarding studies on the application of rotary forming of aluminum castings. There also is limited data available on the flow stress behaviour of as-cast A356 at elevated temperatures. Therefore, it is not specifically known what impact rotary forming at elevated temperatures will have on mechanical properties, nor on the in-service fatigue performance of components processed in this manner. Moreover, the multiaxial fatigue resilience of this material processed in a standard manner, i.e. processing paths excluding rotary forming, has not been previously characterized. There are multiple opportunities for contributions to improved understanding of this industrially relevant material for a wide range of applications, however, multiaxial fatigue characterization is particularly pertinent for automotive wheel manufacturers. This is owed to the significant in-service multiaxial cyclic loading these components endure, outside of regions of the wheels that can be potentially processed by rotary forming.

The following section provides a description of Al-Si-Mg casting alloy structure. Microstructural modifications induced by heat treatment will be discussed, and the overall impact on mechanical properties is then presented to appreciate the potential effects of rotary forming on the alloy.
1.1 Structure of Al-Si-Mg casting alloys

Aluminum alloy A356 is a hypoeutectic Al-Si-Mg foundry alloy with an as-cast (AC) microstructure that consists primarily of aluminum dendrites (α-Al), surrounded by an Al-Si eutectic (Fig. 1.1). Other tertiary phases, such as α-intermetallics or β-intermetallics (β-Al₅FeSi and π-Al₈FeMg₃Si₆), may be present due to melt impurities. The β-intermetallics, appearing as elongated needles within the microstructure, are deleterious to mechanical properties. The solidification sequence of this alloy starts with the nucleation and growth of primary dendritic α-Al. α-intermetallics may form at this point if Mn and Cr are present [4, 5]. This is followed by the β-Al-Si eutectic and β-intermetallics. The remaining liquid, enriched in Si, Mg and Fe, forms Mg₂Si precipitates and engages in complicated ternary and quaternary reactions, producing π-intermetallics [6].

![Figure 1.1: A356 microstructure in the as-cast (AC) condition displaying DAS, (a) α-Al, (b) eutectic, (c) intermetallic and (d) a secondary Mg-Si rich region.](image)

Overall, the Al-Si-Mg alloy system offers excellent castability, reasonable strength and adequate fatigue resistance. The castability is primarily owed to the ~ 7%-wt Si content that enhances the fluidity of the melt. Mechanical properties, including strength and fatigue resistance are influenced by grain/structure refinement, chemical modification and heat treatment. Alloying additions, such as Al-Ti-B, are used to refine grain size and further improve castability [7]. Chemical modification, achieved through the addition of small amounts of Na and Sr to the melt, changes the morphology
CHAPTER 1. INTRODUCTION

of the eutectic-Si from acicular to more fibrous and refined structures. A356 with additions of Na and/or Sr is referred to as ‘modified’ and ‘unmodified’ without. Modified A356 permits eutectic particles which are both smaller and more spherical after heat treatment [8]. These alloys are rarely employed in the AC condition owing to a lack of homogeneity and detrimentally coarse plates of Si present in the eutectic. Particularly if β-intermetallics form, high levels of Fe are also deleterious. Several heat treatment schedules are commercially employed, with the most prominent being T6. Most of these schedules consist of solutionizing, water quenching and then a combination of natural and artificial aging. Both the duration and temperature at which these treatments are carried out decide the final mechanical properties.

Microstructure refinement, which results in a corresponding strength increase, can be achieved by decreasing the solidification time during casting. By decreasing solidification time, the cooling rate during solidification is increased which results in decreased primary and secondary Dendrite Arm Spacing (DAS). Fig. 1.2a compares the DAS with cooling rate data provided by various sources for A356 and shows excellent correlation. A reduction in DAS is directly related to decreased grain size in the material. Many wrought alloys have been found to have mechanical properties correlated to grain size, due to grain-boundary strengthening according to the Hall-Petch relationship. However, the relatively large and irregular grain structure that is a characteristic of foundry alloys means that it is non-trivial to relate strength and cooling rate (Fig. 1.2b).

Therefore, the microstructural feature typically used to compare the final mechanical properties of Al-Si-Mg alloys is secondary DAS [7, 12, 13]. Directional solidification in commercial castings can produce two principal types of dendritic structure. Columnar dendritic growth is stabilized by high cooling gradients, and has a microstructure exhibiting well-defined primary dendrites. Equiaxed structure occurs where gradients are less pronounced, with primary arms often difficult to identify in particularly coarse material. While tertiary spacing is occasionally used for very coarse material [14], secondary DAS is a relatively facile measurement that spans both types of dendritic structures. For the remainder of this thesis, DAS or \( \lambda \) refers to secondary dendrite arm spacing as shown in Fig. 1.1.

With DAS typically on the order of 10-100 µm, depending on the solidification rate, there are smaller microstructural features affecting strength. After heat treatment, the typical eutectic-Si particle size is approximately one order of magnitude less than this, and Mg\(_2\)Si precipitate occur

throughout the microstructure on the nanometer scale. Heat treatments applied to these alloys serve to primarily modify the eutectic structure and refine/redistribute the Mg$_2$Si particles. In general, compared to the AC condition, the T6 treatment optimizes strength and ductility and is one of the more commercially common tempers for modified Al-7Si-0.3Mg (A356) alloy.

1.2 Heat treatment

According to the current ASTM standard, the T6 heat treated condition involves solution treatment at 540°C for 4-12 hours, quenching in water between 65-100°C, followed by artificial aging (precipitation treatment) at 155°C for 2-5 hours [15]. The solution treatment time may be reduced as required if the melt contains modifiers. The time spent at ambient conditions after quenching, referred to as natural ageing, should be minimized as it reduces the precipitation driving force available for effective artificial ageing. While the current standard does not explicitly mention natural aging, in previous versions of the same standard, allowances of up to 8 hours at room temperature were permitted between the water quench and artificial aging.
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1.2.1 Solution treatment

The solution treatment is applied to induce three phenomena to occur: dissolution of Mg$_2$Si particles, chemical homogenization and eutectic-Si structure modification. The Mg$_2$Si precipitate that forms during the last stages of solidification is readily soluble in $\alpha$-Al at the typical solutionizing temperatures, as indicated by the phase diagram in Fig. 1.3 and will dissolve given enough time. In order to maximize the amount of Mg and Si in solution, a solutionizing temperature as close as possible to the equilibrium eutectic temperature is desirable. A temperature of 540°C is high enough such that incipient melting is avoided at the grain boundaries, which can lead to permanent mechanical strength reduction.

In the AC state, solute elements are typically highly segregated due to dendrite formation. Solution treatment serves to chemically homogenize the casting, thereby improving solid solution strengthening. Closset et al. [17] conducted micro resistivity measurements on dendrites in A356 in the AC condition and after varying solution times. In the AC condition, it was found that Si content was highest at the centre of the dendrite while the Mg content was marginally higher on the edge of the dendrite compared to the centre. The dendritic distribution of Mg in the AC condition was also confirmed by Colley [18] for an Al-Si-Mg alloy with a near eutectic composition of Al-11Si-0.22Mg. Closset et al. reported that homogenization was complete after 30 minutes for A356, while Colley reported 3 hours for Al-11Si-0.22Mg. This suggests that while the presence of large amounts of Si may influence the distribution of Mg, homogenization of the primary alloying elements is complete after 3 hours at 540°C regardless of Si content in A356. Colley’s observations

Figure 1.3: Mg$_2$Si in $\alpha$-Al pseudo-binary phase diagram [16]
underscore this point by providing a limiting case. However, Fe-rich intermetallics are quite stable at typical solutionizing temperatures and take relatively longer to dissolve, occasionally remaining after heat treatment [6, 19].

The changes to the eutectic-Si structure imparted by solution treatment also play an important role in determining the final mechanical properties. While modified Al-Si-Mg alloys contain fairly refined fibrous eutectic-Si, this is further refined during solution treatment by the processes of fragmentation and spheroidization. The $\alpha$ fibres break into particles at elevated temperature and gradually spheroidize in order to minimize surface energy of the Al-Si interface. With longer treatment times, coarsening occurs. Larger Si particles develop facets and coalesce with other nearby particles to minimize surface area in regions of high Si concentration [20, 21].

### 1.2.2 Quenching

For the quench operation, the water temperature is selected to maximize cooling rate while concurrently limiting thermal stress development. A high cooling rate is necessary to suppress precipitation when cooling from the solution treatment temperature to room temperature. This produces a high degree of solute supersaturation as well as retaining a larger number of matrix vacancies. If the cooling rate is too slow, non-uniform precipitation will occur, localized at grain boundaries or sites of high dislocation density. The resulting decrease in supersaturation reduces the maximum yield strength that may be achieved by ageing [22]. The critical temperature range where high cooling rates are required is between 450 and 200°C for most Al alloys and the time spent in this temperature range during quenching should therefore be as short as possible to avoid preemptive precipitation [23].

### 1.2.3 Artificial ageing

Artificial ageing is a precipitation heat treatment process. It consists of taking previously solution treated components and holding at a static temperature for a period of time. The process is necessary to precipitate small particles coherent with the surrounding matrix which are finely dispersed particles to resist dislocation glide. In Al-Si-Mg alloys, the supersaturated solid solution ($\alpha_{SSS}$) resulting from the solutionizing process transforms to a stable phase plus a metastable precipitate phase, $\beta$. The rate of precipitation, as well as the precipitate morphology, is dependent on temperature, time,
degree of supersaturation and diffusivity. At high temperatures, diffusion occurs rapidly even when supersaturation is low. The inverse is true for low temperatures.

The precipitation sequence of $\text{Mg}_2\text{Si}$ specific to an Al-$\text{Mg}_2\text{Si}$ system has been described by Edwards et al. [24]. While A356 contains a significantly elevated level of Si, it is assumed that the principal sequence of precipitation in A356 is only marginally affected from that described by Edwards et al. according to:

\[
\begin{align*}
\alpha_{\text{SSS}} & \rightarrow \text{Mg clusters} + \text{Si clusters} & \rightarrow \text{Dissolution of Mg clusters} & \rightarrow \text{Formation of Mg-Si co-clusters} \\
\text{Equilibrium } \beta^\prime & (\text{Mg}_2\text{Si platelets}) & \rightarrow \text{Metastable } \beta^\prime (\text{Mg}_2\text{Si rods}) & \rightarrow \text{Metastable } \beta^\prime (\text{Mg}_5\text{Si}_6 \text{ needles}) & \rightarrow \text{Small, equiaxed precipitates}
\end{align*}
\]

The maximum strength due to precipitation hardening arises when the particle spacing is small. Edwards et al. have identified that the particle morphology coinciding with peak ageing is the $\beta^\prime$ precipitate, having a nanometer scale, needle-like structure. More recent research by Hasting et al. [25] have shown that the composition of $\beta^\prime$ is not just limited to Mg and Si, but may also contain up to 20 atomic % aluminum. Further ageing generates equilibrium $\text{Mg}_2\text{Si } \beta$ platelets or metastable $\beta^\prime$ rods and decreases strength, coinciding with an over-aged condition. Colley [18] found that the peak aged condition was reached after approximately 1 hour at 200°C, 3 hours at 180°C or 8 hours at 150°C. The time to reach an over-aged condition was found to diminish as the ageing temperature increased.

### 1.3 Mechanical properties of Al-Si-Mg alloys

To this point, the structure of Al-Si-Mg alloys has been discussed, along with the aspects of typical heat treatment. This section will discuss the impact of structure on mechanical properties. This will be accomplished by tabulating the results of several studies on Al-Si-Mg alloys with variations in composition and casting methodology to demonstrate the relative impact of structure as characterized by DAS on mechanical properties. These properties include yield strength ($\sigma_y$), ultimate tensile strength ($\sigma_u$), strain to fracture ($\epsilon_f$), as well as hardness. This will be followed by a discussion of the behaviour of this material at elevated temperatures. This is necessary to evaluate the potential implications for rotary forming on A356.
1.3.1 Effect of DAS

Five studies have been summarized in Table 1.1 and their resulting mechanical properties are plotted versus DAS in Figure 1.4. These studies span a variety of modified versus unmodified alloys over a wide range of DAS with varying heat treated conditions. These results are principally for A356, but also include that of A357, a high strength variant of A356 containing a high amount of Mg. Cast Al-Si-Mg alloys often have property-DAS relationships expressed in terms of a power law equation [26] having the form of:

\[
\sigma_y, \sigma_u, \epsilon_f = a + b\lambda^n
\]

where \(a\), \(b\) and \(n\) are power law coefficients and \(\lambda\) is DAS. Data corresponding to the T6 and AC conditions have been grouped and a non-linear least squares fitting technique was used to generate the power law coefficients for Eq. [1.1], which are presented in Table 1.2. For all studies, the overall trends indicate that all properties diminish with increased DAS in both the AC and T6 conditions. There is significant variability in the degree to which the fitted expressions represent the data, quantified principally by Root Mean Square Error (RMSE). The expressions for material in the T6 condition show an order of magnitude higher RMSE as compared to those for material in the AC condition. The larger variance is attributed to differences in the heat treatment and chemistry amongst the different studies.

Table 1.1: Various literature compositions and T6 solutionizing (Sol.), quench (Q) and artificial ageing (AA) schedules employed for mechanical property studies.

<table>
<thead>
<tr>
<th>Reference</th>
<th>Composition (wt%)</th>
<th>T6 Schedule (°C/hours)</th>
<th>Fig. 1.4 series</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Si</td>
<td>Mg</td>
<td>Fe</td>
</tr>
<tr>
<td>Ran et al. [27]</td>
<td>7.04</td>
<td>0.3</td>
<td>0.17</td>
</tr>
<tr>
<td>Wang [28]</td>
<td>7.0</td>
<td>0.41</td>
<td>0.14</td>
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<tr>
<td></td>
<td>6.8</td>
<td>0.39</td>
<td>0.13</td>
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<tr>
<td>Ceschini et al. [29]</td>
<td>7.0</td>
<td>0.6</td>
<td>0.19</td>
</tr>
<tr>
<td>Boileau et al. [30]</td>
<td>6.7</td>
<td>0.35</td>
<td>0.05</td>
</tr>
<tr>
<td>Shabani et al. [10]</td>
<td>6.9</td>
<td>0.33</td>
<td>0.4</td>
</tr>
</tbody>
</table>

† Specimens were naturally aged for 20 hours prior to artificial ageing.
‡ Castings were held at 495°C for an hour, then air quenched prior to solution treatment.

The data employed for this comparison precludes accurate comparisons between material with
Table 1.2: Coefficients for DAS-property relationships for T6 and AC conditions.

<table>
<thead>
<tr>
<th>Property</th>
<th>Condition</th>
<th>Coefficient (Eq. 1.1)</th>
<th>Fit characteristics</th>
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<tbody>
<tr>
<td></td>
<td></td>
<td>$a$</td>
<td>$b$</td>
</tr>
<tr>
<td>$\sigma_y$ (MPa)</td>
<td>AC</td>
<td>81.30</td>
<td>$-9.43 \times 10^{-5}$</td>
</tr>
<tr>
<td></td>
<td>T6</td>
<td>273.94</td>
<td>$-1.39 \times 10^{-2}$</td>
</tr>
<tr>
<td>$\sigma_u$ (MPa)</td>
<td>AC</td>
<td>177.55</td>
<td>$-2.58 \times 10^{-1}$</td>
</tr>
<tr>
<td></td>
<td>T6</td>
<td>341.90</td>
<td>$-1.76 \times 10^{-2}$</td>
</tr>
<tr>
<td>$\varepsilon_f$</td>
<td>AC</td>
<td>-0.0168</td>
<td>0.8360</td>
</tr>
<tr>
<td></td>
<td>T6</td>
<td>-0.9155</td>
<td>1.300</td>
</tr>
</tbody>
</table>

equivalent DAS, as it spans a large variety of compositions, microstructural scales and heat treatments. However, the following conclusions can be made. Regardless of specific heat treatments applied, there is little change to yield and ultimate strength for DAS less than 40 µm. Beyond 40 µm, the ultimate tensile strength is found to decrease more rapidly with increased DAS than seen with yield strength, i.e. larger $b$ and $n$ values are observed for $\sigma_u$ versus $\sigma_y$. AC data shows that the yield strength is increased by nearly three times for small DAS, and does not change significantly as DAS increases. Based on the fits to the elongation data, for values of DAS less than 40 µm, heat treatment does not affect the ductility. At higher levels of DAS, the AC material retained ductility, while elongation was exhausted for the heat treated material at DAS values of approximately 100 µm.

A further review of the impact on mechanical properties of other microstructural factors beyond DAS, such as compositional changes, porosity and eutectic particle morphology is presented in Table 1.3. This includes the mechanical properties discussed previously ($\sigma_y$, $\sigma_u$, $\varepsilon_f$) as well as the strain hardening rate, $\Theta$. This data shows that increased amounts of Mg and Si increase yield and ultimate tensile strength and decrease elongation. As demonstrated by Kashyap et al. [31], Fe diminishes all properties, particularly when the content is above 0.2%-wt. Lee [32] found that microporosity (0.2-1 volumetric %) left yield strength unaffected (< 1% change) for material with an estimated DAS of 25µm. Elongation and ultimate tensile strength, however, were reduced by greater than 50% and 17%, respectively. Boileau et al. [30] used Hot Isostatic Pressing (HIP)\(^1\) to reduce or eliminate porosity in specimens with a larger range of DAS than Lee. Boileau et al. found that that yield,\(^1\)520°C for 3 hours at 105 MPa
ultimate tensile strength and elongation are only slightly affected. This finding was echoed by Ran et al. [27].

Shabestari and Shahri [33] demonstrated that eutectic particle sizes and aspect ratios increased with DAS, as well as showing that the number of discrete particles per unit area (distribution) decreases with DAS. These findings are similar to those of Wang [11], who conducted targeted studies comparing specimens with the same DAS. Wang also considered the shape and size of the eutectic-Si particles in modified and unmodified A356 samples. With an aspect ratio of 1 being a perfect sphere, unmodified eutectic particles were found to have an aspect ratio of 2.5, compared to 1.6

\[ \sigma_y, \sigma_u \text{ and } \varepsilon_f \text{ versus DAS after (I) Ran et al. [27], (II) Wang [28], (III) Ceschini et al. [29], (IV) Boileau et al. [30] and (V) Shabani & Mazahery [10] with fitted expressions coinciding with T6 and AC condition.} \]
following modification. Particle size was twice as large in the former compared to the latter. Wang reported a significant increase in the yield strength for A356 with larger particles. Wang did not demonstrate a significant change in $\sigma_u$ in unmodified material. However, appreciable changes in $\sigma_y$ and $\epsilon_f$ were found. Cáceres et al. reported similar findings as Wang for $\epsilon_f$ [8].

Table 1.3: Relative microstructural effects on mechanical properties of Al-Si-Mg alloys in T6 condition. $\sigma_y$, $\sigma_u$ and $\epsilon_f$ are presented in addition to strain hardening rate $\Theta$, where available.

<table>
<thead>
<tr>
<th>Increase in:</th>
<th>$\sigma_y$</th>
<th>$\sigma_u$</th>
<th>$\epsilon_f$</th>
<th>$\Theta$</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>Silicon content</td>
<td>↑</td>
<td>↑</td>
<td>↓</td>
<td>NA</td>
<td>[31,34]</td>
</tr>
<tr>
<td>Magnesium content</td>
<td>↑</td>
<td>↑</td>
<td>↓</td>
<td>↑</td>
<td>[11,31]</td>
</tr>
<tr>
<td>Iron content</td>
<td>↓</td>
<td>↓</td>
<td>↓</td>
<td>NA</td>
<td>[31]</td>
</tr>
<tr>
<td>Porosity</td>
<td>↓</td>
<td>↓</td>
<td>↓</td>
<td>NA</td>
<td>[27,32]</td>
</tr>
<tr>
<td>Eutectic particle conditions (↑/↓ with DAS)</td>
<td>[28,33]</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Size (↑)</td>
<td>↑</td>
<td>~</td>
<td>↓</td>
<td>↑</td>
<td>[8,11,28]</td>
</tr>
<tr>
<td>Aspect ratio (↑)</td>
<td>↑</td>
<td>~</td>
<td>↓</td>
<td>↑</td>
<td>[8,11,28]</td>
</tr>
<tr>
<td>Distribution (↓)</td>
<td>↑</td>
<td>↑</td>
<td>↓</td>
<td>↓</td>
<td>[11,33]</td>
</tr>
</tbody>
</table>

### 1.3.2 Hardness

Macrohardness measurements are commonly used to empirically quantify the effects of heat treatment on strength ($\sigma_y$, $\sigma_u$) [35]. This is possible through correlation of the flow stress at some level of strain. The basic quality being measured during a hardness test is the ability of the material to resist local plastic deformation, which is imparted by an indenter with standardized geometry under quasi-static loading. Macrohardness is preferred over microhardness especially for cast aluminum alloys, as larger indents provide an improved sampling of the overall or average microstructural effect. Developed as an alternative to ball-type macrohardness measurements such as Brinell or Rockwell, the pyramidal diamond Vickers test allows for accurate measurements across a variety of loads [36,37]. The hardness from ball-type measurements is sensitive to both indenter geometry and load, while Vickers measurements are sensitive to indentation size, particularly for materials which exhibit coarse microstructures.

In tracking the effects of Fe and Mn on Al-Si alloys, Tash et al. [4] reported the results of Brinell hardness ($H_B$) measurements for both modified and unmodified A356 in the AC, solutionized and artificially aged conditions. In this study, artificial ageing was performed on several different tem-
temperatures for four hours (Fig. 1.5). These results demonstrate that there is an appreciable difference in hardness for the different conditions. For material with low Fe content, in all conditions, the modified material displayed elevated hardness values as compared to the unmodified. This finding, coupled with Table 1.3 indicates that simultaneous refinement of eutectic size and aspect ratio due to modification are reflected in increased hardness values. However, the unmodified material was found to be harder when Fe content was very high (>0.4 wt% Fe) in the overaged condition which was attributed to the contribution of significantly harder intermetallics to the overall measurement. Thus it is interesting to note that as Fe content negatively impacts yield strength (Table 1.3), a higher hardness measurement does not necessarily indicate a higher yield strength.

A recent study by Tiryakioğlu et al. [38] tracked both the yield strength and the Rockwell hardness ($H_{RF}$) in samples of Al-7wt%Si-Mg alloys containing varying amounts of Mg. The solution treatment in the study was identical for each sample, but the ageing time at 200°C was varied from 2 minutes to several days. It was found that there were discrete trends in hardness versus yield behaviour correlated to Mg content, which were valid up to the point of peak ageing. Peak hardness and the rate with which hardness increased with yield strength was found to increase with Mg content. In the overaged condition, there was no discernible difference in the $H_{RF}$-$\sigma_y$ trend with different Mg content. The data from Tiryakioğlu et al., presented in Fig. 1.6, exhibits a hysteresis associated with the irreversible thermodynamic change associated with over-aging. This presents
a caveat when comparing hardness to yield strength: hardness values can be correlated to yield strength, but the two discrete regimes behaviour for over or underaged must be considered. It appears that for a given composition, there is a distinct relationship for hardness to yield strength until peak age is attained.

![Figure 1.6: $\sigma_y$ versus $H_{RF}$ after Tiryakioğlu et al. for Al-7wt%Si-Mg alloys plotted according to Mg content in wt%. Dashed lines indicate power-law trend fitting.](image)

1.3.3 Behaviour at elevated temperatures

Estey et al. [39] measured the constitutive behaviour of A356 in the solutionized condition to provide input for a mathematical model aimed at predicting the distortion in an automotive wheel following heat treatment. Estey et al. performed uniaxial compression tests on AC then solutionized samples over a range of temperatures (200 to 500°C) and strain rates ($\dot{\varepsilon} = 0.001$ to 1 s$^{-1}$) to characterize the mechanical behaviour for conditions relevant to the quenching. Besides demonstrating the flow stress (i.e. $\sigma$ versus $\varepsilon$) behaviour of solutionized A356 at elevated temperatures, the results of Estey et al. showed that the material demonstrates decreasing work hardening rates as temperature is increased. Between 400-500°C, near-steady state flow stresses were observed.

McQueen et al. [40] compared the mechanical behaviour of both AC A356 and a Si-C reinforced A356 MMC. Torsion tests targeting temperatures of 300, 400, 500 and 540°C were performed at equivalent strain rates of 0.1, 1 and 5 s$^{-1}$. The results for A356 are presented in Fig. 1.7. McQueen et al. converted torque-twist measurements to equivalent stress versus strain according to the Fields and Backofen method [41]. This methodology requires fitting torque-strain rate and work hardening
coefficients in order to properly resolve equivalent stresses. The overall trend observed in this data is that ductility (Fig. 1.7b) increases with temperature and decreasing strain rate.

A peak flow stress accompanied by a gradual decline with increased strain as depicted in Fig. 1.7a would suggest ‘classic’ dynamic recrystallization (DRX) behaviour [42], as is the case for many fine-grained, precipitate-bearing aluminum alloys. This phenomena clearly manifests in the data supplied by McQueen et al., however, there is a gradual decline in flow stress as opposed to oscillating flow stress indicative of DRX [43]. McQueen et al. posit that dynamic recovery (DRV) is prominent at elevated temperatures as opposed to recrystallization, with DRV localized in the eutectic phase. The rationale provided is that there is little solute within the α-Al to provide grain nucleation sites. McQueen et al. also precluded eutectic-Si particles coalescing to further contribute to the strain softening observed, citing their stability. Observations made on the fracture surfaces by McQueen et al. suggested that failure at lower temperatures was due to cracks nucleating from coarse Si particles and this phenomena diminished at elevated temperatures. The main fracture planes coincided with the transverse plane of maximum shear stress. Cracks were found to predominantly pass through the eutectic-Si particle/matrix interface as opposed to directly through Si particles.

Kim et al. [44] studied the suitability of an Al-Si-Mg alloy for a casting-forging process. It was found that chemical refinement employing titanium boride and Zr was necessary to attain the desired
ductility for defect-free processing. The material was deformed at 450°C with tooling surfaces held at 250°C, presumably to stabilize the workpiece temperature during the forging operation. The authors reported that the aspect ratio of the microstructure was refined consistent with the forging path. After forging and applying a T6 temper, the hardness of the forging was found to have increased compared to a unforged component of the same composition and temper.

1.4 Fatigue behaviour of A356

The uniaxial fatigue behaviour of A356–T6 has been studied by a number of researchers [45–52]. These studies clearly established a direct link between microstructure, defects and fatigue resistance. In almost all related studies, casting defects such as intermetallic inclusions, porosity, and oxide films have been shown to be present at the origin of the failure. With few exceptions, these studies have only considered the fatigue behaviour under uniaxial loading. The multiaxial fatigue behaviour of A356–T6 was studied by De-Feng et al. [53] using thin-walled tubular specimens but under loading conditions leading to very low cycle fatigue; as such these results are not directly applicable to HCF conditions. McDowell et al. [54] performed torsional HCF testing, however these tests were conducted with deformation control, which requires an indirect translation to stress-based criterions.

1.4.1 The role of defects

Aluminum castings are negatively affected by casting defects such as macro and micro porosity, shrinkages, and oxide films. These defects in the microstructure are stress concentrators and provide sites for early crack initiation, thus shortening fatigue life. In some cases, the nucleation step in fatigue crack growth may be eliminated if there is a pre-existing flaw. This induces a localized stress concentration that initiates and speeds crack propagation. Pores act as such crack initiation sites near the surface of the material and the propagation of the crack is decided by both the applied stresses and the local strength of the material.

Boileau et al. reported significant increases in fatigue limit following HIP, as shown in Fig. 1.8. Processing in this manner provided an order of magnitude increase in $N_f$ for a constant $\sigma_a$ and DAS. This testing was conducted under fully reversed conditions, i.e. the maximum stress matched the minimum rendering a load ratio $R_L$ equal to -1. Gao et al. [55] also found similar increases in fatigue resistance with HIP-processed A356–T6 material with $R_L = -1$. While it is
clear that defects are observed at the initiation point on fracture surfaces, very few studies have identified the critical defect size that diminishes the fatigue limit for cast and heat treated aluminum alloys. While processing these alloys via a HIP step has been shown to increase fatigue resistance, the changes in behaviour cannot be limited solely to the elimination of pores due to microstructural changes caused by the thermomechanical processing. Furthermore, HIP is not economically viable for many commercial applications. It is therefore necessary to determine a critical defect size, or the permissible size of a defect before the fatigue endurance limit is affected.

Brochu et al. developed a Kitagawa (also referred to as Kitagawa-Takahasi [56]) relationship for rheocast A357. This analysis consisted of comparing fatigue limits to defect sizes that were measured directly from fracture surfaces. The authors employed the Murakami [57] approach which assumes pores act as cracks and are characterized by the parameter $\sqrt{\text{area}}$. This parameter is defined as the equivalent length of a defect projected in 2D onto the fracture surface. Brochu et al. determined that the critical defect size is 150 $\mu$m under fully reversed tensile loading. This experimental result was larger than the 100 $\mu$m critical defect size suggested by Fan et al. [48] for A356–T6, identified via modelling efforts.

**Figure 1.8:** DAS and porosity effect on AC fatigue properties demonstrated by Boileau et. al. [30] employing HIP. Fatigue testing was conducted with $\sigma_a = 158$ MPa and $R_L = -1$. 
1.4.2 Behaviour in defect-free material

In defect-free cast aluminum alloys, the first cracks are known to initiate either at Fe-rich inter-
metallic particles [55], or in the eutectic [54]. In cases where there are no defects or intermetallics,
the DAS can be correlated with the fatigue life [30, 54, 55]. Boileau et al. showed that \( N_f \) can be
increased by an order of magnitude in AC material when DAS is reduced by a third (Fig. 1.8). Gao et
al. demonstrated that halving the DAS increased the fatigue limit by 20 MPa in A356–T6 according
to a Basquin relationship fitted to their data (Fig. 1.9). While the effects of porosity and DAS are
not mutually exclusive, the role DAS plays is less important when the material contains defects.

\[
\begin{align*}
\lambda &= 23 \pm 4 \mu m \\
\lambda &= 47 \pm 5 \mu m \\
1471(2N_f)^{-0.19} \\
1374(2N_f)^{-0.21}
\end{align*}
\]

**Figure 1.9:** \( S – N \) data demonstrating DAS effect on A356–T6 fatigue properties after Gao et
al. [55]. Fatigue testing was fully reversed (\( R_L = -1 \)).

The main reason that the DAS correlates well with fatigue resistance is the linkage to eutectic
particle condition. Fatigue crack initiation has been found to occur at these particles [54, 58] in
samples that are free of defects. Coarse microstructure, as characterized by a large DAS, contains
large, irregularly shaped and spaced eutectic particles which act as stress concentration points within
the surrounding matrix. In the case of friction-stir processed material, the eutectic particle sizes
are significantly refined and fatigue properties were significantly enhanced, resulting in a >80%
increase in endurance limit (\( R_L = 0.1 \)) as compared to the AC condition [59].
1.5 Rotary forming and related processes

Rotary forming or spinning is a near net-shape, hot or cold metal-working process for manufacturing seamless, dimensionally precise, rotationally symmetric products. In this process, the workpiece is impinged between a tailstock and a rotating mandrel. It is then incrementally deformed by contact with a tool. Usually the tool employed consists of a roller to minimize friction. The mandrel may support the workpiece throughout the length of the intended final profile, however, some spinning operations can be successfully conducted without the benefit of support on the internal diameter.

Industrial classification of rotary forming operations spans three discrete categories (Fig. 1.10) based on the change in wall thickness of the component and the internal stresses developed. In terms of wall thickness, spinning induces little to no change while shear forming or cone spinning induces a uniform change according the sine rule [60, 61], as detailed in Fig. 1.10b. If a flat blank is used, then the final thickness of the product is the starting thickness times the sine of the imposed forming angle. The sine rule can still be applied if the operation is to be conducted on a previously deformed component already containing an included angle. In flow forming or tube spinning, the final wall thickness may be estimated by conservation of volume if the workpiece wall thickness is reduced uniformly.

Depending on the profile, the principal stresses induced in the workpiece by spinning may be either compressive or tensile. A review of the mechanics of spinning by Music et al. [62] states the most desirable stress state under the roller is pure shear, which does not induce any workpiece thickness reduction. This is achieved by balancing the circumferential stresses imposed by the roller with the radial stresses developed through thickness of the workpiece. Moving to larger unbalanced stress states has been shown to cause defects, as shown in Fig. 1.11. Music et al. observed that high circumferential compressive stresses result in buckling of the flange, leading to wrinkling in the workpiece flange ahead of the roller. As shown in Fig. 1.11a these wrinkles develop progressively throughout the spinning operation, eventually manifesting as pronounced lobes at the end of the workpiece. Music et al. state that high radial tensile stresses may induce the formation of circumferential cracks, or cracks which propagate in the $R - \theta$ plane. In a review of rotary forming practices, Wong et al. [63] report that radial cracks, or cracks that propagate in the $z - R$ plane, occur when wrinkles are further worked. These cracks predominantly appear between
Figure 1.10: Spinning process classifications and final workpiece thickness estimates after Runge [61]. Spinning induces little change to the workpiece thickness, while shear spinning proceeds according to the sine rule. Final wall thickness in flow forming may be estimated by conservation of volume.

between lobes.

Shear forming and flow forming principally cause compressive stresses, predominantly through-thickness. The initial workpiece shape distinguishes the latter and the former processes: nominally a flat blank is employed in shear forming whereas flow forming employs a bushing or tubular blank. The classifications and depictions of each of these processes occasionally overlap in the literature, as shear forming process parameters can be changed to approach those of flow forming. A practice called ‘overspinning’ where the final target thickness is less than that dictated by the sine rule produces forming conditions approaching forward flow forming, rendering shear spinning a nominal process description.

In flow forming, a rotating workpiece mounted on a mandrel is induced to incrementally deform between a mandrel and a roller traversing the axis of the workpiece (Fig. 1.12). Depending on how the workpiece is supported, the material can be induced to move in either direction along the mandrel as compared to the motion of the roller(s). Flow forming allows for significant and con-
CHAPTER 1. INTRODUCTION

Figure 1.11: Typical spinning defects reviewed by Music et al. [62] and Wong et al. [63]. Originally described by Romanowski [64].

trollable thickness reductions axially through the part length by employing combined rolling and extrusion/drawing deformation mechanisms. The coupling of these mechanisms generates plastic strains that are much larger than would be realized by either mechanism on its own at the same forming loads [65–67]. In order to improve throughput, most commercial operations employ multiple rollers which operate at the same time in a coordinated fashion. Often these rollers are offset axially and radially along the length of the workpiece, having different geometric profiles. Since conventional spinning, shear spinning and flow forming process descriptions often overlap [68], the collective label of rotary forming processes may be used as a less complicated description; this label will be employed throughout this thesis.

The application of rotary forming results in complicated, highly non-linear tooling and workpiece interactions that are dependent on a myriad of process variables. These process variables include the rates of tooling travel, rotation rate of the workpiece, forming zone geometry, lubrication condition and workpiece material properties. These process variables echo those characterizing standard metal turning, leading to some researchers attempting to employ rotary forming techniques on solid cylindrical components [69,70] with some success.

Due to the complexity of the interactions between the part, tool, and mandrel during rotary form-
CHAPTER 1. INTRODUCTION

Figure 1.12: Flow forming configurations [66].

ing techniques, standard approaches to calculate forming loads and material responses using classical analytical techniques, such as plane strain approximations and slab analysis, are not directly applicable. Numerous attempts to develop analytical descriptions of flow forming have resulted in different formulations for contact area and therefore forming loads [71]. For example, the analytical tooling contact area proposed by Jahazi et al. [72] versus that of Gur and Tirosh [73] are vastly different. Furthermore, they describe forming conditions that do not persist in the process. These assumptions are then compounded by the use of simplified representations for flow stresses and friction. Employing the results obtained by the aforementioned formulations has therefore lead to potentially erroneous conclusions [74]. Developing another analytical description of rotary forming processes from a shear spinning perspective, such as that by Chen et al. [75], requires the assumption that deformation obeys the sine rule throughout the process and may not be extended to flow forming.

Finite Element Analysis (FEA) techniques have been successfully employed to provide better approximations of both forming loads and the material response during rotary forming, where elements of spinning and flow forming may manifest. As there is a great deal of literature regarding
CHAPTER 1. INTRODUCTION

this topic, the focus of the following subsection has been limited to the experimental studies of rotary forming on aluminum alloys. For a further review of rotary forming processes, the reader is directed to literature reviews by Music et al. [62] and Wong et al. [63].

1.5.1 Experimental studies of rotary forming

Experimental studies on this process conducted at ambient temperatures on wrought aluminum alloys have demonstrated that large amounts of plastic deformation may be imparted. Haghshenas et al. [76] reported that equivalent strains of up to 17 may be imparted to an aluminum workpiece. Applications of this forming technique specifically to cast aluminum alloys have shown the potential to reduce or eliminate porosity entirely with heavy plastic deformation, significantly improving fatigue performance. However, due to the lack of ambient ductility, spinning of cast aluminum alloys requires deformation at elevated temperatures in order to achieve a sound product.

Mori et al. [77] conducted spinning experiments at temperatures between 350 and 400°C on cast A357 alloy blanks machined from larger castings. A schematic of the experimental apparatus is shown in Fig. 1.13a. It consisted of a stepped mandrel directly driven by a motor with a 2 mm thick blank bolted to the face. A steel enclosure surrounded the immediate space around the workpiece into which air at 700°C was introduced. When the blank was at the intended forming temperature as identified with an infrared thermocouple (IR-TC), the blank was spun into contact with a numerically controlled roller actuated in both the $z$ and $R$ directions. Cracking was observed at 350°C in the deformed blank corresponding with the step in the mandrel (Fig. 1.13a). When the temperature was increased to 400°C, cracking was not observed and post-deformation analysis found that porosity had been eliminated at wall thickness reductions of 25% and greater.

While not quantified, it was reported that the DAS was reduced in-line with the wall reduction level. As compared to undeformed material, it was reported than an increase in yield strength of the deformed material was found after T6 heat treatment (Fig. 1.14). Ductility, as characterized by elongation, was increased by approximately twofold over all deformation levels. However, the specific T6 schedule followed by Mori et al. was not detailed, and therefore it is difficult to differentiate between the effects on strength due to deformation and heat treatment.

Zhao et al. [79] conducted elevated temperature spinning experiments on strontium modified LPDC A356 tubes with a starting wall thickness of 23 mm. At severe wall thickness reductions,
the dendritic structure was no longer recognizable in some locations. Average dendrite arm spacing was modified from 37.2 to 23 \( \mu \text{m} \) at wall thickness reductions of 80\%. Mechanical testing of this material also showed improvements in the mechanical properties following heat treatment (Fig. 1.14). The undeformed material had a hardness of \( H_B = 69 \), which increased to 80 in material with a 70\% wall thickness reduction. The hardness reported in the undeformed condition matched that of Tash et al. (Fig. 1.5) for material in the solutionized condition suggesting that a non-standard heat treatment was employed. Zhao et al. did not disclose the forming temperature used.

Cheng et al. [78] employed a numerically controlled industrial forming apparatus (Fig. 1.13b) to reduce wall thicknesses of blanks made of A356 with a diameter of \( \sim 400 \text{ mm} \) and a starting wall thickness of \( \sim 8 \text{ mm} \). Maximum thickness reduction was reported as 60\%. The apparatus consisted

Figure 1.13: Experimental apparatus employed by Mori et al. [77] and industrial forming apparatus employed by Cheng et al. [78].
of a mandrel with an actuated tailstock to hold the workpiece in place while being deformed simultaneously by three rollers. These rollers were offset axially and circumferentially about the forming axis, $z$. Cheng et al. found the same effects on microstructure as the two previous studies with a processing temperature of 350°C, however, they also observed a small decrease in Rockwell hardness of the material in the spun condition ($H_{RF} = 90.5 \pm 1.5$ versus $89.3 \pm 0.7$) post solutionizing for 6 hours at 540°C and ageing for 3 hours at 155°C. While tensile properties were not reported by Cheng et al., according to the data presented by Tiryakioğlu et al. (Fig. 1.6), this would represent a slight decrease in yield strength. This is incongruent with measurements reported by Mori et al. and Zhao et al., which indicate that deformation improves properties with increased deformation.

1.5.2 Finite element analysis of rotary forming

The development of rotary forming process models based on material characteristics, process geometry and other operating parameters are highly desirable. This section summarizes the efforts of researchers to address the previously discussed shortcomings of analytical models by employing FEA.
Isothermal finite element analysis of rotary forming

The earliest 3D FEA approaches used to simulate material response during rotary forming were performed by Xue et al. [80] and Xu et al. [1]. These models consisted of a 120° section of the workpiece, with the contact region of the roller pre-defined according to an analytical and experimental characterization of a flow forming process provided by Hayama [81]. The commercial FEA software package ADINA was employed by Xue et al., while Xu et al. employed an implicit method. Meshed tooling was employed in the former, while nodal velocities were directly imposed in the latter. Material response in both studies was implemented on a elastic-perfectly plastic basis, and boundary conditions were identical in both cases. Symmetry boundary conditions were imposed on the edges of the 120° section, which implies three rollers simultaneously in contact on the same axial plane, and all nodes on the inner diameter of the workpiece were radially constrained. This configuration does not accurately reflect the forming conditions in a discrete component as it lacks any description of strain rate effects, nor is the evolution of the elastic loading through the annulus of the workpiece described. Furthermore, all diametral growth of material entering the roller impingement zone was suppressed due to the radial constraint, which prevents the simulation of a common phenomena in the flow forming process [82].

More recently, Mohebbi et al. [83] developed a geometrically identical model to Xu et al. with the exception of using a full 360° workpiece, and a single roller. Both the roller and mandrel were modeled as rigid analytical surfaces, however the same radial boundary conditions imposed by Xu et al. and Xue et al. were retained. The commercial software package ABAQUS Explicit was employed in this work. A Hollomon hardening description for the wrought aluminum alloy was employed. None of the aforementioned studies present quantitative comparisons between experimental and their FEA predictions. However, Mohebbi et al. found that their simulated forming loads agreed with the predictions of an analytical model developed by Hayama [81]. Unfortunately, they did not report the mean flow stress employed nor the friction factor employed in the analytical formulation. Together, the efforts of Xue et al., Xu et al., Mohebbi et al. and Hayama provide a description of the principal deformation modes along principal axes of the workpiece, as shown in Fig. 1.15.

Wong et al. [69] used both the implicit and explicit versions of ABAQUS to investigate rotary forming of solid lead cylinders using flow forming techniques. Wong et al. found that implicit
techniques provided more accurate predictions of forming loads and final deformation based on comparisons with experimental data. However, it was found that the computational penalty for using implicit formulations was excessive, suggesting that explicit formulations are the best approach in terms of computational resources used. Using the results of the implicit FEA, it was shown that an explicit formulation with a large mass scaling factor agreed quite well with the implicit results while providing a near ten-fold reduction in computational time. Wong et al. were the first to summarize the main challenges faced when using FEA to analyze flow forming and related processes [67, 69].

They are as follows:

- Flow forming is an incremental forming operation, as such only a small portion of the surface of the workpiece comes into contact with the roller at any given time. The contact area is continually changing as the roller moves leading to difficult contact formulations.
- A rotating workpiece induces a computational penalty as all nodal positions are affected during each time step.
- The deformation process has no plane of symmetry, requiring a 3D domain with very high element counts.

All of the aforementioned studies have made comparisons between experiments and simulations on a qualitative basis. Only recently have quantifiable comparisons of experimental measurements
to those from FEA been attempted for rotary forming. The most recent study containing this comparison was conducted by Wang and Long [84–86] on a conventional spinning process of steel workpieces. The sole comparison provided was the final thickness of the part versus normalized part length, which exhibited 15% error. Owing to the lack of through-thickness reduction inherent in spinning, this error may not reflect the overall accuracy of the model. A model of slit spinning of aluminum by Zhan et al. [87] found an overall maximum error of 22% in all dimensions, with the maximum discrepancy coinciding with regions of the workpiece having the largest departure from the sine rule. The dimensions involved in this comparison were not provided. Huang et al. [88] found that a model of slit spinning under-predicted forming forces by 16%. However, the model was inconsistent with the experimental process on which it was based [89] in terms of geometry and material description.

All of these preceding studies did not consider the evolution of temperature within the workpiece, employing material descriptions that were based on a single forming temperature. This is particularly tenuous when considering that a number of these modelling efforts attempt to describe the process for strain rate and temperature dependent materials.

**Coupled thermomechanical finite element analysis of rotary forming**

Michel et al. [90] developed a fully coupled explicit model for spin extrusion using MARC. The spin extrusion process consisted of a combination of flow forming and backwards extrusion, or a hybrid between the Mannesmann[2] and flow forming processes. This is the only model reported in the literature on rotary forming that couples the thermal and mechanical aspects of the process to account for the heat generated due to inelastic deformation, or the conversion of strain to thermal energy. This is characterized as the inelastic heat fraction or Taylor-Quinney coefficient [91], which is a ratio of strain energy converted to heat during deformation. Heat generation within the workpiece and subsequent transfer to the surrounding environment and to the tooling was considered in the model. However, frictional heating at the tooling interface was ignored. Michel et al. did not compare their model against any experimental data, nor document their material description.

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2Cross-roll piercing process for fabricating seamless pipe
1.6 Scope and objectives

While the characteristics of A356–T6 have been well documented in the literature in terms of the net effect of heat treatment on mechanical properties, there is a paucity of data regarding the material in the AC form. The effect of holding aluminum alloys in the AC condition at the elevated temperatures necessary for forming purposes, followed by deformation, has unknown implications on the final microstructure and mechanical properties of the heat treated material. The properties of heat treatable aluminum casting alloys are dependant on microstructural features spanning several length scales, which are all affected by processing in this manner. Understanding how the microstructure changes during the different stages of this processing route may help explain the mechanical properties of rotary formed cast aluminum reported in the literature.

The complexity of rotary forming has precluded the development of effective analytical techniques to estimate important process parameters such as forming loads. Rotary forming process models employing FEA to accurately determine process parameters and final product dimensions would increase adoption of the process and reduce the need for costly experimentation. However, to date, studies in this area have used simplistic material models, or have made other incongruous assumptions. Furthermore, the surveyed literature indicates that there is a need to improve the material behaviour description in rotary forming models and to assess their predictions through comparison to experimental data.

Additionally, there is a scarcity of multiaxial HCF data for A356–T6, including the associated fracture mechanisms and an experimental basis for a multiaxial fatigue criteria. There is also a lack of data regarding the critical defect size required to diminish the endurance limit of A356–T6. This characterization would set a foundation for identifying the specific methods by which rotary forming improves fatigue resilience.

1.6.1 Objectives

The objectives of the present work are to investigate the effects of rotary forming on cast aluminum alloys, specifically, A356 and to develop a model to investigate the evolution of stress state, workpiece deformation and deformation rate. Beyond investigation of the general effects of rotary forming, the objectives also include augmenting fatigue characterization of A356–T6, to distinguish the effects of rotary forming on fatigue resilience.
To achieve these objectives, the following research tasks were identified:

- Characterize the flow stress behaviour of AC A356 at various elevated temperatures and strain rates in the AC condition, and ascertain an appropriate constitutive behaviour suitable for FEA;

- Develop a simplified Experimental Forming Apparatus (EFA) to produce rotary formed material with known processing histories, capable of imparting deformation at temperatures indicative of a commercial rotary forming process;

- Experimentally ascertain the effects of the rotary forming processing route (deformation followed by heat treatment) on a workpiece;

- Develop a process model of rotary forming capable of capturing thermal and deformation history of the material, based on the inputs dictated by the EFA;

- Study crack growth and the impact of porosity of A356–T6 under multiaxial fatigue conditions; and

- Characterize the improvement in fatigue resilience of A356–T6 imparted by rotary forming

The research tasks represent substantive and novel contributions based on the literature reviewed previously in this chapter. While a significant number of studies have been undertaken to study the deformation behaviour of A356, all have focused on the material following various thermal treatments. Deformation behaviour has been characterized in the AC condition at different strain rates and temperatures under torsion, however, this was for a select number of conditions. While there are several published examples of isothermal FEA models of rotary forming, there have been no fully coupled thermomechanical models reported and validation has been mostly limited to qualitative observations. There has been limited work conducted on tracking the effects of thermomechanical processing on A356 in the AC state. Although many uniaxial studies on the fatigue behaviour of A356–T6 have been published, an attempt at identifying the critical defect size from a multiaxial fatigue standpoint has not been previously attempted. While an improvement in specific mechanical quantities for rotary formed cast aluminum has been reported, there is a paucity of published quantitative data on the effects of rotary forming on fatigue properties of any alloy.
1.6.2 Scope

The degree to which the aforementioned research tasks were pursued has been grouped according to five main analysis types, as described in Fig. 1.16. This figure also describes the source of each material type and the processing applied to each material prior to each analysis. These analyses are described in detail below.

Figure 1.16: Material and data flow outlining the present work.

In order to address the lack of deformation data for A356 at elevated temperatures and strain rates in the AC condition, a large number of isothermal compression tests were conducted in an effort to develop a comprehensive constitutive equation. The material for this study was sourced from a wedge cast in an apparatus configured to generate specimens with similar microstructure to material cast in a commercial LPDC operation. Analysis of the experimental data was analyzed within a constitutive framework, with the end result being an expression capable of describing the flow stress as a function of temperature, strain and strain rate over a wide range of conditions. This expression is necessary for the coupled thermomechanical model.

A fully coupled thermomechanical model of a rotary forming operation was developed. The scope of the model was limited to predicting final part geometries compared to several different forming experiments. These forming experiments, conducted with the EFA, encompass workpiece
wall thickness reductions ranging from conventional spinning to flow forming. The forming conditions imposed by the EFA, combined with the constitutive behaviour description provided the main model inputs.

The effect of holding A356 in the AC state at elevated temperatures for forming purposes, followed by deformation has unknown implications on the final mechanical properties following heat treatment. The properties of heat treatable aluminum casting alloys are dependant on microstructural features spanning several length scales, all of which are affected by processing in this manner. Therefore, the effects of forming operations mirroring those employed with the EFA on the microstructure was characterized, both in the as-formed and T6 condition. This was accomplished through microstructural observations on specimens with various thermomechanical histories, coinciding with extensive hardness measurements, Scanning Electron Microscopy (SEM) and Energy-Dispersive X-ray spectroscopy (EDX) analysis.

Pursuant to the thermomechanical characterization, the fatigue properties of A356–T6 and material with varying degrees of deformation in the same heat treated condition were determined. Commercially formed material, in addition to material processed by the EFA, was included in this characterization. The commercially processed material embodied significantly more deformation than that delivered by the EFA. Performance was gauged via uniaxial HCF testing.

The current work contributes to understanding the multiaxial HCF behaviour of A356–T6, which has seen little attention in the literature. Generating a wide variety of microstructure seen commercially, multiaxial fatigue mechanisms were analyzed through fracture surface observations to compare against existing uniaxial understanding. Employing a Kitigawa-type analysis, the influence of casting defects on the HCF behaviour was also examined, permitting the evaluation of a critical defect size.
This chapter details the sources of the materials employed in the present work, along with descriptions of the apparatus and the experimental methods used.

2.1 Material

All of the A356 materials used in this study were sourced from a North American wheel manufacturer in various states, ranging from melt to LPDC wheels in the AC, T6 and rotary formed-T6 conditions. The characteristic composition of all material is given in Table 2.1:

Table 2.1: Composition (%-wt) of modified A356, balance Al.

<table>
<thead>
<tr>
<th>Si</th>
<th>Mg</th>
<th>Fe</th>
<th>Ti</th>
<th>Na</th>
<th>Sr</th>
<th>Ni</th>
<th>Cu</th>
<th>Zn</th>
<th>Ca</th>
<th>Zr</th>
</tr>
</thead>
<tbody>
<tr>
<td>6.714</td>
<td>0.334</td>
<td>0.130</td>
<td>0.124</td>
<td>&lt;0.001</td>
<td>0.014</td>
<td>0.005</td>
<td>0.005</td>
<td>0.002</td>
<td>0.001</td>
<td>0.004</td>
</tr>
</tbody>
</table>

A356 in melt form was used to cast wedges using a purpose built apparatus (Fig. 2.1), at the site of the aforementioned wheel manufacturer. Samples were cut from the wedges for constitutive behaviour analysis (Type 1) and multiaxial fatigue characterization (Type 2). The wedge casting apparatus consisted of a three part AISI–4320 steel mould (two halves containing a runner and gating system as well as the sides of the wedge cavity) and a water-cooled chill plate which formed the bottom of the cavity. The mould halves were held together by a combination of clamps and

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1Portions of this chapter have been published in:


33
bolts. The casting methodology was altered slightly for each wedge type. For the Type 1 wedge, the runner and gating system was employed and the chill was cooled with 10°C water supplied through two 6.4 mm diameter channels at 3.8 bar. For the Type 2 wedge, the gate system was blocked and the mould cavity was filled directly with melt, and no cooling water was employed. The motivation for this wedge geometry was to create a gradient in cooling rate varying with height in the wedge. This methodology was employed to generate material with a range of DAS sizes and defects. In both cases, a more refined microstructure was found at the base of the wedge where the cooling rate was the highest as compared to a coarser microstructure at the top where the cooling rate was the lowest. The Type 1 wedge exhibited a finer microstructure overall, consistent with commercial LPDC processes, while the Type 2 wedge provided material consistent with LPDC at the base and becoming coarser with height.

LPDC wheels with the same composition as given in Table 2.1 were also provided in several
2.1.1 Material employed in constitutive behaviour analysis

Cylindrical compression test specimens, nominally measuring 10 mm in diameter by 15 mm in length, were extracted from an AC Type 1 wedge. The mean DAS of this material was found to be 39.9 $\mu$m with a standard deviation of 7.0 $\mu$m. Gas-based porosity was widely dispersed with a mean percent area of 0.154% based on micrographic observations. These measurements are characteristic of the cross-section of the wedge where specimens were extracted and are similar to those observed in the wheel castings employed in parallel studies. Figure 2.2 show micrographs of the typical microstructure observed.

2.1.2 Castings used for experimental forming

AC wheels, produced via a standard LPDC production process, were obtained from the aforementioned North American wheel manufacturer. Blanks suitable for spinning on the EFA were machined from these castings by removing the spokes and hub of the wheel through machining. The blanks

**Figure 2.2:** Typical microstructure of compression test specimens extracted from a Type 1 wedge. Low magnification in (a) and higher magnification in (b).
did, however, retain the original cast surface in the region that was to be deformed. Employing a blank that could be machined from a wheel casting both eliminated the need for a purpose built mould to cast blanks and made use of material generated using the industrial process. The resulting blank measured approximately 140 mm axially, had a minimum internal diameter of 330 mm, and an approximate 10 mm wall thickness. Fig. 2.3 shows an axial cross-section of the blank.

Figure 2.3: Cross-section of AC blank used with EFA. The blank datum employed is indicated at $z = 0$. Fatigue sample extraction region is indicated as well as coupon and circumferential section locations.

As discussed in later chapters and sections, these blanks were formed by varying amounts with the EFA, and characterization was undertaken both prior to heat treatment and afterwards. Hardness profilometry was carried out across the entire axial cross-section, and on a 72° circumferential section. Detailed microstructural analysis and targeted thermal treatments were completed on coupons cut sequentially in the circumferential direction from the AC blank. A single undeformed blank and the most heavily deformed workpiece resulting from the forming experiments had uniaxial fatigue specimens extracted from axial sections. The fatigue specimens were removed following heat treatment of both components.

2.1.3 Material employed in multiaxial fatigue behaviour

The material employed for the multiaxial fatigue study was sourced from both a Type 2 wedge and a LPDC wheel, as shown in Fig. 2.4. The wedge was divided into regions where fatigue specimens were drawn from based on changes observed in the microstructure. One half of the wedge was devoted to multiaxial fatigue specimens (Fig. 2.22), and the other half had a mixture of multiaxial, uniaxial and torsion specimens drawn from it. All specimens were then heat treated to a T6 condition after being removed from the wedge block as described in Section 2.3.1. Multiaxial specimens were
Figure 2.4: Multiaxial fatigue specimen locations in Type 2 wedge (a), extraction points of material from spokes (c) of an LPDC wheel (b) in the T6 condition.

also extracted from the spokes of the LPDC wheel which had the same material composition as the wedge casting. This wheel was received in the T6 condition.

The scale of the microstructure showed a marked increase from the bottom of the wedge to the top (Fig. 2.5) coinciding with the cooling rate differential imposed by the casting practice. This is characterized by the measured DAS which increased from the bottom (39.5 µm) to the top (72.2 µm). Average porosity remained relatively uniform throughout the wedge based on the percent area. The peak pore size, as characterized by maximum \(\sqrt{\text{area found through metallographic analysis}}\) was 126 µm. The porosity measurements also show that the middle of the wedge, at a height of 89-120 mm, contained a region of elevated porosity as characterized by the maximum pore size. The DAS measured in the wheel specimen (36.7 µm) was similar to that measured at the bottom location in the wedge. The area percent porosity throughout the wedge was uniform (≈ 0.12%) and approximate double that measured in the wheel (0.06%). The reduced porosity content is consistent
Figure 2.5: Typical T6 microstructure and porosity in Type 2 wedge casting. Microstructure at the top of the wedge (a) versus (b) bottom with porosity highlighted, detailed in (c).

Table 2.2: Microstructure and porosity summary of material employed in the fatigue study. Standard deviation is provided where applicable.

<table>
<thead>
<tr>
<th>Family</th>
<th>Height (mm)</th>
<th>DAS (µm)</th>
<th>Mean Porosity</th>
<th>$H_{V5}$ (kg/mm$^2$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Wheel (W)</td>
<td>N/A</td>
<td>36.7 ±8.0</td>
<td>0.0603</td>
<td>86.2 ±2</td>
</tr>
<tr>
<td>Wedge Bottom (A*/B*)</td>
<td>28.75</td>
<td>39.5 ±7.6</td>
<td>0.1237</td>
<td>78.9 ±5</td>
</tr>
<tr>
<td></td>
<td>58.75</td>
<td>39.7 ±9.1</td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>88.75</td>
<td>47.6 ±14.0</td>
<td>124</td>
<td></td>
</tr>
<tr>
<td>Wedge Middle (M)</td>
<td>118.75</td>
<td>57.2 ±17.9</td>
<td>105</td>
<td>83.1 ±3</td>
</tr>
<tr>
<td></td>
<td>144.73</td>
<td>58.5 ±21.0</td>
<td>45</td>
<td></td>
</tr>
<tr>
<td></td>
<td>174.73</td>
<td>59.7 ±21.2</td>
<td>93</td>
<td></td>
</tr>
<tr>
<td>Wedge Top (T)</td>
<td>204.73</td>
<td>62.6 ±21.6</td>
<td>48</td>
<td>89.0 ±7</td>
</tr>
<tr>
<td></td>
<td>234.73</td>
<td>72.2 ±28.2</td>
<td>126</td>
<td></td>
</tr>
</tbody>
</table>

$^*$Artificial defects

with the degassing practices employed by the wheel manufacturer which substantially reduce the hydrogen content. A range of mean maximum pore diameters was measured in the wedge with the largest diameters observed in the upper portion of the bottom region (Bottom family, Table 2.2). The increase in pore diameter at this location was a result of the double ladle pouring procedure used to fill the mould cavity where the casting began to solidify between pouring the first ladle and the commencing to pour the second.
Hemispherical artificial defects were introduced to the middle of the gage sections of six fatigue specimens via Electro-Discharge Machining (EDM). In total, two torsion and four uniaxial specimens were drawn from the bottom of the wedge (family A, Table 2.2) and had artificial defects applied post heat treatment with a sink-type machine\(^2\). This technique of generating artificial defects has been qualified in other crack propagation investigations [92, 93]. An example of an artificial defect as it appearing on a fracture surface is shown in Fig. 2.18.

2.1.4 Commercially formed material

Commercial wheel castings intended for rotary forming were also produced by the same North American wheel manufacturer mentioned previously with the outboard side of the wheel (spoke, hub, etc.) near to net shape, and the inboard side cast with a short, thick wall. This inboard portion was then formed at elevated temperatures prior to heat treatment. Fig. 2.6 shows the cross-sectional geometry of the AC blank and the formed result. After casting and before forming, the blank was placed in a holding furnace at 350\(^\circ\)C for up to 6 hours. The blank was then placed on a mandrel and rotary formed by two diametrically opposed rollers. The path taken by the rollers moves material down in the \(z\) direction and out from the mandrel to form the inner tire bead of the wheel. After forming, the wheel was quenched and then passed through a T6 treatment. All wheels received were in the formed and heat treated condition as well as having undergone some machining. A single early prototype AC blank was also obtained.

Radius-gauge tensile fatigue specimens were extracted from 4 different locations in the received wheels, corresponding to different levels of deformation and orientation. As indicated in Fig. 2.6, specimens were drawn from the hoop or circumferential direction (\(\theta\)) in locations that encountered no direct deformation (‘H’) and significant deformation (‘HD’). A similar approach was taken to remove specimens oriented in the axial direction (‘A’ versus ‘AD’).

Preliminary analysis of the microstructure shows that the effects of forming are highly localized. This manifests primarily in changes to the DAS, ranging from lightly compacting the structure or eliminating it entirely, as shown in Fig. 2.7 The microstructure for both the formed-T6 and AC material was characterized and is discussed in greater detail in the following chapters.

\(^2\)Charmilles IsoPulse P25
Figure 2.6: Blank versus formed cross-section geometry showing principal regions affected by forming as well as location of fatigue samples oriented in the axial (A) and hoop (H) directions.

Figure 2.7: Commercially formed T6 microstructure at locations indicated in Fig. 2.6.
2.2 Experimental forming apparatus and methodology

A scaled rotary forming apparatus was developed to conduct instrumented rotary forming experiments. This EFA was constructed based on the geometry of blanks extracted from commercial wheel castings described in Section 2.1.2 and was designed to be flexible enough that different blanks could be employed through the design of modular tooling in follow-on experiments.

The core of the EFA is a Herbert No. 9 capstan lathe (Fig. 2.8a). This lathe was retrofitted with the necessary systems to be capable of performing repeatable forming experiments at elevated temperatures (Fig. 2.8b). All modifications were designed to incorporate commercially available off-the-shelf components where possible and custom fabrication elsewhere. The following sections describe the main components assembled and integrated into the lathe to make this possible.

2.2.1 As-received lathe and modification overview

The received lathe had the following important standard features: a spindle to which a workpiece may be affixed, a tool stand to hold a turning tool which may be moved along bed rails to encounter the workpiece, and a tailstock on the same bed rails placed after the tool mount to support long and heavy workpieces. This lathe was also equipped with a mast which engaged a capstan to control tooling and tailstock flexure during heavy turning operations. This feature, plus the overall mass of the machine made it a good candidate for conversion to rotary forming operations. Additional features and components appearing on the lathe are shown in Fig. 2.8a.

The lathe is belt-driven by a continuously run induction motor rated at 22.4 kW (30 HP). The output is split with a transmission to drive both the spindle and saddle, giving a range of discrete spindle speeds from 19 to 560 RPM. The longitudinal feed, which runs the saddle between the spindle and turret along the z-direction, is proportional to these spindle speeds. As received, the latitudinal feed which runs the tool in the R-direction was damaged, however manual movement of all tooling was retained.

2.2.2 Rotary tooling

The design of the rotary tooling was conducted in concert with the blank (Fig. 2.3) such that the blank could be extracted from a commercial wheel casting and fit the work envelope of the lathe. The geometry (Fig. 2.3) of the blank represents a compromise between removing segments of the
Figure 2.8: Main lathe components: (1) frame, (2) mast, (3) steady rest, (4) tailstock, (5) tool stand location, (6) spindle, (7) tool slide, (8) saddle, and (9) capstan on the as-received lathe in (a) and a CAD depiction in (b) of the same components and additions made for forming purposes.
CHAPTER 2. EXPERIMENTAL METHODS AND APPARATUS

Figure 2.9: Depiction of the rotary tooling assembly. The driveshaft connects the main mandrel weldment to the spindle, and is in turn supported by a center engaging the tailstock plate on the front face. The details of the clamp operation are also shown.

wheel perceived as barriers to forming while at the same time providing a clamping feature to hold a portion of the blank to the mandrel and fitting within the work envelope. The geometry of the blank provided the inputs necessary to further design the tooling to support it through forming, specifically, the clamping system and mandrel. The complete mandrel and clamping system is depicted in Fig. 2.9.

The main components of the mandrel consist of a slightly tapered, hollow weldment that is attached to a spindle with a hollow driveshaft. The mandrel’s taper was specified to match the taper on the blanks in the clamp region, which was then blended to a smaller taper intended to ease the release of the blank post-forming. The mandrel weldment is in turn capped by a centering plate which was brought into contact with a spring loaded ‘live’ centre\(^3\), installed on the tailstock (item 4 in Fig. 2.8a). This centering plate, held in place by a series of bolts, also serves to enclose the mandrel cavity, which contains a wiring harness for 10 type-K TCs that are welded into the surface of the mandrel. Extensions for the TCs are run through the driveshaft and spindle and connected to the wireless DAQ system described in Section 2.2.5. All components were constructed of AISI–

\(^{3}\)Ritens number 431-17124; #4 Morse taper, bull nose
4320 steel and fitted together with grade 12.9 fasteners, allowing the potential for interchangeable tooling for future experiments.

The method chosen to hold the blank to the mandrel during forming was to use 3 compression clamps which force the inner diameter of the blank onto the outer diameter of the mandrel by pushing on the blank flange. The main functional elements of one of the clamps in the clamping system are shown in the top right inset of Fig. 2.9. To engage the clamp, first the clamp bracket is located and fixed to the mandrel with the means of a tapered die pin, passing through the mandrel weldment and the clamp bracket. Clamp element 2 is then fixed perpendicularly to the clamp bracket by a captive shoulder bolt. Afterwards, clamp element 2 is actuated against clamp element 1 with a M16 socket cap screw which passes through the mandrel weldment and threads into the clamp bracket. Tightening this bolt induces clamp element 2 to push the engagement block (element 1) against the face of the blank flange. Fixed to clamp element 2 by socket head cap screws, clamp element 3 is brought into contact with the flange opposite to element 1 as the blank approaches the target forming temperature. Operation of the clamping system requires periodic tightening of the M16 socket cap screws as the blank is heated, with the final ‘hot’ configuration shown in the lower inset of Fig. 2.9 This design accounts for a large range of thermal expansion and maximizes operator access.

2.2.3 Roller and tool stand

The tool stand received with the lathe consisted of three main components: a tool holder mounted on a pivot block, which in turn was fixed to a riser communicating with the saddle. This tool stand was originally intended to hold up to four turning tools orthogonal to the workpiece by a locking pin mechanism located in the pivot block. This mechanism was modified such that the tool holder could be fixed at arbitrary angles from the \( R \) direction by the installation of a jam nut assembly, as shown in Fig. 2.10. A AISI–8620 roller with a diameter of 120 mm and a nose radius of 10 mm was set on an axle with a thrust bearing assembly packed with high temperature grease\(^4\). The AISI–4320 axle was held in place by a pinch-bolt on a bar held captive by AISI–1035 lateral bracing members, which was then mated to the tool holder. The tool stand angle was fixed at 15° from the \( R \) direction for all experiments. The temperature of the roller remained below that of the mandrel, as it was only

\(^4\)Dow Corning Molykote BR2 Plus
directly heated by contact with the workpiece during the course of forming.

**Figure 2.10:** Roller and tool stand detail.

### 2.2.4 Heating system

In order to create a flexible heating system which could be scaled as needed, propane heating was selected as the principal heating method. The heating system consisted of a bank of four propane torch assemblies that are installed on the mast of the lathe. The torches are held in a framework that permits adjustment of the various torch positions and their distances away from the blank (Fig. 2.11). The total heat output of the torch tips employed was rated at 82 kW. Propane is supplied through an automotive-type, nominally closed solenoid and regulator at 34.5 kPa, and is distributed to the torches with a manifold containing individual ball valves. The control solenoid permitted rapid cycling of the propane supply, but the torches were lit manually. The final position of the torches was arrived at by making small adjustments until the blank heated rapidly and uniformly.

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5Rothenberger/Exact model 3135
2.2.5 Rotary DAQ

In order to monitor instrumentation installed on the rotating tooling and workpieces, a ruggedized DAQ solution was necessary that could be operated remotely. To meet these needs, a DAQ system was designed to be directly connected to the spindle and spin with the workpiece (Fig. 2.8b). All of the DAQ components are set on a freely rotating shaft mounted to a coupler that directly communicates with the lathe spindle. Clockwise and counterclockwise angular acceleration/deceleration of the DAQ is dampened with a series of four opposed gas springs, acting about the shaft between the DAQ components and the coupler.

The electrical components of the system consist of a USB DAQ board which is connected to a solid state, small form factor wireless computer. Power for both the DAQ board and the computer is delivered by 12 D-cell NiMH rechargeable batteries. As needed, the computer running logging software is operated remotely to monitor TCs in the mandrel (Fig. 2.9) and blank via a Virtual Network Control (VNC) connection during forming operations.

2.2.6 Experimental forming methodology

Several commissioning exercises were undertaken before forming experiments took place. The rotary tooling was installed on the lathe spindle and adjusted until the outer diameter of the mandrel

6Measurement Computing 2416
7National Instruments LabVIEW
had 0.5 mm circular runout along the length of the forming region with the tailstock centre engaged. This was measured by a dial indicator after wet polishing the surface with 400 grit Si-C paper. The mandrel surface was preheated to approximately 150°C with the heating system, and was then sprayed with a colloidal graphite coating to aid in blank removal post-forming.

A blank was preheated to the same temperature in a box furnace, and the inner diameter was sprayed with refractory-type coating. This coating was applied to assist in blank removal as well as to suppress heat transfer between the blank and the mandrel, diminishing blank heating times. The blank and tooling were then allowed to cool to ambient conditions.

A single blank had 2 type-K TC junctions peened into the surface, offset on the circumference by 30° at different axial locations, and was fitted to the mandrel. These TCs were monitored in conjunction with the other 10 TCs embedded near the surface of the mandrel (Section 2.2.2) to determine the characteristic temperature profile through both the blank and mandrel during preheating and forming exercises. The heating system was engaged, and the mandrel was rotated at ∼20 RPM. Heat was applied until clamp element 2 (Fig. 2.9) loosened. The torches were extinguished via the control solenoid, the clamps were then reseated and tightened before the torches were relit and

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**Figure 2.12:** Rotary DAQ detail.
heating resumed. This continued until a target temperature of 350°C was reached. The clamps were further tightened, the heating system was re-engaged and the mandrel was spun at 281 RPM. The \textit{TC} record for this exercise is presented in Fig. 2.13, where dashed lines indicate where the mandrel rotation and heating system were disengaged to tighten the clamps, and then heating resumed. The solid arrow indicates the point where the mandrel rotation rate was increased from 20 to 281 RPM.

This exercise showed that there was little temperature variation in the blank (within 20°C), with the \textit{TC} closest to the mandrel interface reading the coldest temperature. The channel exhibiting both the highest mean temperature and largest heating rate on the mandrel was located immediately below the blank, with temperature and heating rate dropping axially. There was a 1-2°C difference in temperature between the \textit{TC}s offset circumferentially on the mandrel, and due to this minor difference, the temperature from these \textit{TC}s has been presented as an average value (TM3-7 in Fig. 2.13). Increasing the mandrel speed from the preheating rate of 20 RPM to the forming rate of 281 RPM decreased the heating rate in both the mandrel and blank. The mandrel temperature continued
to increase, while the blank temperature remained nearly constant at the TB2 location over twice the length of time necessary for forming.

Blanks intended for forming did not have TCs installed owing to the interference with forming. Non-contact blank surface temperature measurements were attempted using IR-TCs as per Mori et al. [77] using commercial models\(^9\) however it was found that the low emissivity of the blank material and surface irregularities precluded accurate measurements as compared to a contact method. Therefore, surface temperature measurements of the blank were performed manually with a type-K TC surface probe\(^10\) every 3 minutes during heating at TB1 and TB2 locations, as well as immediately before and after forming. Preheating the blanks to the target temperature of 360±8°C for forming took between 17-23 minutes. The variance in blank heating time is attributed to mandrel fitment; blanks better conforming to the mandrel surface took longer to heat up as heat transfer to the mandrel was improved in spite of the refractory-type coating.

Once the blank was at the appropriate temperature, the spindle speed was increased to 281 RPM and the roller was brought into contact with the blank at approximately 30 mm per minute. The longitudinal thread-cutting feed screw was then engaged to move the roller axially at a rate of 0.21 mm per revolution with the torches engaged to lessen heat loss. These parameters were selected based on information provided on the commercially formed wheels (Section 2.1.4), previous experience with commercial forming operations [82], the rating of the lathe motor and results of calculations made during tooling design. Once the spinning pass was complete, the clamps and blank were removed from the mandrel and left to air cool, avoiding potential distortion from quenching.

This procedure was repeated to produce three blanks with increasing levels of deformation summarily described in Table 2.3 and Fig. 2.14. The specimen with the least deformation (Fig. 2.14b) corresponds to conditions normally seen in spinning operations, and the mid-deformed blank corresponds to conditions of ‘overspinning’ as the start of the forming operation was moved axially towards the fixed end of the blank (Fig. 2.14c). The peak deformed specimen had two forming passes applied, one corresponding to the mid-deformed blank and another with a deformation profile approaching the conditions of flow forming (Fig. 2.14d). The deformation in all cases was such that the inner diameter of the forming region did not contact the mandrel surface for all experiments.

---

\(^9\)Exergen iRT/c.1X-K-440F/220C and iRT/c.10A

\(^10\)Omega model number 88108
During the course of the process, the workpiece temperature remained above 340°C for all trials.

**Table 2.3:** List and description of EFA experiments based on initial location of roller in terms of the initial axial location and penetration into the workpiece, \( P \).

<table>
<thead>
<tr>
<th>Deformation level</th>
<th>Axial forming start (mm)</th>
<th>( P ) (mm)</th>
<th>Passes</th>
</tr>
</thead>
<tbody>
<tr>
<td>Low</td>
<td>100</td>
<td>~0.01</td>
<td>1</td>
</tr>
<tr>
<td>Mid</td>
<td>75</td>
<td>~0.01</td>
<td>1</td>
</tr>
<tr>
<td>High/peak</td>
<td>75 &amp; 44</td>
<td>~0.01 &amp; 1</td>
<td>2</td>
</tr>
</tbody>
</table>

![Figure 2.14: Initial and final cross-sectional geometries of workpieces. A depiction of the initial location of the roller nose is also provided.](image)

**Figure 2.14:** Initial and final cross-sectional geometries of workpieces. A depiction of the initial location of the roller nose is also provided.

### 2.3 Material characterization

Microstructural changes in the material due to processing and defect features following fatigue testing were tracked with several methods. Optical microscopy was used to characterize microstructural
features in conjunction with multiple image analysis tools. SEM techniques were employed primarily in fractography studies and to analyze the effect of heat treatment on precipitation and eutectic particle changes. Altered mechanical properties due to processing were tracked via extensive Vickers hardness tests. Each of these techniques are discussed in greater detail in the following sections.

2.3.1 Sample preparation

Metallographic specimens from each type of material employed in this research were extracted using different combinations of bandsaw and Electro-Discharge Machining (EDM) sectioning. All metallographic samples were manually polished with 120, 320 and 600 grit Si-C paper. This was followed by two secondary polishing steps with 6 and 1 µm alumina. Specimen size permitting, final polishing employed a vibratory polishing operation at 60 Hz for 180 minutes with 0.06 µm colloidal silica. Some samples were mounted in epoxy to aid in polishing.

While the specific schedule is unknown for the material that was received in the T6 condition, the characteristic T6 schedule performed for all material in the present work is as follows: solutionized at 538°C for 3 hours, quenched in water at 60±5°C, and artificially aged at 150°C for 3 hours with no natural ageing. Both solutionizing and artificial ageing were performed in separate box furnaces. A lesser number of small specimens were heat treated for various times and temperatures in a nitrate salt bath (60% potassium nitrate, 40% sodium nitrate). These latter samples were instrumented with a thermocouple (TC) which was used to continuously monitor the temperature during heat treatment.

2.3.2 Optical microscopy

Photomicrographs were taken of the metallographic specimens with several different pieces of equipment. These included:

- a Nikon Epiphot 300 inverted microscope and QImaging digital camera, with Clemex Vision PE software;
- a Nikon Eclipse MA200 inverted microscope and Nikon DS Fi1 digital camera, with NIS-Elements software;

11 Buehler Vibromet II
12 System Three Cold Cure
• a Leica DM 1LM inverted microscope and Leica DC100 digital camera, with Leica software.

Additionally, low magnification images were acquired with a Vickers-Armstrong microscope fitted with an adapter and a digital SLR camera, as discussed further in Section 2.3.4.

Microstructural characterization (DAS, porosity, eutectic-Si particle characteristics) was carried out on digital images with several software tools including Clemex Vision PE software, Fiji [94] and custom tools written in MATLAB employing the Image Analysis Toolbox. For wedge characterization, 2D characteristic measurements of microstructural features were made on a series of images representing a composite area greater than 85 mm². Locations and quantities of measurements made on formed material are provided in further chapters.

DAS measurements were carried out with a program written in MATLAB, shown schematically in Fig. 2.15. The program first loaded a micrograph and then accepted the graphical input of the centre of active secondary dendrite arms. A line of best fit was applied to the arm centres, and the centres were then projected on this line. The distance between each arm centre on this line was then calculated and scaled accordingly for each discrete measurement. This was repeated at all sites where active secondary dendrite arms could be identified, and coincides with the methodology discussed in the literature [14].

Porosity and defect sizes were characterized by the \( \sqrt{\text{area}} \) parameter discussed previously in Section 1.4.1. This was accomplished by using one of the aforementioned tools to threshold the image such that the feature was completely highlighted. The resulting area of the feature was then calculated based on the number of pixels and scaled accordingly. An example result of this process is provided in Fig. 2.16. In a similar manner, the eutectic particles had their areas measured, and Equivalent Circle Diameter (ECD) was calculated based on this area. Particle aspect ratio was calculated by applying best-fit ellipses to binarized images of the eutectic, as shown in Fig. 2.17. The aspect ratio was then calculated by dividing the length of the major axis of each ellipse by the length of the minor.

2.3.3 SEM techniques

SEM was carried out on several pieces of equipment. The principal tool employed throughout was a Hitachi S3000N scanning electron microscope, which was used for general imaging purposes, fractography studies as well as Energy-Dispersive X-ray spectroscopy (EDX). For EDX analysis, an
Advanced Analysis Technologies detector was employed. This detector has an energy resolution of 133 eV and was operated at 2500 to 3000 samples per second with the SEM accelerating voltage at 10 keV. Two other scanning electron microscopes were also employed for fractography studies, including a Hitachi S2300 and JEOL JSM 6400. All were operated in backscatter electron mode, with accelerating voltages ranging from 7 keV to 20 keV.

Some polished specimens were deep-etched to examine eutectic Si morphology changes during purely thermal and thermal-mechanical processing. This was accomplished by immersing the speci-
**Figure 2.17:** Example of best-fit ellipses applied to eutectic particles to measure aspect ratio.

**Figure 2.18:** Example of manual defect size measurement carried out on the fracture surface of a fatigue specimen containing an artificial defect ($\sqrt{\text{area}} = 416 \mu m$).

imens in a strong Keller’s etchant composed of 10% HF and 5% HC acid by volume for 50 minutes, as suggested by Colley [18]. These specimens were then imaged via SEM.

For tabulating defect sizes employed in the fractography study, a similar process to that employed for the porosity measurements discussed above. However, the SEM images of the fracture surface did not have the contrast necessary for automated processing. The defects were therefore manually traced and the result scaled accordingly, as demonstrated in Fig. 2.18.
2.3.4 Hardness testing

Changes in mechanical properties due to thermal and thermomechanical processing were quantified via extensive macrohardness ($H_{V5}$) measurements, and a lesser number of microhardness ($H_{V0.01}$) measurements to track changes in the eutectic and $\alpha$-Al phases. Macrohardness measurements were performed on a Vickers-Armstrong Hardness Test Machine (VHTM) which was customized to assist with specimen positioning and measurement processing. Microhardness measurements were performed with a Buehler Micromet 3 Micro Hardness Tester. All hardness tests were conducted and validated according to the practices detailed in Appendix A. Measurements of indentation diagonals were accomplished either directly with the filar micrometer on each machine, or indirectly on digital images taken of the indentation site employing the analysis tools listed above.

The VHTM used for macrohardness measurements was augmented with a custom, Computer Numeric Control (CNC) stage (Fig. 2.19). This was accomplished through the design of a double axis rigid stage with ball-screw assemblies driven by high-resolution stepper motors and translator circuits. The translator circuits were in turn controlled by a computer running interpreter programs. The existing microscope on the VHTM was augmented such that the eyepiece lens and filar micrometer was replaced with a digital SLR camera. The camera was controlled with software running on a second computer, employed to image each indentation site. This system provided the ability to program the stage to repeatedly position a specimen to within 20 µm accounting for ball-screw backlash, perform an indentation, and then digitally image the result.

In order to create complete hardness profiles of the processed material, blanks with various processing histories were first sectioned, mounted and polished according to the methodology described in Section 2.3.1. These section profiles were then scanned at 1200 DPI. The scans were processed to provide specimen outlines. The outlines were then uniformly meshed with triangular elements, such that the element edge lengths were a minimum of $\pi$ times the maximum indent diagonal of 450 µm, using a MATLAB script. The resulting mesh was then manually verified and edited such that the distance between each node was at least 450π µm apart. The nodal coordinates were then programmed into the stage controller, and indents were placed at each node. The controller

13 Ability Systems Corporation Indexer LPT and G Code Controller
14 Canon EOS Rebel T2i fitted with a Martin Microscope MM-SLR
15 Canon EOS Utility software
16 Employing a Hewlett Packard ScanJet 4200C
program was run again following indentation to collect images of each indent. These digital images were then processed with MATLAB scripts to generate measurements of the indent diagonals. This automated method of measuring indents was found to agree with manual measurements within 2% over 500 indents.

2.3.5 Compression testing methodology

Deformation tests were conducted on a Gleeble\textsuperscript{17} 3500 thermomechanical simulator fitted with isothermal tungsten carbide Iso-T (isothermal) anvils, treated with a nickel-based lubricant\textsuperscript{18}. The specimens were deformed by actuating the anvils hydraulically. During each test, the temperature was controlled and recorded with a type-K TC mounted to the centre of each specimen. Instantaneous diametral deformation was measured with a Linear Variable Differential Transformer (LVDT) actuated by quartz mandibles. This arrangement is depicted in Fig. 2.21.

\textsuperscript{17}Gleeble is a trademark of Dynamic Systems, Inc., Poestenkill, NY.
\textsuperscript{18}Loctite 77124 Nickel Anti-Seize
The specific test procedure followed for each test was:

1. The sample was first loaded between the platens with a preload of less than 0.5 kN.

2. The hydraulic actuator was retracted by 2 mm to allow for thermal expansion. Pre-load on the sample was maintained via friction.

3. Joule heating was applied to the sample at a rate of $5^\circ C \text{s}^{-1}$ and then held at the target test temperature for 60 seconds.

4. The sample was deformed at the prescribed deformation rate to target strains of 0.2, 0.4 and 0.55 based on the specimens nominal length at ambient temperature. Deformation rates were calculated based on specimen dimensions at ambient temperatures.

5. Based on the prescribed target strain rate, the data acquisition rate from the three transducers was greater than 200 samples per unit strain during deformation.
At elevated strain rates, heat generated during deformation was greater than could be removed by the anvils. This caused a slight increase ($\sim 10^\circ$C) in the average deformation temperature $T_d$, and was more predominant at lower temperatures and higher strain rates. Although the preload at all temperatures was kept below 0.5 kN, creep occurred at elevated temperatures during the hold stage prior to deformation. While the creep rate differs with temperature, the amount of creep was nearly identical from test to test as the pre-deformation heating cycle was the same. Differing amounts of thermal expansion owed to variations in specimen geometry render quantification of the amount of creep prior to deformation intractable. However, due to the low load each sample experienced some amount of pre-deformation. This amount of creep is small and has been considered part of the thermal expansion included in the initial diameter, $D_0$. Uniform plastic deformation conditions were assumed to prevail in all cases. Therefore, the instantaneous true strain $\varepsilon$ was found from the instantaneous diameter $D$ according to:

$$\varepsilon = -2\ln\frac{D}{D_0}$$

(2.1)

The instantaneous strain record was then differentiated to provide the instantaneous strain rate, $\dot{\varepsilon}$.
Figure 2.22: Fatigue specimen types: multiaxial (a), torsion (b), straight-gauge uniaxial (c) and radius-gauge uniaxial (d & e).

The average strain rate, $\dot{\varepsilon}_a$, is taken as the mean strain rate during deformation, and was only used to qualify each test. This is also true for the average temperature, $T_a$, as well. All experimental data points of temperature, load, strain and strain rate have been incorporated in the analysis that is presented in Chapter 3.

2.3.6 Fatigue testing methodology

Numerous fatigue sample types were employed to characterize both the multiaxial and uniaxial fatigue behaviour of material with different processing histories. All fatigue tests in the present work were conducted under fully reversed loading ($R_L = -1$). These different specimen types were necessary to facilitate testing with different apparatus and sample sources, as given in Fig. 2.22. Specimen types ‘a’ through ‘c’ were drawn from wedge and wheel castings, shown in Fig. 2.4a and 2.4c, respectively. Specimen type ‘d’ was drawn from commercially formed material, as shown in Fig. 2.6b. Specimen type ‘e’ was extracted from EFA workpieces, at locations given in Fig. 2.3. The testing apparatus, testing frequency and applicable loading condition is summarized in Table 2.4.

The tests which employed the Instron servo-hydraulic apparatus, located in the Laboratory of
Table 2.4: Fatigue test equipment, operating conditions and specimen type

<table>
<thead>
<tr>
<th>Machine</th>
<th>Frequency (Hz)</th>
<th>Sample type (Fig. 2.22)</th>
<th>Loading</th>
</tr>
</thead>
<tbody>
<tr>
<td>Instron servo-hydraulic</td>
<td>11</td>
<td>a</td>
<td>Tension-torsion</td>
</tr>
<tr>
<td>Amsler-Vibrophone</td>
<td>45</td>
<td>b, c</td>
<td>Tension, torsion</td>
</tr>
<tr>
<td>Sonntag eccentric</td>
<td>33</td>
<td>d, e</td>
<td>Tension</td>
</tr>
</tbody>
</table>

Mechanics and Materials Physics at the École Nationale Supérieure de Mécanique et d’Aérotechnique (ENSMA), were conducted under sinusoidal load control conditions to failure. The Sonntag apparatus, available in the Department of Materials Engineering at UBC, precluded this type of control, and therefore testing was conducted under a constant load amplitude as measured with a load cell at the start of testing. The Sonntag tests were stopped based on a change in elastic response as monitored by an LVDT and the use of die penetrant for verification. A similar stopping criteria was inherent for the resonant tests, conducted with the Amsler-Vibrophone machine (located at ENSMA) whereby the test was stopped when a significant change in resonant frequency was recorded. Specimens that were not fractured at the conclusion of testing were cooled with liquid Ni and manually broken for fractographic analysis purposes.

The principal utility of the Instron and Amsler-Vibrophone tests was to investigate multiaxial A356–T6 HCF behaviour. Multiaxial testing was conducted using the step technique originally outlined by Maxwell and Nicholas [95], to target the endurance limit \((\sigma_f, \tau_f)\) at \(10^6\) cycles. Both tension and torsion components were imposed in-phase. With this technique, each specimen undergoes cyclic loading at an initial load level estimated based on previous testing experience. Samples that do not fail after \(10^6\) cycles are then cycled again at a higher stress amplitude. For specimens that failed before \(10^6\) cycles without a preceding loading step, the endurance limit was estimated using a Basquin coefficient, \(b\), of 0.17 based on experimental results found in the literature [47, 55, 58, 96, 97]. For specimens that failed before \(10^6\) cycles but after a minimum of one loading step, the endurance limit was calculated as the average stress amplitude of the failure and prior step.

The Sonntag apparatus was used to generate uniaxial \(S – N\) or Wöhler curves, with no step testing. All data obtained with this apparatus consists of ‘runout’ results, or results from employing a single load amplitude. The same HCF target of \(10^6\) cycles was employed for these uniaxial tests,
with minimum loads identified from the multiaxial testing regimen.
In order to address the lack of information on the constitutive behaviour of as-cast A356 at elevated temperatures across a range of strain rates, a large number of isothermal compression tests were conducted according to the methodology described in Section 2.3.5. The data generated with these tests supported the development of a comprehensive constitutive equation. The two main constitutive frameworks that were considered in developing this equation were an extended Ludwik-Hollomon expression and a Kocks-Mecking approach. This chapter presents the experimental data, analysis following each constitutive framework, and a discussion of the performance of each.

3.1 Experimental results

In total, 55 successful compression tests were carried out with a range of temperatures, strains and strain rates. Temperatures of 30°C, 100°C, and between 200 to 500°C in increments of 50°C and strain rates of 0.1, 1, 5 and 10 s\(^{-1}\) were targeted. Predominantly in the high strain rate range, the simulator was not entirely capable of delivering test results without overshooting the target strain or strain rate. Therefore the average strain rates achieved, \(\dot{\varepsilon}_a\), are reported. These rates were calculated based on the strain versus time information, and are grouped in ranges of 0.06 to 0.11 s\(^{-1}\), 0.57 to 0.98 s\(^{-1}\), 3.59 to 5.33 s\(^{-1}\) and 7.58 to 12.24 s\(^{-1}\), which are referred to further as \(R_1\), \(R_2\), \(R_3\) and \(R_4\), respectively. Flow curves for each of these strain rate ranges are shown in Figure 3.1, while data for all test conditions is summarized in Table 3.1. The variation in target strain rates and strains was attributed to thermal expansion and variation in specimen length, therefore the actual plastic strain achieved (\(\varepsilon_p\)) as calculated from the LVDT has been reported. A number of additional tests were performed at the highest target strain rates in each temperature range, however these tests

---

were not successful owing to non-uniform deformation and barreling. Tests where these conditions developed were defined as failed tests and their data is not shown nor used for subsequent analysis.

The yield stress for each test (shown in Figure 3.1 as circle symbols on each curve) was found with a 0.2% offset method employing temperature-corrected shear and elastic modulii [98]. \( \mu \) and \( E \), according to:

\[
\mu = \mu_0 \left( 1 + \frac{T - 300}{T_{melt}} \right) \frac{T_{melt}}{\mu_0} \frac{d\mu}{dT}
\]

\[
E = 2\mu (1 + \nu)
\]

Table 3.1: Compression test characteristics: \( T_\alpha \), \( \dot{\varepsilon}_a \), \( \sigma_y \) and \( \varepsilon_p \) attained for all tests (continued)

(a) \( \sim 30^\circ C \)

<table>
<thead>
<tr>
<th>No.</th>
<th>( T_\alpha ) (°C)</th>
<th>( \dot{\varepsilon}_a ) (s(^{-1}))</th>
<th>( \sigma_y ) (MPa)</th>
<th>( \varepsilon_p )</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>33.4</td>
<td>0.06</td>
<td>97</td>
<td>0.13</td>
</tr>
<tr>
<td>2</td>
<td>36.1</td>
<td>0.07</td>
<td>112</td>
<td>0.37</td>
</tr>
<tr>
<td>3</td>
<td>34.0</td>
<td>0.07</td>
<td>109</td>
<td>0.18</td>
</tr>
<tr>
<td>4</td>
<td>41.0</td>
<td>0.57</td>
<td>114</td>
<td>0.42</td>
</tr>
<tr>
<td>5</td>
<td>38.1</td>
<td>0.71</td>
<td>102</td>
<td>0.19</td>
</tr>
<tr>
<td>6</td>
<td>36.8</td>
<td>0.71</td>
<td>105</td>
<td>0.19</td>
</tr>
<tr>
<td>7</td>
<td>37.5</td>
<td>0.72</td>
<td>113</td>
<td>0.19</td>
</tr>
<tr>
<td>8</td>
<td>33.7</td>
<td>0.74</td>
<td>107</td>
<td>0.14</td>
</tr>
<tr>
<td>9</td>
<td>45.4</td>
<td>3.95</td>
<td>137</td>
<td>0.35</td>
</tr>
</tbody>
</table>

(b) \( \sim 100^\circ C \)

<table>
<thead>
<tr>
<th>No.</th>
<th>( T_\alpha ) (°C)</th>
<th>( \dot{\varepsilon}_a ) (s(^{-1}))</th>
<th>( \sigma_y ) (MPa)</th>
<th>( \varepsilon_p )</th>
</tr>
</thead>
<tbody>
<tr>
<td>10</td>
<td>103.7</td>
<td>0.07</td>
<td>95</td>
<td>0.36</td>
</tr>
<tr>
<td>11</td>
<td>102.4</td>
<td>0.07</td>
<td>104</td>
<td>0.17</td>
</tr>
<tr>
<td>12</td>
<td>104.3</td>
<td>0.70</td>
<td>120</td>
<td>0.14</td>
</tr>
<tr>
<td>13</td>
<td>112.4</td>
<td>0.74</td>
<td>102</td>
<td>0.45</td>
</tr>
<tr>
<td>14</td>
<td>106.9</td>
<td>0.77</td>
<td>106</td>
<td>0.20</td>
</tr>
<tr>
<td>15</td>
<td>104.7</td>
<td>0.77</td>
<td>117</td>
<td>0.15</td>
</tr>
<tr>
<td>16</td>
<td>106.6</td>
<td>0.78</td>
<td>111</td>
<td>0.19</td>
</tr>
<tr>
<td>17</td>
<td>121.5</td>
<td>4.23</td>
<td>121</td>
<td>0.45</td>
</tr>
<tr>
<td>18</td>
<td>126.9</td>
<td>8.27</td>
<td>117</td>
<td>0.39</td>
</tr>
</tbody>
</table>

(c) \( \sim 200^\circ C \)

<table>
<thead>
<tr>
<th>No.</th>
<th>( T_\alpha ) (°C)</th>
<th>( \dot{\varepsilon}_a ) (s(^{-1}))</th>
<th>( \sigma_y ) (MPa)</th>
<th>( \varepsilon_p )</th>
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</thead>
<tbody>
<tr>
<td>19</td>
<td>201.3</td>
<td>0.08</td>
<td>85</td>
<td>0.19</td>
</tr>
<tr>
<td>20</td>
<td>199.5</td>
<td>0.08</td>
<td>84</td>
<td>0.32</td>
</tr>
<tr>
<td>21</td>
<td>204.9</td>
<td>0.72</td>
<td>84</td>
<td>0.31</td>
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<tr>
<td>22</td>
<td>203.7</td>
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<tr>
<td>23</td>
<td>188.7</td>
<td>3.86</td>
<td>105</td>
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<tr>
<td>24</td>
<td>201.6</td>
<td>7.58</td>
<td>111</td>
<td>0.27</td>
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</table>

(d) \( \sim 250^\circ C \)

<table>
<thead>
<tr>
<th>No.</th>
<th>( T_\alpha ) (°C)</th>
<th>( \dot{\varepsilon}_a ) (s(^{-1}))</th>
<th>( \sigma_y ) (MPa)</th>
<th>( \varepsilon_p )</th>
</tr>
</thead>
<tbody>
<tr>
<td>25</td>
<td>249.8</td>
<td>0.08</td>
<td>95</td>
<td>0.42</td>
</tr>
<tr>
<td>26</td>
<td>251.6</td>
<td>0.08</td>
<td>91</td>
<td>0.20</td>
</tr>
<tr>
<td>27</td>
<td>247.5</td>
<td>0.08</td>
<td>89</td>
<td>0.33</td>
</tr>
<tr>
<td>28</td>
<td>257.9</td>
<td>0.70</td>
<td>101</td>
<td>0.44</td>
</tr>
<tr>
<td>29</td>
<td>259.9</td>
<td>4.39</td>
<td>106</td>
<td>0.40</td>
</tr>
<tr>
<td>30</td>
<td>258.5</td>
<td>8.24</td>
<td>111</td>
<td>0.30</td>
</tr>
</tbody>
</table>

(e) \( \sim 300^\circ C \)

<table>
<thead>
<tr>
<th>No.</th>
<th>( T_\alpha ) (°C)</th>
<th>( \dot{\varepsilon}_a ) (s(^{-1}))</th>
<th>( \sigma_y ) (MPa)</th>
<th>( \varepsilon_p )</th>
</tr>
</thead>
<tbody>
<tr>
<td>31</td>
<td>297.8</td>
<td>0.08</td>
<td>90</td>
<td>0.33</td>
</tr>
<tr>
<td>32</td>
<td>300.5</td>
<td>0.09</td>
<td>85</td>
<td>0.21</td>
</tr>
<tr>
<td>33</td>
<td>307.6</td>
<td>0.80</td>
<td>101</td>
<td>0.47</td>
</tr>
<tr>
<td>34</td>
<td>310.2</td>
<td>3.59</td>
<td>108</td>
<td>0.34</td>
</tr>
</tbody>
</table>
Table 3.1: Compression test characteristics: $T_a$, $\dot{\varepsilon}_a$, $\sigma_y$ and $\varepsilon_p$ attained for all tests

<table>
<thead>
<tr>
<th>No.</th>
<th>$T_a$ (°C)</th>
<th>$\dot{\varepsilon}_a$ (s$^{-1}$)</th>
<th>$\sigma_y$ (MPa)</th>
<th>$\varepsilon_p$</th>
</tr>
</thead>
<tbody>
<tr>
<td>35</td>
<td>350.1</td>
<td>0.08</td>
<td>60</td>
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$\sim$ 350°C

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<tr>
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<td>56</td>
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<td>407.2</td>
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<td>46</td>
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$\sim$ 400°C

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<th>$\varepsilon_p$</th>
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$\sim$ 450°C

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<th>$\varepsilon_p$</th>
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<td>502.8</td>
<td>12.24</td>
<td>38</td>
<td>0.19</td>
</tr>
</tbody>
</table>

$\sim$ 500°C

Based on pure aluminum, with $T$ in Kelvin, the temperature dependence of modulus ($\frac{\mathrm{du}}{\mathrm{dT}}$) equal to -0.5, the shear modulus at 300 K ($\mu_o$) equal to $2.64 \times 10^4$ MPa and $\nu$ equal to 0.33. The liquidus temperature of A356 (612.5°C [7]) was employed as $T_{\text{melt}}$. A relationship based on pure aluminum is appropriate as the temperature dependence of $E$ for aluminum alloys is relatively insensitive to solute [99].

The flow stress data shows appreciable strain hardening at all strain rates with little strain rate sensitivity until approximately 300-350°C. Above this temperature, there is little to no strain hardening and increasing strain rate sensitivity owing to dynamic recovery. This is also demonstrated by the trend in $\sigma_y$ (refer to Table 3.1). For the low temperature tests, $\sigma_y$ remains relatively insensitive to strain rate as compared to tests at elevated temperatures. To further demonstrate the change in strain-rate sensitivity, Figure 3.2 shows selected experimental flow stress curves at constant temperatures for $\dot{\varepsilon}_a \in R_1, R_4$. For the tests conducted at ~ 30 and ~ 100°C, $\sigma_y$ for $\dot{\varepsilon}_a \in R_1, R_2$ was found to be approximately 108 MPa varying by ±10%. As the systematic error produced by the transducers used in this study is two orders of magnitude lower than this range, the bulk of this variance is attributed to small variations of DAS, coarse Mg2Si precipitates and segregation in the as-cast structure. Owing to the low area percent and the fact that the applied load will tend to close pores,
Figure 3.1: Characteristic experimental flow stresses. Measured stress-strain results for all $\dot{\varepsilon}$ ranges is presented. Average test temperature $T_a$ is reported in $^\circ$C.

Porosity is not considered to influence the compressive strength of the material [39].

For comparison, the $\sigma_y$ observed at $\sim 30^\circ$C in this study is half that of this material in the T6 condition [100] and twice that of the solutionized condition [39]. The main cause of this difference is surmised to be the result of the distribution and state of Mg$_2$Si precipitates. In the solutionized condition, any precipitates formed during casting have been dissolved and the solute atoms are in solution. This results in a substantively smaller $\sigma_y$ compared to the as-cast condition. The artificial ageing step in the T6 process serves to nucleate and grow a more evenly distributed variety of Mg$_2$Si precipitates as compared to the coarse and randomly distributed variety in the as-cast condition. This
in turn creates a more effective barrier to dislocation glide, resulting in a higher $\sigma_y$.

For tests conducted at the lowest strain rate, $R_1$, there is evidence of slight strain-softening at temperatures above 300°C (Figure 3.1a), which diminishes at higher temperatures for tests in the same strain rate range. This behaviour was also observed in the 300°C test for $R_2$ (Figure 3.1b). These results are evidence of the dynamic phenomena known to occur for this alloy system at these temperatures, such as recovery [101]. As such, 300°C has been identified as the start temperature for the transition to strain-rate dependence, the material behaviour may be conservatively characterized such that below 350°C, the material is dislocation interaction dominated. Above this temperature, it is strain rate/diffusion dominated. As there is a temperature/strain-rate range where both dislocation

**Figure 3.2:** Comparison of the experimental flow stress $\sigma$ to flow stresses predicted by the Ludwik-Hollomon $\hat{\sigma}_{LH}$ and Kocks-Mecking $\hat{\sigma}_{KM}$ constitutive expressions.
interaction and diffusion processes are both active, a constitutive relationship that traverses these two regimes will inherently have difficulty describing the flow stress with accuracy in this temperature range [102].

3.2 Constitutive equation development

In general, the phenomenological deformation behaviour of aluminum alloys is dominated by strain hardening at low temperatures, which transitions to a time or strain-rate dependent (creep) response at elevated temperatures. Models of aluminum casting processes consider the material as having some elastic behaviour in addition to either time-dependent creep, or plasticity with strain dependence. The former formulations considered the Sellars-Tegart [103] or Garofalo relationships [40,104-106], while the latter [107-109] used extended Ludwik-Hollomon expressions. Constitutive behaviours of AC wrought aluminum AA1050, AA3104 and AA5182 have also been described with a simultaneous combination of time and strain dependence [110] employing the Ludwik-Hollomon approach.

As standard Sellars-Tegart relationships are not able to effectively describe work hardening observed in the current experimental data at temperatures below 350°C, a Ludwik-Hollomon expression was identified as the most suitable phenomenological model to apply to AC A356. While being quite effective in predicting flow stresses, the main drawback of using phenomenological approaches is that they are only valid within the experimental strain range from which the constitutive expression has been derived. Physically-based models, such as the Zerilli-Armstrong [111] which considers grain size or the more general Kocks-Mecking [112-115], employ fundamental material parameters to predict flow stress. As grain size in cast Al-Si alloys is not typically correlated to strength as compared to other microstructural features such as DAS, the Kocks-Mecking approach is likely the best physically-based model to apply for AC material. However, this type of approach has been used primarily to describe materials in the work hardening regime.

While the experimental results may be divided into two discernible regimes, from an analysis standpoint it is necessary to construct an equation or series of equations that successfully predicts the flow stress across both regimes. The following sections describe the development of two models to do so. The analysis was accomplished through linear least squares fitting where possible and a Nelder-Mead technique when non-linear fitting was necessary. The fitting procedure for each
approach will be presented initially, after which the results of each model will be discussed. Fitted coefficients for approach are summarized in Table 3.2.

3.2.1 Extended Ludwik-Hollomon

The extended Ludwik-Hollomon expression is a phenomenological model that is frequently employed for cast aluminum alloys. In order to capture the simultaneous evolution of both strain hardening and strain rate effects with temperature for aluminum, van Haaften et al. [110] proposed an expression with static fitting coefficients, while other authors [108, 109] have employed expressions where the fitting coefficients $K$, $M$ and $N$ are functions of temperature. The temperature dependent, extended Ludwik-Hollomon expression is:

$$\sigma_{LH} = K(T) \varepsilon^N(T) \left( \frac{\dot{\varepsilon}}{\dot{\varepsilon}_1} \right)^M(T)$$

where $\dot{\varepsilon}_1$ is a normalization strain rate of $1 \text{ s}^{-1}$. The advantage of this phenomenological approach is that $N(T)$ and $M(T)$ are both independent functions which when factored from the flow stress, enable $K(T)$ to be determined via regression.

Values of the work hardening parameter, $N$, were found first via the slope of experimental ln $\sigma$-ln $\varepsilon$ curves for each test. The strain-rate term, $M$, was taken to be the slope of ln $\sigma$-ln $\dot{\varepsilon}$ curves built from data at strains between 0.05 and 0.12 with increments of 0.01. The resulting values of $M$ and $N$ versus $T_a$ are shown in Figure 3.3a. The $N$ values plotted versus average temperature demonstrate that there is a near linear trend for all strain rates until approximately $350^\circ$C. As mentioned in Section 3.1, above this temperature, diffusional effects become significant. The values of $N$ for $\dot{\varepsilon}_a \in R_1$ are distinctly negative, while the rest of the strain rates result in $N$ values close to zero. Based on these observations, $N(T)$ was characterized such that:

$$N(T) = \begin{cases} n_1 T + n_2 & T \leq T_t \\ n_3 & T > T_t \end{cases}$$

where $T_t$ is the transition temperature ($\sim 350^\circ$C) between the two regimes. The constants $n_1$ and $n_2$ were found from data corresponding to $T_a < 300^\circ$C and $n_3$ from $T_a \geq 300^\circ$C. $T_t$ is the intersection of the two resulting linear expressions.
As the physical basis for the trend in the $M - T_a$ data (Figure 3.3a) is an observed increase in strain rate sensitivity with temperature, a continuous function has been employed to describe $M(T)$:

$$M(T) = m_1 T^{m_2} + m_3 \quad (3.5)$$

$N$ and $M$ were fit with Eq. 3.4 and 3.5, and then were used to extract values for $K_{LH}$ (Figure 3.3b). The correlation of the strength coefficient with temperature, $K_{LH}(T)$, was also assumed to be a continuous function and was found to be best described by:

$$K_{LH}(T) = k_1 T^2 + k_2 T + k_3 \quad (3.6)$$

The fitted functions $M(T)$ and $K_{LH}(T)$ show good agreement with the calculated $M$ and $K$ values based on experimental values. $N(T)$ describes the experimental values of $N$ much better in the work-hardening regime as compared to the rate-dependent, with the slowest strain rate data being poorly described. Fitted coefficients for each function are provided in Table 3.2.

### 3.2.2 Kocks-Mecking

The Kocks-Mecking hardening model is the most common physically-based approach used to describe constitutive behaviour. The underlying premise of the Kocks-Mecking hardening model is

![Figure 3.3: Extended Ludwik-Hollomon coefficients plotted versus temperature.](image-url)
that the work hardening rate $\Theta = d\sigma / d\varepsilon$ approaches zero at some saturation stress, $\sigma_s$ which is a function of strain rate and temperature. As a result, plotting $\Theta/\mu$ versus $\sigma/\sigma_s$ creates a master curve that is accurate over a large range of hardening such that:

$$\frac{\Theta}{\Theta_0} = f \left( \frac{\sigma}{\sigma_s} \right)$$

Typically, Eq. (3.7) is assumed to have the form of a linear Voce-type relationship:

$$\Theta = \Theta_0 \left( 1 - \frac{\sigma}{\sigma_s} \right)$$

with an integrated form of:

$$\frac{\sigma - \sigma_s}{\sigma_y - \sigma_s} = \exp \left( -\frac{\Theta_0 \varepsilon}{\sigma_s} \right)$$

The saturation stress, $\sigma_s$, is expressed as:

$$\sigma_s = \sigma_{s0} \frac{\mu}{\mu_0} \left( 1 - \left( \frac{g}{g_0} \right)^{1/q} \right)^{1/p}$$

which is a function of a normalized activation energy term $g$, defined as [115]:

$$g = \frac{k_B T}{\mu b^3} \ln \frac{\dot{\varepsilon}_0}{\dot{\varepsilon}}$$

where $\dot{\varepsilon}_0$ is the minimum strain rate that best converges all data to a single function relating $\sigma_s/\mu$ to $g$. Physically, $p$ and $q$ (0 ≤ $p$ ≤ 1 and 1 ≤ $q$ ≤ 2) in Eq. (3.10) represent the shape of dislocation obstacle profiles [114], which in turn decide the values of $\sigma_{s0}$ and $g_0$. These constants are phenomenological [115, 116] and in the present work have not been tailored to distinguish between discrete obstacles (e.g. precipitates) or grain boundaries. The yield strength, $\sigma_y$, can also be expressed as a function of $g$, which renders Eq. (3.9) a $\sigma$-$\varepsilon$ relationship that is a function of strain rate and temperature.

Values of $\sigma_s$ were first identified from linear intercepts of experimental $\Theta - \sigma$ data between $\Theta = 0$ and fully developed yield, $\Theta = \mu/20$ [115, 117]. The result of this process is demonstrated in Figure 3.4 which shows $\Theta/\mu$ versus $\sigma/\sigma_s$ for strain rate ranges $R_1$ and $R_4$. The tabulated $\Theta/\mu$
versus \( \sigma/\sigma_s \) data demonstrates a continuous function satisfying Eq. 3.7. While the large strain, high temperature data below \( \Theta = \mu/20 \) may be approximated by Eq. 3.8, the overall data distribution is non-linear. Above \( \Theta = \mu/20 \), the data corresponding to small strains and low temperature represents a large initial work hardening rate. The average initial work hardening rate, \( \Theta_0 \), across all tests has been identified as 2875 MPa, which is significantly higher than pure aluminum in an annealed state where \( \Theta_0 \) is in the range of 1120-1720 MPa [118].

![Figure 3.4: Compression test work hardening rate versus flow stress. \( \Theta/\mu \) versus \( \sigma/\sigma_s \) for \( \dot{\varepsilon}_a \in R_1 \) and \( R_4 \) at all temperatures is shown. The horizontal line indicates \( \Theta = \mu/20 \).](image)

Figure 3.5 shows a plot of \( \sigma_s/\mu \) versus \( g(\dot{\varepsilon}_a, T_a) \), and the resulting fitted expressions. Here, the baseline values of the phenomenological constants identified by Kocks and Mecking [115], \( p = 1/2 \), \( q = 2 \) and \( \dot{\varepsilon}_0 = 10^7 \text{s}^{-1} \) were employed. The constants \( \sigma_{s0}/\mu_0 \) and \( g_0 \) were determined from the \( y \) and \( x \) intercepts, respectively, of the fitted \( \sigma_s/\mu \) versus \( g^{1/4} \) relationship. A sensitivity analysis conducted on the values of \( p \), \( q \) and \( \dot{\varepsilon}_0 \) did not show appreciable improvement of the fit within the range defined by their physical basis.

Figure 3.5 also shows the relationship between \( \sigma_y/\mu \) and \( g \). Coinciding with the two regimes identified in Section 3.1, there is an identifiable threshold value of \( g \), where \( \sigma_y/\mu \) decreases; \( \sigma_y/\mu \) remains constant until this threshold is reached. For larger values of \( g \) beyond this transition, \( \sigma_y/\mu \) is taken to approach \( g_0 \). In order to characterize this transition, a bi-linear function was fitted to data
corresponding $g^{1/q} > 0.6$ and $g^{1/q} \leq 0.6$ for each linear component:

$$
\sigma_y(g) = \begin{cases} 
  \mu c_y^{1/p} & g < \left( \frac{c_y}{C_y} + g_0^{1/q} \right)^{1/p} \\
  \mu \left( C_y \left( g^{1/q} - g_0^{1/q} \right) \right)^{1/p} & g \geq \left( \frac{c_y}{C_y} + g_0^{1/q} \right)^{1/p}
\end{cases}
$$

(3.12)

where $C_y$ and $c_y$ are constants. The values of all constants are given in Table 3.2.

As both $\sigma_s$ and $\sigma_y$ are able to be scaled as a function of temperature and strain rate, an expression for flow stress is possible through Eq. 3.9. However, this relationship is incapable of describing a large portion of the flow stress accurately, owing to the particularly high initial work hardening rate, and the non-linear strain-hardening rate at low temperatures and strains observed in the experimental data (Figure 3.4). Having an accurate integrated form is necessary as many commercial FEA codes require expressions for equivalent stress and strain. Using a modified equation to describe the $\Theta - \sigma$ behaviour [113] has been found to provide a better phenomenological representation of $\sigma$ versus $\varepsilon$ post yield for all conditions such that:

$$
\hat{\sigma}_{KM} = \left( \sigma_s^2 + \left( \sigma_y^2 - \sigma_s^2 \right) \exp \left( \frac{-\Theta_0 \varepsilon}{\sigma_s} \right) \right)^{1/2}
$$

(3.13)

Both in this expression and in Eq. 3.9, $\Theta_0$ is taken to be a static value, where $\sigma_s$ and $\sigma_y$ are based on preselected phenomenological parameters $(p, q)$ for an absolute strain rate of $\dot{\varepsilon}_0$. This

![Figure 3.5: Saturation and yield stresses versus normalized activation energy $g$, corrected for obstacle profiles by $p = 1/2$ and $q = 2$.](image)

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CHAPTER 3. CONSTITUTIVE BEHAVIOUR OF AS-CAST A356

differs from other approaches taken to assessing these parameters for Al-Mg alloys, which involve reggressively solving an objective function \([101]\), and can potentially predict negative flow stresses. The other caveat of using a static value of \(\Theta_0\) is that the slight strain hardening followed by softening discernible at 300°C cannot be captured. However, this phenomena is only pronounced for a small range of thermomechanical states. Overall, this approach provides a reasonable description of the entire range of behaviour observed without resorting to scaling \(\Theta_0\).

<table>
<thead>
<tr>
<th>Model</th>
<th>Eq.</th>
<th>Coefficients</th>
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<tbody>
<tr>
<td>Extended Ludwik-Hollomon</td>
<td>3.4</td>
<td>(T_0 = 348.8^\circ\text{C}) (n_2 = 0.1892) (n_1 = -5.6919 \times 10^{-4}) (n_3 = -9.2773 \times 10^{-3})</td>
</tr>
<tr>
<td></td>
<td>3.5</td>
<td>(m_1 = 9.101 \times 10^{-11}) (m_3 = 0.0232) (m_2 = 3.4131)</td>
</tr>
<tr>
<td></td>
<td>3.6</td>
<td>(k_1 = 8.1200 \times 10^{-4}) (k_2 = -1.1570) (k_3 = 407.4)</td>
</tr>
<tr>
<td>Kocks-Mecking</td>
<td>3.10</td>
<td>(\sigma_0/\mu_0 = 0.0297) (g_0 = 1.1798)</td>
</tr>
<tr>
<td></td>
<td>3.12</td>
<td>(c_y = 0.0656) (C_y = -0.1361)</td>
</tr>
<tr>
<td></td>
<td>3.13</td>
<td>(\Theta_0 = 2850\ \text{MPa})</td>
</tr>
</tbody>
</table>

3.3 Comparison of constitutive expressions

Figure 3.2 shows examples of the measured flow stress compared to the predicted flow stresses calculated by the different constitutive expressions. In these calculations, experimental data points for strain, strain rate and temperature are directly substituted into each respective constitutive expression for flow stress. Figures 3.2a and 3.2b demonstrate the characterization of the Ludwik-Hollomon and Kocks-Mecking expressions at temperatures exhibiting strain hardening (100°C and 200°C), while Figures 3.2c and 3.2d show the comparison for the strain rate sensitive regime (450°C and 500°C). Owing to the negligible strain rate sensitivity at 100°C, a single strain rate is presented corresponding to \(R_1\) in Figure 3.2a, while the latter plots in Figure 3.2 show comparisons to the lowest and highest experimental strain rate ranges, \(R_1\) and \(R_4\) at each target temperature.

At lower temperatures, all applicable relationships describe the experimental flow stress well. The Ludwik-Hollomon fits the data best for these conditions, fitting the data for both strain rates particularly well at 200°C (Figure 3.2b). The Kocks-Mecking underestimates the low strain rate
flow stress at both 100°C and 200°C, but is acceptable for the elevated strain rate at 200°C. At elevated temperatures, the Ludwik-Hollomon and Kocks-Mecking expressions are fairly accurate at low strain rates, but overestimate the flow stress at elevated strain rates.

Calculating the sum of the mean difference as well as the RMSE between the predicted and measured flow stresses for all tests provides a measure of each models performance. Based on these metrics, the Kocks-Mecking slightly underestimates the flow stresses by less than 1% with 3% RMSE and the Ludwik-Hollomon overestimates the flow stress by 2% with 2.5% RMSE. In terms of specific strain rate and temperature ranges best described by each expression, the general trend for the strain-hardening conditions is that higher error is observed with increasing temperature, with peak error occurring in the transition regime (300-350°C). Beyond this transition temperature range, the error diminishes with increased temperatures. The accuracy of deformation models using these constitutive expressions will therefore be reflected in this behaviour.

As this material exhibits a range of temperature and strain rate dependent deformation response, selecting a sole constitutive expression to capture the full range of behaviour observed requires compromises in accuracy. The extended Ludwik-Hollomon model provides flexibility in a phenomenological approach, however requires a large number of fitting coefficients to approximate the flow stress effectively. The physically-based Kocks-Mecking is more efficient in terms of fitting coefficients, however it requires well characterized yield and saturation stresses. In order to consider strain dependence, the accuracy of the Kocks-Mecking expression at low strains is dependent on the integrated form of Eq. 3.7. The Kocks-Mecking expression in the integrated form can only describe the strain-softening behaviour of the material by scaling the initial strain hardening rate according to temperature and strain rate. This requires a large number of tests to accurately assess, which may also be a limitation when considering inherent material variations.

3.4 Summary

The constitutive behaviour of as-cast A356 has been experimentally characterized through an extensive set of compression tests. The data was used to fit both a phenomenological and physically-based constitutive expression. The material in the as-cast form shows a diverse range of thermomechanical behaviour characteristic of Al-Si-Mg alloys, with a transition from strain-hardening to strain rate dependent behaviour. Below 350°C, the material was observed to strain harden, whereas above
this temperature it becomes more strain-rate dependent with increasing temperature. This behaviour poses a challenge for developing a single constitutive expression that accurately describes all experimental results across the temperature and strain rate ranges encompassed experimentally. As a result, each constitutive expression has differing degrees of accuracy and efficiency in predicting flow stress for particular ranges of temperature, strain rate and strain. Specifically, over the experimental data tested:

- The extended Ludwik-Hollomon expression is the most flexible and therefore provides the best prediction across all temperatures and strain-rates. This expression overestimated the flow stress by an average 2% with a RMSE of 2.5%. This phenomenological approach necessitated a large number of fitting coefficients.

- The Kocks-Mecking relationship on average slightly underestimated the flow stress by 1%, but exhibited a larger RMSE. Flow stresses for elevated strain and strain rate conditions showed better agreement with the relationship. This model is physically-based, and does not have as many fitted coefficients as the Ludwik-Hollomon.

Based on this analysis, the constitutive expression that most accurately describes AC A356 across all temperatures and strain rates is the extended Ludwik-Hollomon expression.
Previous studies on the rotary forming of cast aluminum alloys (Section 1.5.1) have shown improved mechanical properties following processing, however, they did not provide insight on the underlying microstructural changes that lead to these property improvements. The effects of holding the AC material at elevated temperatures for forming purposes, followed by deformation has unknown implications on properties following heat treatment. The properties of heat treatable aluminum casting alloys are dependant on microstructural features spanning several length scales and will be affected by the thermomechanical processing schedule applied. This chapter investigates the microstructure of A356 in the AC condition, following rotary forming operations of varying intensity, and following heat treatment. This is accomplished through microstructural observations on specimens with various thermomechanical histories, as well as concurrent macro and microhardness measurements.

4.1 Microstructure and hardness

As discussed in Section 1.3.2, macrohardness measurements can be used to infer yield strength of A356. Fig. 4.1 shows selected yield strength versus hardness results reported by Tiryakioğlu et al. [38] for underaged A356 with low and high levels of Mg, converted to $H_V$. The original data was reported as $H_{RF}$ (Fig. 1.6), and was converted according to the method given in Appendix B. A non-linear least-squares fit of the data provided by Tiryakioğlu et al. shows that $\sigma_y = f(H_V)$ can be described by a power-law relationship, as suggested by Colley [18]. Comparing data provided by Colley for A356 (with DAS equal to 30 µm) in both the over and underaged conditions shows good agreement, particularly at lower hardness values. The overall goodness of fit of the power-law relationship is greater than 0.95, and the RMSE is approximately 14 MPa over all experimental data.

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1A version of this chapter is intended for publication in Materials Science and Engineering: A
CHAPTER 4. CHARACTERIZATION OF ROTARY FORMED MATERIAL

Figure 4.1: $\sigma_y$ to $H_V$ relationship based on converted measurements made by Tiryakioğlu et al. [38] on underaged Al-7%wtSi-0.2%wtMg and Al-7%wtSi-0.4%wtMg. Data from Colley [18] for over and underaged Al-7%wtSi-0.3%wtMg plotted for comparison.

This indicates that this data is sufficient for inferring local yield strength directly from hardness values for the underaged condition.

Using the EFA developed at UBC, three workpieces were rotary formed as described in Section 2.2.6 to varying levels of deformation. Taken from a combination of undeformed and rotary formed workpieces, 9 sections were analyzed using the hardness profiling methodology described in Section 2.3.4. Between 950-1350 hardness measurements were performed on each section. Axial and circumferential sections, extracted from a blank, as well as axial sections, extracted from as-deformed workpieces, were analyzed prior to heat treatment. Axial sections from a blank and the workpieces were also analyzed following a T6 heat treatment.

4.1.1 Experimentally formed material

As-cast and as-deformed material

Fig. 4.2 shows the results of hardness measurements performed on axial and circumferential ($72^\circ$) profiles taken from the locations indicated in Fig. 2.3. Also shown on the axial section are 11 equidistant segment markers to assist in tracking changes through processing. Hardness measurements performed on the two profiles showed a similar range of hardness. Axially, the highest hardness was found to be at either end of the sample, with the centre being the softest. Circumferentially, a gradient in hardness was observed. The variations in hardness in both directions is thought to be
 CHAPTER 4. CHARACTERIZATION OF ROTARY FORMED MATERIAL

**Figure 4.2:** Axial and circumferential hardness profiles of the AC blank. Circumferential section represents a symmetric portion of the wheel from which the blank was machined from.

Due to differences in DAS and eutectic phase fractions caused by variations in solidification time and the transport of Si-enriched liquid during solidification.

The hardness profiles of each of the as-deformed sections, shown in Fig. 4.3, demonstrate a large change from the AC condition. In all cases, the mean hardness has dropped significantly. The point of initial roller contact in each specimen and direction of roller travel has been identified with an arrow in Fig. 4.3. In the undeformed regions, the hardness distributions show similar trends to the AC condition albeit with a reduction in average hardness. Particularly evident in the highly deformed specimen is a region of elevated hardness appearing immediate to the initial forming site. Elevated hardness is also observed near the tip of the specimens (segments 10 and 11) in the deformed regions similar to the AC condition. The remaining positions in the deformed regions show increased hardness values relative to the peak hardness when compared with the AC condition. Additionally, the increased hardness values in these areas (segments 8 and 9, 6–9, and 4–9 on the deformed cross-sections, respectively), are higher at locations closer to the outer diameter. This
CHAPTER 4. CHARACTERIZATION OF ROTARY FORMED MATERIAL

Figure 4.3: Comparative hardness profiles of as-deformed axial sections. Arrows indicate start of forming.

A type of hardness distribution has also been seen in the rotary forming of steel [68].

In order to track the effects on DAS, five optical micrographs, or fields, were selected at random along each of 9 blank/workpiece sections at a depth between 2-3 mm from the outer diameter. Approximately 300 discrete measurements were performed across each of the 5 fields per segment, according to the methodology discussed in Section 2.3.2. The results of this analysis are shown in Fig. 4.4.

In comparing the results of the AC and deformed material, it appears as though forming had little impact on the mean DAS. Comparing measurements in the most heavily deformed region
Chapter 4. Characterization of Rotary Formed Material

Figure 4.4: Mean DAS measurements of undeformed blank and deformed workpieces. Arrows indicate start of deformed section for each specimen. Error bars indicate ±1 standard deviation.

of each specimen (corresponding to segment 10 and 11) shows a slight decrease in the mean DAS with deformation. Only the highly deformed material showed a consistent decrease in the mean DAS measurement in all segments affected by forming over the undeformed (AC) material. However, this trend is not conclusive. A comparison of the mean DAS in the undeformed regions (segments 1 and 2) displays approximately the same variance. Generally speaking, it is clear that DAS increases with cross-section thickness in both the AC and deformed material. As larger DAS typically corresponds to lower yield strength and a correspondingly lower hardness value, it is postulated that the cause for elevated hardness in sections 1–3 for all specimens may be related to the presence of elevated levels of eutectic phases.

To evaluate the link between high hardness and the presence of increased fractions of eutectic, the indentation fields from the axial section of the least-deformed workpiece were reprocessed to quantify the distribution of eutectic. This specimen was the most uniformly polished and therefore had fields which were best suited for microstructure evaluation of all specimens. The reprocessing commenced with extracting a subfield absent of the indent. This was followed by determining the greyscale level for each subfield image separating the eutectic and $\alpha$-Al phase. This was accomplished by differentiating the peak counts of pixels to identify an inflection greyscale value. An example is shown in Fig. 4.5. The distributions of hardness (Fig. 4.3a) and eutectic fraction (Fig.
4.5d) are quite similar. As such, the condition of the eutectic is a key factor in determining the resulting hardness of this alloy.

**Heat treated material**

As demonstrated in Fig. 4.6, the range of hardness values measured in each sample after conducting the T6 heat treatment show an overall increase in hardness compared to the AC condition and a dramatic change in the hardness distribution. A comparison of the hardness in the deformed regions of each section following the T6 heat treatment indicates decreased hardness with increased deformation. The elevated hardness region found at the end of the undeformed specimen, approximately 115 kg/mm$^2$, progressively decreases with deformation to approximately 100 kg/mm$^2$ in the deformed specimens at the same location. Additionally, the regions of elevated hardness appearing immediately beyond the start of forming as seen in the as-deformed specimens appears to be eliminated. This is particularly evident in the peak-deformed workpiece (Fig. 4.6d vs. 4.3c). Regions in

![Figure 4.5: Eutectic fraction ($\phi_{Eu}$) distribution in the least deformed workpiece (Fig. 2.14d) in the as-deformed condition.](image)
deformed specimens (e.g. segment 1) that were outside the deformation zone exhibit a fractionally higher hardness (approximately 5 kg/mm$^2$) than similar locations in the undeformed specimens. This small increase in hardness may be due to increased fragmentation caused by longer coarsening times resulting from heating and holding at forming temperatures. According to the $H_V$-$\sigma_y$ relationship presented earlier (Fig. 4.1), the 15 kg/mm$^2$ difference in hardness between undeformed and heavily deformed material (segment 11) represents a 60.6 MPa drop in $\sigma_y$.

This indicates that regions that saw mechanical and thermal processing were softened by the process, while regions which were only processed thermally saw increased hardness.

### 4.1.2 Commercially formed material

The commercially formed material was analyzed using the same procedures that were applied to the experimentally formed material. Specimens for this analysis were extracted from an AC blank and from a commercially flow-formed wheel following a T6 heat treatment. The hardness profiles of the AC and deformed-T6 conditions are shown in Fig. 4.7. The DAS was measured at select segments roughly corresponding to where the dendritic structure could be identified. The measured DAS for both sections is summarized in Fig. 4.8.

The hardness distribution and DAS profile of the AC blank for the commercial flowing process show similar ranges of values as the AC blank for the EFA in the regions corresponding to segments 1–4. However, outside these positions, significantly higher hardness values are observed, particularly near the end or tip of the blank (segment 12). In comparing this result to that of the DAS measurements, a tentative explanation for this change is due to locally elevated levels of eutectic owed to the coarse microstructure in segments 6–12. This latter point is speculative and based on the comparison made with the least-deformed experimental specimen; while the indentation fields permitted accurate hardness and DAS measurements for the commercial specimens, they were not suitable for eutectic evaluation.

The effect of forming on the microstructure is readily apparent from the decreased DAS in the deformed regions. DAS measurements on the deformed-T6 sample between segments C and D could not be performed because the high level of deformation had broken up the dendritic structure resulting in the microstructure shown in Fig. 2.7b and 2.7c. Unfortunately, the non-uniform deformation profile coupled with rough machining prior to heat treatment precludes a quantitative positional
Figure 4.6: Comparative hardness profiles of deformed-T6 axial sections. Arrows indicate forming start point.
CHAPTER 4. CHARACTERIZATION OF ROTARY FORMED MATERIAL

Figure 4.7: Commercially formed material in the AC condition (a) and deformed-T6 in (b). Dashed lines indicate DAS measurement locations (Fig. 4.8).

Figure 4.8: Mean DAS measurements of undeformed and commercially formed material. Arrow indicates forming initiation point. Error bars indicate ± 1 standard deviation.
4.2 Effects of processing on microstructure

In order to ascertain the effects of holding the AC material at an elevated temperature before forming, coupons (location and size given on Fig. 2.3) were extracted from an AC blank and held at elevated temperatures in a nitrate salt bath (described in Section 2.3.1) for varying lengths of time. Samples were left to air cool upon removal from the salt bath, reflecting the EFA procedure. Hardness measurements were made on each sample before the treatment and within 30 minutes of cooling to ambient temperature. The microstructure of select samples was also assessed using optical and electron microscopy. This work was aimed at determining the effects of holding the material at an elevated temperature (or ageing) prior to forming.

The temperatures selected for this investigation were based on potential forming temperatures and include: 300, 350, and 400°C, as well as the solutionizing temperature of 540°C. Selected to span the potential breadth of forming operations, target hold times were 2, 10, 20 and 50 minutes. The temperature history of each sample was monitored with a thermocouple. The approximate time to cool to 100°C for all specimens was 3.5 minutes. Air cooling as opposed to water quenching was selected as it best matched the procedure employed with the EFA.

4.2.1 Hardness observations

The average hardness and standard deviation for each hold temperature are plotted as a function of hold time in Fig. 4.9. For all temperatures below 540°C, there is a clear power law drop in hardness versus hold time, with better agreement at 350 and 400°C. Furthermore, as temperature increases, the standard deviation in the hardness measurements diminishes. This is likely due to the additional time at elevated temperature encountered during air-cooling for specimens held for shorter periods of time. At these temperatures, diffusion mechanisms are active and both Si and Mg have increased solubility in the matrix. Apelian et al. [20] reported that the equilibrium solubility of Mg and Si in
solid aluminum when both Mg$_2$Si and Si are present as:

$$\%\text{-wt Si} = 1.69 \times 10^{-9} \cdot T^{3.19}$$  \hspace{1cm} (4.1) \\
$$\%\text{-wt Mg} = 1.45 \times 10^{-9} \cdot T^{3.16}$$  \hspace{1cm} (4.2) \\

for $310^\circ C < T < 575^\circ C$. This implies that the solubility of Mg and Si increase by approximately 2.5 times when the temperature is increased from 300 to 400$^\circ C$. Thus, the potential evolution of the Si-rich eutectic phase and Mg-bearing structures precipitated during casting is enhanced, effecting the following phenomena:

- Coarsening of Mg$_2$Si precipitates that formed during initial casting;
- Complete or partial dissolution of small Mg$_2$Si precipitates;
- Eutectic spheroidization beyond initial fragmentation; and
- Eutectic coarsening beyond spheroidization.

The initial casting procedure produces relatively coarse and unevenly distributed Mg$_2$Si precipitates. Coarsening of these precipitates may occur as Mg and Si transport via diffusion is enhanced with increasing temperatures or is allowed to occur through longer hold times. This primarily affects Mg$_2$Si, but also intermetallics to a lesser extent owed to their stability. These microstructural changes significantly decrease the overall mean hardness as precipitation strengthening diminishes.

Simultaneously, the AC eutectic structure is also affected. The potential for eutectic Si-phase fragmentation as well as spheroidization is more likely with increasing hold temperatures. The former is driven in small part by thermal stress arising from the property mismatch of brittle Si in the Al matrix [119], and more predominantly by diffusion [120]. Spheroidization occurs solely by diffusion of Si. The changes to eutectic morphology are expected to accelerate with increased holding temperature. The collective change in hardness due to changes in precipitation and eutectic structure can be expressed as a function of time $t$ and temperature $T$:

$$\Delta H_V = \left(-2.93 \times 10^{-2}T + 4.21\right) \log t$$  \hspace{1cm} (4.3)
While this expression is only valid through the temperatures demonstrating power law behaviour, an extrapolation of this relationship implies that there is no thermal effect on the microstructure \((\Delta H_v = 0)\) below 144°C. The flow stresses presented in Fig. 3.1c show a much higher drop in work hardening rate between tests conducted at \(\sim 120°C\) and \(\sim 190°C\) as compared to the drop from \(\sim 45°C\) to \(\sim 120°C\). Since diffusive transport is expected to be low at these temperatures, this implies that the extrapolated temperature may be when fragmentation of the eutectic commences.

In the case of the 540°C results, there is no power law drop in hardness versus time observed as with the other temperatures. Consistent with samples tested at other temperatures, there was a large initial drop observed in the specimen held for 2 minutes. However, the hardness increases from this point on. While the eutectic-Si morphology is expected to change as in the lower temperature conditions, the effects of precipitation are superimposed. With longer temperature holding times, there is a progressive increase in precipitate dissolution, leading to higher levels in solution. The low air cooling rate and the potential for natural ageing results in increased hardness with time, coincident with increased levels of dissolution attained at temperature.

### 4.2.2 Microstructural observations

To examine the effects of hold temperature on the microstructure, the coupons held at each temperature for 50 minutes were analyzed via optical microscopy and EDX. These specimens were also
deep etched via the methodology given in Section 2.3.3. The results of this analysis, presented in Fig. 4.10, show how the Si-eutectic structure evolves with increased holding temperature. Subtle modification of the eutectic structure is evident from the optical microscopy, while the SEM images following deep etching show coarser features with increasing hold temperature. Holding at 300°C, some of the larger eutectic Si branches have rounded and are joined by less refined fiber morphology. Increasing the temperature to 400°C shows a continued evolution of this morphology resulting in fewer, thicker branches being observed. At the solutionizing temperature of 540°C, the particles are fully fragmented and spheroidization is evident. The EDX results show that localized Mg-bearing structures are present up to 400°C. These are expected to be predominantly Mg$_2$Si (outlined in red/orange in Fig. 4.10-4.11); however, intermetallics may also be present. Once the solutionizing temperature is reached, there is no evidence of these localized Mg-bearing structures. While the EDX observations do not show the evolution in distribution of Mg at 300 and 400°C from the AC state, the absence of regions containing concentrated Mg in specimens held at 540°C is congruent with the observations made regarding the macrohardness results (Fig. 4.9).

A sample of undeformed material following the complete T6 heat treatment was also analyzed using this methodology. The results of this analysis are also presented in Fig. 4.10. The distribution of Mg in this sample is approximately the same compared to the sample held at 540°C for 50 minutes. The eutectic-Si in the T6 sample has also spheroidized to a greater extent and some coarsening has occurred as characterized by larger and fewer particles with the same field size.

This methodology was also applied to analyze deformed material before and after a T6 heat treatment. Specimens were extracted from a location approximately 1 mm from the roller interface in the sample that experienced the largest deformation in the EFA. The axial location of the specimens coincided with the undeformed specimens employed to evaluate the effect of hold temperature/time. The resulting micrographs for this material are shown in Fig. 4.11. Prior to heat treatment, the eutectic-Si particle size appears to have decreased compared to the undeformed specimens and has been compacted in line with the deformation. There is less evidence of spheroidization having occurred, as the Si morphology is observed to be small, short fibers/plates. The EDX maps suggest that the Mg-bearing structures have consolidated on the edges of the dendrite arms, appearing as plates oriented parallel to the forming direction. The morphology and distribution of these structures explains the hardness profiles seen in the spun material prior to heat treatment (Fig. 4.3c), where
Figure 4.10: Eutectic particle morphology of undeformed specimens held at various temperatures and times, provided by optical images of the microstructure, element maps generated via EDX and SEM images of eutectic particle morphologies following deep etching. This was conducted on specimens in the AC condition (a), held for 50 minutes at 300 (b), 400 (c) and 540°C (d) and the T6 condition (e).
4.2.3 Eutectic particle shape and size

The Lifshitz, Slyozov and Wagner (LSW) coarsening model \([121, 122]\) provides a means of quantifying eutectic particle size evolution with time:

\[
k_{\text{LSW}} = \frac{\bar{d}^3 - \bar{d}_0^3}{t}
\]  

(4.4)

While no data in the present study was collected regarding initial particle diameters, the model demonstrates how eutectic particles are dependent on a temperature dependent constant and time. Competitive coarsening driven by diffusion, as described by the LSW model, predicts a steady-state lognormal distribution of particle sizes, centred about \(\bar{d}\), once fragmentation is complete \([123]\). The

Figure 4.11: Eutectic particle morphology of as-deformed and deformed-T6 material, provided by optical images of the eutectic microstructure, element maps generated via EDX and SEM images of eutectic particle morphologies. This was conducted following deep etching specimens in the as-deformed (a) and deformed-T6 condition (b). Arrow indicates forming direction.

regions of elevated macrohardness were found coinciding with deformed regions. In the deformed-T6 condition, localized Mg is absent as in the case of the undeformed material, however the eutectic structure differs. While spheroidized, the eutectic particles are found to be appreciably smaller in count and size for equivalent field sizes than those observed in the undeformed material.
presence of these hard Si particles within the softer Al matrix results in a strengthening effect due to the eutectic phase consistent with Metal Matrix Composite (MMC) theory [40,124].

In order to quantify the effects of different processing paths on the eutectic particles, particle analysis using optical microscopy was conducted on specimens of AC material solutionized at 540°C for 50 minutes, undeformed-T6 material, peak deformed-T6 material and commercially formed-T6 material. The particle characteristics were quantified with ECD and aspect ratios measured from best-fit ellipses. The measurements were then fit to a log-normal Probability Density Function (PDF) according to:

\[ P = \frac{1}{x \sqrt{2\pi s^2}} \exp \left( - \frac{(\ln x - m)^2}{2s^2} \right) \]  

(4.5)

where \( x \) is ECD or aspect ratio, \( m \) and \( s \) are the mean and standard deviation of the natural logarithm of \( x \). The resulting statistics in terms of arithmetic mean, mode, \( m \) and \( s^2 \) are summarized in Table 4.1 and Fig. 4.12. This analysis indicates that the aspect ratio does not vary significantly between the different processing paths. As the melt was chemically identical for all specimens, a possible explanation for this is that the modification technique produces a narrow range of aspect ratios after fragmentation. Wang [28] showed that the distribution of aspect ratio in modified A356 and A357 was nearly identical and otherwise identical unmodified material showed a distinct difference. A clear difference was noted in the ECD for each sample, with the commercially deformed material having the smallest ECD, followed by the solutionized material, then EFA-T6 and finally the undeformed material having the largest particle size. The ECD and aspect ratio measurements of the undeformed material are comparable those of Wang et al. [28] for modified A356-T6, and the ECD measurements are approximately half of those found for unmodified A357-T6 [28,38].

The results for the solutionized material and the undeformed material in the T6 condition are consistent with phenomena described in Section 1.2. As reflected in the aspect ratio and ECD measurements, the solutionized material did not coarsen to the same extent as the T6 specimen owing to the longer time at temperature for the latter material. Assuming the \( k_{LSW} \) coefficient (Eq. 4.4) is the same for both deformed and undeformed material and taking \( \bar{d} \) as the mode value, these results indicate that deformation fragments the eutectic-Si to a much greater extent, leading to smaller eutectic particle sizes after heat treatment. Increased levels of deformation advance the degree of fragmentation, which explains the size difference between the EFA and commercially formed mate-
Table 4.1: Eutectic particle statistics.

<table>
<thead>
<tr>
<th>Material</th>
<th>Statistic</th>
<th>Aspect ratio</th>
<th>ECD (µm)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Mean</td>
<td>1.60</td>
<td>1.81</td>
</tr>
<tr>
<td></td>
<td>Mode</td>
<td>1.40</td>
<td>1.42</td>
</tr>
<tr>
<td>Solutionized (Fig. 4.10d)</td>
<td>m</td>
<td>0.421</td>
<td>0.511</td>
</tr>
<tr>
<td></td>
<td>s²</td>
<td>8.20 x 10⁻²</td>
<td>0.162</td>
</tr>
<tr>
<td>T6 (Fig. 4.10e)</td>
<td>Mean</td>
<td>1.51</td>
<td>2.28</td>
</tr>
<tr>
<td></td>
<td>Mode</td>
<td>1.35</td>
<td>1.70</td>
</tr>
<tr>
<td></td>
<td>m</td>
<td>0.372</td>
<td>0.728</td>
</tr>
<tr>
<td></td>
<td>s²</td>
<td>7.43 x 10⁻²</td>
<td>0.195</td>
</tr>
<tr>
<td>EFA-T6 (Fig. 4.11b)</td>
<td>Mean</td>
<td>1.57</td>
<td>2.01</td>
</tr>
<tr>
<td></td>
<td>Mode</td>
<td>1.38</td>
<td>1.51</td>
</tr>
<tr>
<td></td>
<td>m</td>
<td>0.403</td>
<td>0.600</td>
</tr>
<tr>
<td></td>
<td>s²</td>
<td>7.89 x 10⁻²</td>
<td>0.189</td>
</tr>
<tr>
<td>Com. T6 (Fig. 2.7c)</td>
<td>Mean</td>
<td>1.51</td>
<td>1.63</td>
</tr>
<tr>
<td></td>
<td>Mode</td>
<td>1.36</td>
<td>1.26</td>
</tr>
<tr>
<td></td>
<td>m</td>
<td>0.373</td>
<td>0.398</td>
</tr>
<tr>
<td></td>
<td>s²</td>
<td>6.62 x 10⁻²</td>
<td>0.164</td>
</tr>
</tbody>
</table>

Figure 4.12: PDFs of eutectic particle ECD and aspect ratios.

4.3 Phase-specific effects of processing

In an attempt to ascertain the degree to which processing history effects the primary and eutectic phases, microhardness tests with a low load were employed to selectively test each phase of material. This was conducted on undeformed material with various thermal histories, and the peak deformed material processed by the EFA. Commercially formed material was excluded from this analysis due to the inability to accurately isolate and test each phase. The results of these measurements, presented in Fig. 4.13, show the relative contribution of each phase to the macro hardness and overall strength. The mean microhardness and standard deviation of 30 individual measurements is given for the breadth of conditions presented in the previous section. Indentation locations were chosen such that the plastically affected zone was retained within each phase, as shown in Fig. 4.13b and 4.13c.

In the AC condition, the eutectic shows a significantly higher hardness as compared to the primary α-Al phase. The drop in the hardness of the eutectic in samples held at 300 through to 400°C for 50 minutes is nearly identical, surmised to be mostly attributed to eutectic fragmentation and
some Mg dissolution which increases precipitate coarsening. The decrease in hardness of the primary is identical for specimens in the AC conditions to those for hold temperatures of 300 through to 350°C, and decreases further to a minimum at 400°C, which can be entirely attributed Mg dissolution.

Below holding temperatures of 540°C, it appears that the hardness in the eutectic stabilizes after an initial drop, while the peak primary α-Al phase hardness decreases consistently with temperature. This suggests that the time-dependant decrease in macrohardness (Fig. 4.9) is driven by Mg diffusion beyond the initial effects of eutectic fragmentation. In the 540°C case, both the eutectic and primary phase show increased hardness relative to the other specimens held at lower temperatures. Both benefit from Mg₂Si precipitation expected due to slow cooling and eutectic spheroidization. The hardness of the eutectic in the deformed material without heat treatment is approximately the same as the the specimens held at temperatures below 540°C, and the primary α-Al phase is some-

![Figure 4.13: Comparative microhardness of the α-Al phase and eutectic in various conditions (a). Micrographs demonstrate an example of an indent made in α-Al in the AC condition (b) and one made in a eutectic region in the T6 condition (c). Dashed circles representing the estimated plastic zone in (b-c) have diameters 2.5× the indent diagonals.](image)
where between the specimens held at 350 and 400°C. This indicates that prior to heat treatment, the majority of the modification in strength can be attributed to changes in microstructure due to thermal effects.

Following T6 heat treatment, both the eutectic and primary hardness increase appreciably, with more effective precipitation and spheroidization. The hardness of the primary $\alpha$-Al phase in the deformed material is approximately the same as the undeformed material, having a mean hardness within a standard deviation of the undeformed. The mean hardness of the deformed material’s eutectic phase is 28% less than that of the undeformed, which indicates that the principal cause of the hardness decrease observed in deformed samples in the T6 condition (Fig. 4.6) is due to changes localized to the eutectic.

Overall, these phase-selective microhardness measurements indicate that the primary $\alpha$-Al phase is slightly softer in deformed material, but not appreciably different than that of undeformed material. This indicates that precipitation strengthening is not greatly affected by deformation. The same measurement has type attributed the drop in macrohardness to a significantly softer eutectic phase. It is evident that the condition of the eutectic is a better indicator of hardness, and by virtue, strength of this material as opposed to DAS. This not only includes the shape and size of particle beyond heat treatment at the micro scale, but also the overall phase fraction on the macro scale.

### 4.4 Surface defects

As demonstrated in Fig. 4.14 and 4.15, the mid-level and peak deformed EFA workpieces showed surface defects in the form of cracking on the outer diameter in the deformed regions. Cracking did not manifest on the least-deformed sample, but was most prominent on the mid-level deformed sample where the highest number of cracks throughout the deformed region were observed. Cracks appearing early in the forming pass alternate between opening counter to the forming direction (A, C) to predominately opening with the forming direction (B, D–U). The cracks at this stage of deformation do not extend any further than 140 $\mu$m into the bulk of the material, measured radially, or normal to the forming direction. With the exception of one crack (C) extending up to 320 $\mu$m in length along the axis of the workpiece, other cracks are shorter and do not not exceed 200 $\mu$m. The defects observed were larger in scale than those reported by Mori et al. [77], who reported a maximum size of 60 $\mu$m. Similar to Mori et al., however, cracks were found to predominantly
penetrate the surface of the sample with an angle of approximately $45^\circ$ from the radial direction, occurring in eutectic-rich regions.

The peak deformed sample embodied two discrete forming passes, with the first pass characteristic of the mid-level deformed specimen. This sample also exhibited surface cracks, albeit with fewer, larger cracks than characterized by the mid-deformed specimen, as shown in Fig. 4.15. It appears that minor cracks created in earlier passes were closed, while larger cracks are deformed along with the bulk material and are elongated as a result. None of the cracks observed extended any further than 100 $\mu$m into the bulk of the material, however the characteristic axial lengths (a maximum of 1.2 mm) were much longer than less deformed material. The penetration angle has also
changed to approach the forming direction, however, cracks were still observed to follow eutectic-rich regions as was seen with the mid-deformed workpiece.

This type of localized failure is characteristic to the forming process, as it has been characterized in other forward rotary forming operations [77, 82]. Cracking or ‘fish scaling’ in rotary forming is caused by highly localized shear occurring both ahead of and behind the roller interface. The extent of cracking is dependent on processing parameters and the local strength of material. Both of these
factors decide the overall crack morphology. Thus, the brittle eutectic phase appears to be a weak point in A356 where cracks initiate.

To eliminate cracking, either the forming temperature may be increased, or the forming parameters altered to avoid high levels of radial shear. Increasing the forming temperature from 350 to 400°C was confirmed to arrest the formation of cracks by Mori et al. [77] for similar forming parameters. This is also inline with the torsion testing results of McQueen et al. [40], who reported significant increases in strain to fracture moving from 300°C to 400°C, particularly at lower strain rates. Thus, in addition to temperature, forming speed may be one of the process parameters that may be changed to reduce the frequency and severity of surface defects. The effects of further parameter changes are discussed in the following chapter.

4.5 Summary

Rotary forming of A356 at elevated temperatures has shown that the microstructure is affected by a number of different factors across several length scales. Combined hardness profile and microstructural analysis shows that the DAS has less of an effect on hardness than the distribution and condition of eutectic-Si phase. Heating the AC material prior to deformation initiates diffusion-driven coarsening of precipitates and modifies the eutectic structure. An extrapolation of the data from targeted static thermal experiments suggest that the AC material is stable up to approximately 144°C. Prior to heat treatment, rotary formed material exhibits a decreased macrohardness in-line with the time spent at elevated temperature, indicating that the decrease in hardness between the AC undeformed state to the as-deformed state is principally a thermal effect. After heat treatment, there was a small macrohardness increase observed in the regions unaffected by forming in the EFA processed material as compared to unprocessed material with the same heat treatment. This coincided with a large decrease in macrohardness in heavily deformed regions over unprocessed material due to changes in eutectic particle size.

Eutectic particle size and shape analysis showed that rotary forming fragments the eutectic structure, with particle sizes fragmenting to a greater extent with higher levels of deformation. Furthermore, increased levels of deformation create smaller eutectic particles after heat treatment, which was correlated to lower macrohardness. As a result, it is surmised that flow formed material in the T6 condition may exhibit decreased yield strength as compared to undeformed material in the
same state, despite smaller eutectic particles observed in the deformed material.

Eutectic regions were also observed to coincide with surface defects in the form of cracks found in EFA processed material. These cracks occurred extensively along the length of the mid-deformed part, originating and propagating through eutectic-rich regions. Further deformation of the cracked surface seemed to close some smaller cracks, while elongating others. The cause of these surface defects is a combination of forming temperature (affecting the local strength of the material) and processing parameters such as forming speed.
CHAPTER 5

MATHEMATICAL MODELLING OF ROTARY FORMING

As discussed in Section 1.5.1, rotary forming imparts significant levels of localized workpiece deformation. The degree of deformation is dependent primarily on tooling and feed-rates, and workpiece geometry as well as deformation history; often the roller(s) will pass over the same workpiece location several times to incrementally achieve bulk deformation. This poses several challenges to develop a model capable of accurately predicting the strain path and overall workpiece deformation. Chief among these challenges is considering the effects of temperature and strain rate on the constitutive behaviour, which has not been previously considered. The following sections describe how these challenges were addressed in developing and applying a coupled thermomechanical model of the EFA process detailed in Section 2.2.6.

Specifically, the overall model reflecting the EFA process included submodels for workpiece preheating, forming at elevated temperatures, and cooling once forming was complete. For the preheating submodel, an axisymmetric domain assumption was made for both the mandrel and the blank. Thermal and mechanical boundary conditions were imposed in order to reflect the geometry and temperature of the workpiece prior to forming. The results from this model were used to generate a 3D meshed description of the workpiece. Forming of the workpiece was then modelled employing an adiabatic assumption using rigid analytical tooling to describe the mandrel and roller. The assumption of adiabatic conditions was applied to the workpiece to capture the effects of heat generated due to plastic deformation. The use of rigid analytical surfaces to describe the tooling precludes calculating their temperature history. Once the forming step was complete, cooling of the workpiece was simulated to permit direct comparison between experimentally achieved and

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Further material is intended for publication in the Journal of Materials Processing Technology.
simulated geometries at ambient temperatures.

5.1 Coupled thermomechanical EFA model development

In order to calculate the evolution of stress, strain and temperature within the workpiece during rotary forming, a fully coupled thermomechanical model was developed using the commercial FEA software package, ABAQUS. With this technique, both the thermal and mechanical states are solved together and are permitted to influence one another via the boundary conditions imposed. A coupled thermomechanical model is needed to describe the thermal effects on the mechanical response and vice versa. Changes in the thermal state have mechanical effects through phenomena such as thermal expansion, and changes in material properties such as diminished yield strength and plasticity. Mechanically, heat is generated by the conversion of plastic strain energy, in turn affecting the thermal state. Coupling the states allows the model to include the interdependent effects in the solution. As rotary forming involves high levels of localized plasticity, a coupled model provides the most accurate description of changes to the workpiece during this process.

ABAQUS provides both implicit and explicit FEA tools that are capable of performing this type of analysis. Implicit finite element methods are well suited to modelling quasi-static thermomechanical processes such as casting because of the long durations and gradual changes in boundary conditions. However, dynamic processes with discontinuities in contact and large plastic deformation are better suited to an explicit approach. Explicit approaches rely on a direct calculation of dependent variables over a given time increment, whereas the implicit approach solves for dependent variables expressed in terms of coupled equations. Both involve numerical time integration to solve for the unknown workpiece displacements and temperatures, which is the basis for the resulting strains and stresses. A detailed comparison between the two solution techniques in terms of a thermomechanical framework has been outlined by Koric et al. [125].

Both implicit and explicit techniques were employed in modelling the rotary forming conditions realized with the EFA. The initial preheating of the workpiece was predicted using an implicit, 2D-axisymmetric model. In addition to predicting the temperature, this model was employed to determine the geometry of the mandrel and workpiece following thermal expansion. The results of this 2D model were then used as initial conditions for an explicit 3D forming model.
5.1.1 Geometry

There are three main components needed to accurately describe the rotary forming process of the EFA: the workpiece, mandrel, and roller. During preheating, only the workpiece and mandrel are involved. Due to the axial symmetry of the mandrel and workpiece during preheating, a 2D-axisymmetric representation was employed. This use of symmetry reduces computational overhead. As discussed in Section 1.5.2, the deformation conditions occurring during rotary forming do not exhibit a plane of symmetry and thus require a 3D representation.

The roller geometry was sized according to that employed with the EFA, and was assumed to retain ambient dimensions. The mandrel, which heats up with the workpiece, was sized based on temperature corrected geometry as described in the following section. The mandrel and roller were defined as rigid analytical surfaces in order to diminish computational overhead. These surfaces are unmeshed, do not deform, and do not permit heat transfer. Boundary conditions or constraints to degrees of freedom of these surfaces are applied via a single reference node for each instance, i.e. one reference node for the roller, and one for the mandrel. Using rigid analytical surfaces to describe the tooling precludes the possibility of deformation, which is a valid assumption based on the difference in flow stress between tool steel and A356 at forming temperatures.

The explicit technique employed for forming simulations requires a mesh which is as uniform as possible. The maximum time increment achieved in explicit FEA is proportional to the shortest path across any element. Variability in element edge length can needlessly increase the number of time increments needed to complete a simulation. For irregular geometries, it is not possible to generate a perfectly uniform mesh and therefore the element edge lengths will vary between a minimum, \( L_{\text{min}} \), and maximum, \( L_{\text{max}} \). Meshing strategies for explicit models seek to minimize the difference between the two lengths. For the preheating model, a 2D section of the workpiece was decomposed into quadrilateral elements with a minimum element edge length, \( L_{\text{min}} \), of 2 mm. The resulting mesh had 573 nodes and 487 2D-axisymmetric, quadrilateral elements (Fig. 5.1). Five nodes on this 2D section were designated as tracking nodes to assist in later assigning boundary conditions. An axial section of the mandrel was meshed with slightly coarser, elements having a \( L_{\text{min}} \) close to 4 mm, resulting in 673 nodes and 515 elements. Four tracking nodes were designated on the surface of the mandrel to track thermal expansion, in addition to the five on the outer surface.
of the workpiece.

The elements for the axisymmetric model featured hourglass control and reduced integration (CAX4RT). While these element options were potentially unnecessary for the preheating simulation, they were necessary for later translation of the workpiece mesh from 2D to 3D. In displacement-based FEA, reducing the number of integration points is a standard technique to reduce computational costs, and is often necessary to enable the solution of problems involving large plastic deformation. Especially when employing reduced integration, hourglass control is necessary to prevent elements from ‘locking’. This phenomena occurs when boundary conditions cause elements to develop zero stiffness, which may occur depending on deformation and the number of integration points [126]. A more detailed stability analysis of explicit finite elements has been conducted by Ling and Cherukuri [127].

To generate the mesh for the forming model, the 2D-axisymmetric mesh was revolved about its axis of symmetry to form the 3D mesh. In order to minimize the element count, the full 360° section of the workpiece was discretized into 297 circumferential segments, corresponding to the $L_{\text{max}}$ of the 2D mesh (Fig. 5.2). This resulted in 170181 nodes and 144639 brick/hexagonal elements (C3D8RT), which inherited the reduced integration and hourglass controls from the 2D-axisymmetric elements. This mesh density was selected primarily on the basis of minimizing the computational resources, and the effects of this density will discussed further in Section 5.3.

5.1.2 Material properties

A coupled thermomechanical model requires the definition of both thermal-physical and mechanical properties for each of the different materials employed. Thermal-physical properties include density,
Figure 5.2: Workpiece mesh employed for forming simulations. A detailed view of the 2D mesh employed for preheating simulations is shown in (a) with an inset showing \( L_{\text{max}} \). The revolved 3D mesh is shown in (a), constructed from the 2D mesh shown in (a). A detail view of this 3D mesh is given in (c).
specific heat capacity and thermal conductivity. Mechanical properties include modulus of elasticity, Poisson’s ratio, the coefficient of thermal expansion and a description of the flow stress if plasticity is to be considered. The full set of property data was implemented in both the preheating and forming simulations for the A356 workpiece. Plasticity was considered in the preheating simulation to ascertain appropriate boundary conditions. The material definition of the AISI-4320 mandrel was only necessary for the preheating simulation, and did not require a description of plasticity. This is because the meshed instance of the mandrel was replaced by an analytical surface in the forming simulation, and all deformation encountered during the course of preheating was elastic. Material properties for both materials was assumed to be isotropic.

Most of the FEA of incremental forming in the literature has employed isothermal, quasi-static material properties as outlined in Section 1.5.2. However, as demonstrated in Chapter 3, the flow stress behaviour of A356 is temperature and strain rate dependent. Thus, in the current work, temperature dependent material properties, based primarily on literature reported values (with the exception of the flow stress relationship), were implemented with the goal of quantifying the mechanical response at elevated temperatures and strain rates.

The extended Ludwik-Holloman behaviour (Eq. 3.3) discussed in Chapter 3 was implemented in the model for A356 via the user subroutine UHARD for the implicit solution (ABAQUS Standard) and VHARD for explicit cases (ABAQUS Explicit). These subroutines calculate the local flow stress based on the temperature, strain and strain rate at each respective integration point. The temperature-corrected elastic modulus and Poisson’s ratio were implemented based on literature values [98], given in Eq. 3.2.

The thermal-physical properties of A356 were taken from Hétu et al. [128]. Thermal conductivity and specific heat capacity ($k_c$ and $C_p$, respectively) as a function of temperature were implemented in the model using a data table. These functions which are shown in Table 5.1. A constant density, $\rho$, of 2670 kg/m$^3$ was implemented as the density of A356 does not change significantly over the range of processing temperatures. The expression for thermal expansion, $\alpha$, given by Hétu et al. is also provided in this table. ABAQUS requires that $\alpha$ be declared relative to a reference
Table 5.1: Thermal properties of A356 and from AISI–4320. Properties for A356 are those employed by Hétu et al. [128] and AISI–4320 properties are from handbook values [129].

<table>
<thead>
<tr>
<th></th>
<th>$C_p$ (J/kg °C)</th>
<th>$k_c$ (W/m °C)</th>
<th>$\alpha$ (°C$^{-1}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A356</td>
<td>$898.7 + 0.4270T$</td>
<td>$7146 + 4.150T$</td>
<td>$2.260 \times 10^{-7} + \left(2.39 \times 10^{-8}\right)T$</td>
</tr>
<tr>
<td>AISI–4320</td>
<td>$452.1 + 0.4740T$</td>
<td>$43.33 - \left(1.090 \times 10^{-2}\right)T$</td>
<td>$6.50 \times 10^{-7}$</td>
</tr>
</tbody>
</table>

Temperature was taken as the total thermal expansion as opposed to differential form [126] such that:

$$\alpha'(T) = \frac{(T - T_0)^{-1}}{T_0} \int_{T_0}^{T} \alpha(T)\,dT \quad (5.1)$$

where $\alpha'(T)$ is the total thermal strain required by ABAQUS at temperature $T$, with reference temperature $T_0$ taken to be 25°C.

In order to describe the thermal expansion of the mandrel solely during the preheating phase of the process, thermal-physical properties for AISI-4320 were necessary. The values for $k_c$, $C_p$, and $\alpha$ were entered in tabular form according to handbook values [129], summarized in Table 5.1. Density was taken to be 7850 kg/m$^3$ for AISI-4320, a modulus of 205 GPa and Poisson’s ratio of 0.29 was prescribed.

### 5.1.3 Initial conditions

As described in Section 2.2.6, workpieces were preheated to 350°C via propane torches. This preheating process was modeled assuming axisymmetry using the implicit solution technique because of the long times involved in the heating process. The results from this implicit simulation were then used as initial conditions for the forming simulation. This included the elastic stress state and temperatures throughout the workpiece, and the geometry of the workpiece and mandrel after the thermal expansion that occurred during preheating.

The resulting geometry and position of the workpiece on the mandrel are shown in Fig. 5.3b. The displaced coordinates of the tracking nodes on both the mandrel and blank were employed to define tooling positions for the forming simulation. In the case of the blank, the tracking nodes were used to define the initial position of the roller. In the case of the mandrel, the tracking nodes were used to define an analytical surface that replaced the meshed
instance of the mandrel weldment as discussed previously.

### 5.1.4 Boundary conditions

Both the preheating and forming simulations require mechanical and thermal boundary conditions to represent the sequence of steps taken during rotary forming, described in Section 2.2.6. Thermal boundary conditions were necessary to heat the workpiece and mandrel during preheating. During forming, the thermal boundary conditions of the workpiece were changed to an adiabatic description and included heat generation due to deformation. The adiabatic state of the blank was then modified as boundary conditions were applied to reflect cooling of the workpiece post-forming. Mechanical boundary conditions were used throughout these stages to control the relative movement of the workpiece and tooling as well as describe the contact conditions.

In the preheating analysis, the mandrel section had node-based constraints applied to the spindle end, such that axial movement was suppressed. This causes the mandrel to expand away from this constraint as the mandrel temperature increases. This represents the conditions present in the EFA where the mandrel is rigidly fixed to the end of the spindle and expands when heated towards the tailstock via the spring-loaded centre. A mechanical contact boundary condition was applied
between the outer diameter of the mandrel and the inner diameter of the workpiece. Most rotary forming simulation studies have assumed frictionless contact [80,83]. However, a recent study employed a Coulomb friction coefficient of 0.2 for tooling interactions encountered during unlubricated spinning of steel [84]. A Coulomb friction coefficient of 0.1 was assigned in the present work to all surface interactions as some friction was expected in spite of the graphite lubricant employed during experiments. A 10 kN force ($F_c$ in Fig. 5.3a) was evenly applied to the clamp face of the workpiece, causing the workpiece to maintain mechanical contact throughout the heating cycle and resulting thermal expansion. The load selected did not induce plastic deformation in the workpiece, but was sufficient to maintain contact.

In developing the thermal boundary conditions for the preheating simulation, inverse thermal analysis using a Levenberg-Marquardt technique [130], was attempted with the experimental TC record (Fig. 2.13) and the domain described in Fig. 5.1. This approach was meant to solve for the time dependent surface flux occurring during flame heating, for use in the preheating model to predict the heat-up of the workpiece from ambient conditions to the forming temperature. It was found that the problem was severely ill-posed owing to the cropped mandrel domain, and the transient heat transfer between the workpiece and the mandrel. The variability observed in the heating rates between experiments further compounded this analysis. The heat rate variability was attributed to geometric variance in the workpieces, resulting in non-uniform axial and circumferential contact with the mandrel. It was determined that the heat transfer between the workpiece and mandrel could not be accurately represented with a displacement-based interfacial heat transfer coefficient. As such, uniform heating conditions were assigned in the preheating model to the mandrel and workpiece to reach the measured temperatures of $T_m = 220^\circ\text{C}$ and $T_b = 350^\circ\text{C}$, respectively, ramped from $30^\circ\text{C}$ linearly over 20 minutes. These boundary conditions ($dT_m/dt$ and $dT_b/dt$) are shown in Fig. 5.3a.

Although the meshed instance of the mandrel was replaced by an analytical surface, the mechanical contact boundary condition between the workpiece and the mandrel employed during preheating was retained. An additional contact boundary condition was specified between the roller and the outer diameter of the workpiece. Adhering to the guidelines set out by Wong et al. [67,69] to reduce the computational expense discussed in Section 1.5.2, the workpiece was kept stationary and the roller was moved rotationally about and along its axis. The workpiece was held station-
ary by replacing the distributed force $F_c$ employed in the preheating simulation (Fig. 5.3a) with a surface constraint fixing the location of this surface (Fig. 5.4a).

During experiments, the roller was moved into contact with the workpiece radially, and then set to move across the workpiece. In order to reflect this, the forming simulation had a roller path, starting from a small radial clearance from the outer diameter of the workpiece, that initially moved radially towards the workpiece to make contact and then transitioned to combined rotational and axial motion. This roller path was calculated based on the geometry and location (on the mandrel) of the preheated workpiece and then implemented in the model via a data table prescribing the motion of the reference node for the rigid analytical surface representing the roller.

In implementing the roller path, the tracking nodes corresponding to the surface of the workpiece were used to define geometry to achieve the forming profiles reported in Table 2.3. The initial
radial position of the roller nose was determined by linearly interpolating between the coordinates of the tracking nodes and an initial clearance of 0.1 mm. The initial axial position for each experimental condition was found by measuring the axial position of the node representing the start of the formed portion of the workpiece, correcting for dilatation and translating this position to the result obtained from the preheating simulation. The path of the roller was then set to move first radially from the initial position \( u_o \) to point \( u_n \) with a penetration of \( P = 0.1 \) mm, while rotating about the \( z \)-axis. Upon reaching \( u_n \), the roller was then moved axially to point \( u_f \) while rotating about the \( z \)-axis. The final axial position of the roller, \( u_f \), was defined as the length of the corresponding experimental workpiece, corrected for temperature, plus 5 mm. This 5 mm clearance was imposed to ensure that the roller moved past the point of contact with the workpiece at the end of the simulation. The proscribed axial movement of 0.21 mm per revolution and circumferential movement of 281 RPM of the roller matched those employed experimentally. This resulted in simulated forming process times \( t_p \) of 61.93 and 87.76 seconds.

As mentioned previously, forming models were run adiabatically, to characterize the heat developed due to dissipation or inelastic heat generation without any heat loss from the workpiece. A Taylor-Quinney factor of \( \beta = 0.9 \) was used.

Once the forming pass was complete, the workpiece was cooled down in the model by applying uniform surface heat fluxes to the surfaces of the workpiece (refer to Fig. 5.4c). A flux of \( q_f = 4.3 \times 10^3(T - 25) \) W/m\(^2\) was applied to both the inner and outer surfaces of the workpiece and \( q_c = 6.4 \times 10^3(T - 25) \) W/m\(^2\) was applied to the clamp surface. With these heat fluxes applied, the workpiece cooled to ambient temperature in 7.5 seconds. The fluxes and the length of time required to cool the formed workpiece do not reflect the experimental conditions where the workpiece was allowed to air cool for 40 minutes. The heat fluxes were selected to limit the simulation time required to reach ambient temperature and to induce no further plasticity in the workpiece. This was conducted in order to compare the experimental workpiece dimensions at ambient temperatures with those predicted by the model.

### 5.2 Material model validation

The material characterization testing documented in Chapter 3 was modelled both to verify the implementation of the extended Ludwik-Hollomon constitutive expression. A common method to
reduce computational overhead in explicit FEA is to employ time or mass scaling, which is discussed subsequently. This model further provided a framework to investigate the effects of this technique. Implicit and explicit solutions of this model were used to validate the constitutive expression with thermal effects directly imposed. Working from these baseline simulations, thermal conditions were then indirectly specified, and various degrees of time and mass scaling were applied. These models were used to evaluate the potential effects of scaling on the fully coupled explicit forming simulation.

A 2D axisymmetric model of a compression test was constructed as shown schematically in Fig. 5.5 and described in Section 2.3.5. The mesh employed consisted of 1200 2D axisymmetric elements (CAX4RT) with a uniform element length of 250 µm in each direction \( (L = L_{\text{max}}, L_{\text{min}}) \). Two mechanical boundary conditions were imposed beyond those inherent due to axisymmetry. First, axial movement of the nodes on the base of the specimen were suppressed, and nodal displacements \( u \) corresponding to the experimental displacement record were imposed on the face opposite to this. This latter boundary condition was implemented on a tabular basis.

5.2.1 Mechanical validation

In order to validate the implementation of the material model and the mechanical boundary conditions, implicit and unscaled explicit simulations were conducted without considering heat transfer. The results from these simulations were compared against the flow stress measured during compression testing. The first test selected (test number 39, Table 3.1) for comparison best reflected forming conditions, as it was conducted at the highest strain rate at approximately 350°C. A second test at approximately the same strain rate, but at a lower temperature that exhibited strain hardening behaviour (test number 24, Table 3.1) was also selected.
CHAPTER 5. MATHEMATICAL MODELLING OF ROTARY FORMING

In the same manner that $u$ was assigned from the experimental displacement record, the experimental temperature record was directly assigned to all nodes. For the purposes of directly comparing the simulation to experimental results, strain was extracted from the nodal displacement record at point ‘X’ on the centerline in Fig. 5.5. The average von Mises stress at the integration point of the elements along the centerline was extracted. The flow stresses for all simulations are compared directly to the relevant experimental flow stresses in Fig. 5.6 and 5.7.

These two figures demonstrate that there is no difference between unscaled explicit and implicit solutions. Furthermore, aside from the flow stress record at low strain, there is reasonable agreement between the model results and the experimental flow stresses. The discrepancy at low strain for both tests is owed to the combined effects of error inherent in the constitutive expression as well as potential non-uniform deformation occurring during the test.

5.2.2 Thermal assessment

As the previous models had temperatures and displacements applied directly through boundary conditions, the absence of thermomechanical coupling in this model formulation is insufficient to fully assess the effects of time and mass scaling. To address this, the thermal boundary conditions were modified. Using the previously discussed compression test data that exhibited work hardening.

**Figure 5.6:** FEA results versus experimental flow stress at forming temperatures. Experimental, implicit and unscaled explicit simulated flow stresses are compared at $T_a = 352^\circ C$ & $\dot{\varepsilon} = 8.82 \text{ s}^{-1}$, validating the material model.

**Figure 5.7:** FEA results versus experimental flow stress with strain hardening. Experimental, implicit and unscaled explicit simulated flow stresses are compared at $T_a = 202^\circ C$ & $\dot{\varepsilon} = 7.58 \text{ s}^{-1}$, further validating the material model.
(\(T_a = 202^\circ C\) and \(\dot{\varepsilon} = 7.58\) s\(^{-1}\), Fig. 5.7), thermal boundary conditions were developed to approximate the heat transfer conditions. The initial temperature of 202\(^\circ C\) was applied uniformly across all nodes, and heat generated due to inelastic deformation was implemented with a Taylor-Quinney factor of \(\beta = 0.9\). The heat transfer from the sample to the IsoT anvils used in the Gleeble was approximated with a heat flux, \(q_H = h(T - T_\infty)\), applied to both ends of the specimen (Fig. 5.5), where \(h = 4 \times 10^3\) W/m\(^2\)\(^\circ C\) and \(T_\infty = 180^\circ C\) throughout the simulation.

The thermal boundary conditions were established through recursive implicit simulations holding \(\beta\) constant and modifying both \(T_\infty\) and \(h\) to establish values close to describing the experimental temperature evolution. In reality, these values vary throughout the compression test, however, these boundary conditions were determined to adequately approximate the heat transfer conditions. The purpose of this exercise was not to accurately assess the heat transfer during a compression test, but to increase the degree of thermomechanical coupling to match the potential degree encompassing a rotary forming simulation.

The flow stresses and temperature response at point ‘X’ (Fig. 5.5) predicted by models with both the explicit and implicit formulations were found to be identical. Fig. 5.8 shows the contours of \(\sigma_{VM}\) and \(T\) for the two solution techniques. Throughout the computational domains, the distribution of stresses are nearly identical, being within 2 MPa at all points. Temperature contours are indistinguishable. Thus, the explicit simulation with these thermal conditions and the aforementioned mechanical boundary conditions can be used to evaluate the effects of time scaling and mass scaling.

5.2.3 Effects of time and mass scaling

While both time and mass scaling strategies are equivalent in terms of reducing computational effort, they are quite different in their implementations. Mass scaling seeks to increase the length of time increments (\(\Delta t\)) by scaling the material density by a factor \((f_m)\) and thereby increasing \(\Delta t\) by decreasing dilatational wave speeds. Time scaling reduces computation time by applying loads faster than in actuality, decreasing the total simulated time \((t_p)\). Both methods are proportional, where the time scaling factor \(f_t = \sqrt{f_m}\). In a coupled framework, mass scaling requires the density of the material and dependent boundary conditions to be scaled. Time scaling requires that material rate sensitivities and thermal boundary conditions be amended.
Implementing a time scaling strategy is more complicated as it requires both the rate dependency of the material and the thermal boundary conditions to be modified. The rate dependency may be accommodated by directly scaling the strain rate in the constitutive behaviour. When applying time scaling to the thermal conditions, the Fourier and Biot (Fo and Bi) numbers in the scaled problem must remain the same according to:

\[
Fo = \frac{(k_c f_i)(t_p/f_t)}{\rho C_p L^2} \quad (5.2)
\]

\[
Bi = \frac{(h f_i)L}{k_c f_i} \quad (5.3)
\]

where the change in process time \(t_p\), which is scaled by \(f_t\), is accommodated by factoring the conductivity \(k_c\) and boundary heat transfer coefficients, \(h\), to retain the same Biot and Fourier numbers.

The baseline explicit model was time scaled by \(f_t\) equal to 10, 25, 50 and 100. These simulations were compared to models with equivalent amounts of mass scaling applied \((f_m = 100, 625, 2500, \text{ and } 10000)\). ABAQUS allows for various implementations of mass scaling such that it is only applied to the mechanical solution and does not affect the thermal characteristics of the model. Although selective \cite{131} and variable mass scaling techniques are possible, uniform mass scaling was employed. Here, \(t_p\) was specified directly and the solver increased \(\rho\) in all elements by the same amount.
amount to decrease the dilatational wave speed. As the constitutive expression renders flow stress a function of temperature, the principal metric to compare the ability of each simulation to track the evolution of stress, strain and temperature was flow stress.

A comparison of the predicted flow stresses with various levels of scaling is provided in Fig. 5.9. This comparison shows that flow stresses for both time and mass scaling are similar up to scaling of factors of 25 and 625, respectively, as compared to unscaled results. Increasing time scaling to 50 also agrees well with unscaled results, however, there is a significant departure apparent for equivalent mass scaling. While the flow stresses predicted by the \( f_t = 50 \) and unscaled simulation are nearly identical, the equivalent mass scaled simulation does not predict the same yield, nor predict the same level of strain. Fig. 5.10 displays the same type of behaviour for the predicted temperature, with temperatures scaled results departing from unscaled results at \( f_m = 2500 \).

Both mass and time scaling strategies result in significantly shortened computation times. This can be quantified with an acceleration factor, \( \bar{f} \), which is defined as the ratio of unscaled simulation computation time to scaled simulation computation time. Table 5.2 provides \( \bar{f} \) for each simulation type and clearly demonstrates the large improvements to computation time that occur with scaling. The \( f_t = 10 \) simulation ran 7.1 times faster and the \( f_m = 100 \) simulation ran 8.86 times faster than the baseline unscaled simulation. Beyond this scaling point, the incremental increase in \( \bar{f} \) becomes smaller for mass scaling as compared to time scaling owed to comparably smaller time increments caused by solution instability (locking). This is because locked portions of the domain

---

**Figure 5.9:** Predicted flow stresses for time and mass scaling.

**Figure 5.10:** Predicted temperatures for time and mass scaling.
require highly distorted elements elsewhere, which causes $L_{\text{min}}$ to decrease much more than a stable solution. However, if the simulation corresponding to $f_t = 10$ is considered to be at the limit of numerically stability, then the equivalent mass scaling factor ($f_m = 100$) provides a 20% improvement in computation time.

Table 5.2: Simulation execution time acceleration factor $\bar{f}$ for time and mass scaling

<table>
<thead>
<tr>
<th>$f_t$</th>
<th>$\bar{f}$</th>
<th>$f_m$</th>
<th>$\bar{f}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>10</td>
<td>7.10</td>
<td>100</td>
<td>8.86</td>
</tr>
<tr>
<td>25</td>
<td>17.72</td>
<td>625</td>
<td>12.00</td>
</tr>
<tr>
<td>50</td>
<td>44.29</td>
<td>2500</td>
<td>24.01</td>
</tr>
<tr>
<td>100</td>
<td>95.15</td>
<td>10000</td>
<td>46.71</td>
</tr>
</tbody>
</table>

Both time and mass scaling provide a reasonably accurate simulated flow stress at $f_t = 10$ and $f_m = 100$ as compared to the unscaled process (Fig. 5.9). However, a comparison of the von Mises stress contours, shown in Fig. 5.11, with the results from the unscaled model (refer to Fig. 5.8) indicate that the time scaled version is in better agreement than the mass scaled (Fig. 5.11a versus 5.11b). However, a single element lying on axis of the specimen ($R = 0$) has locked in the $f_t = 10$ result, which is indicative of numerical instability. This locked element has in turn affected the stress distribution at the midpoint of the domain. This is likely because the domain is sensitive to instability at this location, as this is where the largest stress and temperature gradients occur in the unscaled simulation (Fig. 5.8c and 5.8d).

At $f_t = 50$ and $f_m = 2500$, there is a significant difference between the time and mass scaled versions: the time scaled flow stress matches the unscaled version, while the mass scaled flow stress severely underestimates the unscaled version. As shown in Fig. 5.11c and 5.11d, the contour plots of von Mises stress for both scaling factors do not agree with the stress and deformation distribution seen in the unscaled version. Both models show evidence of element locking, with the time scaled version showing only a few locked elements in the most sensitive portion of the domain. The extent of element locking in the mass scaled version is on a much larger scale, with only a few elements in the centre of the domain remaining unlocked. Not only do the locked elements affect the solution’s accuracy, but also cause the domain’s $L_{\text{min}}$ to inordinately decrease due to distorted elements elsewhere. This in turn leads to shorter time increments and longer computation times, as shown in Table 5.3.
5.2.4 Maximum scaling factor

An appropriate scaling factor is dependent on the model formulation both in terms of domain and boundary conditions. Models containing structures with high natural frequencies and quasi-static loads will be able to accommodate larger scaling factors without much loss in accuracy. As such, isothermal studies of rotary forming that have employed explicit FEA have reported time scaling factors ranging from $f_t = 100$ for a solid cylindrical lead workpiece [69] to $f_t = 6$ for steel sheet [86]. As demonstrated in the preceding, both time and mass scaling provide reasonable solutions as compared to the baseline model up to factors of $f_t = 10$ and $f_m = 100$. At these scaling factors, time scaling provided a better match to unscaled simulations. With increasing scaling factors, the model provided increasingly unstable results, predominantly in the numerically sensitive areas of the domain. Mass scaling was found to provide more unstable results as compared to equivalent time scaling. The largest scaling factor with predictions closest to both experimental data and unscaled simulations was $f_t = 50$. This is in spite of the fact that numerical instabilities were evident in a few of the elements. As will be shown subsequently, the chance for these instabilities to occur is domain dependent, as $f_t = 50$ was employed for all further simulations.
5.3 Preliminary rotary forming modelling

Following the verification of the constitutive behaviour within a coupled thermomechanical explicit FEA framework, preliminary modelling of rotary forming was carried out on a simple abstract version of the EFA process. This preliminary modelling was necessary to examine mesh sensitivity and to assess the computational expense prior to running full-scale models of the EFA process. This preliminary model also permitted verification of the boundary conditions describing the roller movement. These preliminary models use a simplified workpiece geometry in the form of an annulus with approximately the same thickness and (uniform) inner diameter as the EFA workpiece, but one seventh the axial length. These workpiece dimensions allowed the same roller geometry as the full-scale model to be used in defining the rigid analytical surface. A cylindrical, rigid analytical surface was defined for the mandrel to match the inner diameter of the simplified workpiece.

5.3.1 Model description

Based on the geometry and boundary conditions shown in Fig. 5.12, the model was run three ways: 2D axisymmetric and plane strain formulations, and on a fully 3D basis. The analysis consisted of a single forming step lasting an unscaled time \( t_p \) of 2.21 seconds, with roller movement consistent with the EFA experiments. The roller was first brought into radial contact with the simplified workpiece at an initial temperature of 350°C. The roller was then moved axially (and circumferentially in the 3D case) across the outer surface of the simplified workpiece which was held stationary at \( z = 0 \). The mandrel had an initial radial clearance of 0.1 mm from the inner diameter of the simplified workpiece. Contact surfaces were defined between the simplified workpiece, the mandrel and the roller with friction conditions mirroring those used in the full-scale forming simulation.

For the 2D analyses, the roller reference point started at \( u_o \) with a radial clearance of 0.1 mm from the material at \( z = 22.5 \) mm and was then moved radially to penetrate \( P = 1 \) mm into the surface to arrive at \( u_n \). The roller was then moved axially along the simplified workpiece by 5 mm to \( u_f \), deforming the material against the rigid mandrel. For the 3D analysis, the roller was made to rotate approximately 34 times \( (\theta = 213.774 \) radians) about the \( z \) axis with the same axial and radial displacements imposed as in the 2D cases. Fig. 5.13 shows the evolution of the roller position over the course of each simulation. These boundary conditions were implemented in a tabular fashion, with the simulated result being identical to the tabular values.
Figure 5.12: 2D and 3D depiction of preliminary FEA model

Figure 5.13: Roller positioning validation, as shown by the change in the roller’s radial position as seen by the reference node. Both 2D and 3D simulations are shown with $\Delta R$ and $\Delta R \sin \theta$ series, respectively. The resulting simulated roller position shown by the $\Delta R$ Sim. series and on the inset plot.
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Four sequentially finer mesh densities were used to describe the simplified workpiece for the 2D cases and three for 3D. A uniform mesh \( (L = L_{\text{max}}, L_{\text{min}}) \) was used in each case, consisting of quadrilateral reduced-integration elements with hourglass control (CAX4RT and CPE4RT) for the 2D cases and equivalent elements (C3D8RT) for the 3D cases. The mesh densities, characterized by initial element edge length \( (L) \) ranged from 5 mm to 0.625 mm for 2D and 5 mm to 1.25 mm for 3D.

5.3.2 Mechanical results

The predicted stress state and deformed shape at time \( t_p \) for each of the 2D and 3D models of the workpiece are presented in Fig. 5.14. Here, equivalent stress \( (\sigma_{VM}) \) and plastic strain \( (\varepsilon_p) \) are shown for equivalent mesh densities. Little difference is apparent between the 2D descriptions, implying that the diameter of the workpiece is sufficiently large to support a plane strain assumption. The difference in the stress and strain state between the 5 and 2.5 mm mesh cases is very apparent, suggesting that a 5 mm element edge length is too coarse. The difference between 2.5 and 1.25 mm mesh cases is significantly more muted, suggesting that element edge lengths less than 2.5 mm adequately discretizes the domain. This was confirmed by the 0.625 mm 2D simulations (not shown), which displayed similar results to the 1.25 mm case. Even at the finest mesh density, there is no evidence of elements locking in the same manner as was seen previously with the compression test simulations. This indicates that the time scaling factor employed is appropriate for these conditions and domain.

For all mesh densities, however, there is a significant difference between the 2D and 3D results. In the 3D instances, deformation is highly localized at the roller interface. Here, the material is carrying larger load both in terms of shear and \( \sigma_{VM} \) compared to the 2D cases. This is due to the larger plastically affected zone predicted in the 2D approximations, which extends approximately 2.5 mm into the material directly beneath the roller, compared to approximately half that for the 3D predictions. Another appreciable difference is the material pileup ahead of the roller in the 3D case, which is absent in the 2D. Comparing the 3D model results, it appear that the element edge length must be 2.5 mm or smaller in order to predict this phenomena.

Based on the results from these simulations, modelling rotary forming with a domain similar to that of the EFA process requires a 3D description. Both 2D and 3D modelling efforts dictate that
mesh with a characteristic element edge length of 2.5 mm is required, with finer mesh providing more coherent results.

5.3.3 Computational requirements

The resulting memory footprint and solution times using a single processor are given in Table 5.3 for each of the simulations. This data shows that solution times sharply increase when moving to three dimensions, as the 3D model with finest mesh took 66 hours to complete. It is important to note that this computation time reflects approximately 30 hours per second of simulation, even with an aggressive time scaling factor. The cause is twofold: both the problem size, measured in terms of memory requirements, and the number of floating point operations increase with a larger number of solution variables as the critical time step decreases. The latter is also the reason for
the axisymmetric simulations taking marginally longer to complete than equivalent plane strain formulations; the boundary conditions produced a minimum element length $L_{\text{min}}$ that was slightly smaller during the course of the axisymmetric deformation.

In an effort to gauge the improvement in solution times by employing multiple processors, the $L = 0.625$ mm plane strain model was run with multiple processors on a multicore computer. In these tests, the domain was split across 2 to 8 processors with shared memory. Additional processors did improve the solution times appreciably. Moving to two processors showed a 104% improvement, but the rate of improvement showed diminishing returns with additional processors; moving from 7 to 8 processors only displayed a 2% improvement. This is due to increased communication overhead as ABAQUS holds all element calculations on a single processor, and passes node-based calculations on to each processor. The net result is that a finely meshed 3D domain requires significant computational resources, and despite a high level of time scaling and additional processors, results in very long solution times for very short simulation periods.

### 5.4 Thermomechanical EFA modelling

Following the development and verification of the modelling methodology, full-scale, coupled thermomechanical models of the EFA process were run to analyze the forming conditions achieved in the experiments. The application of this modelling methodology to the EFA process provides a large amount of information regarding the process including the ability to track the evolution of the stress state, strain rate, and temperature in the workpiece during forming. The results of the forming model reflecting the mid-deformed and least-deformed workpiece will be presented as an example of the model capabilities. These results will be followed with a comparison of the final workpiece
geometries generated experimentally and those predicted with the model. Finally, the stress state imposed on the workpiece during mid-level deformation will be presented and discussed within the context of surface defect formation.

### 5.4.1 Model results

The forming model provides a great deal of insight into the overall process. One facet is the evolution of the stress-state and deformation of the overall workpiece during forming. This is demonstrated in Fig. 5.15 which shows the distribution of the equivalent stress on the surface of the least-deformed workpiece at the start, midpoint and end of forming. This corresponds with the roller at \( u_n \), at the midpoint of the roller travel, and just prior to the roller leaving the surface of the workpiece. In this figure, the workpiece has been rotated such that axis of symmetry of the roller is parallel with the centerline of the workpiece, affording an orthogonal view of the workpiece along its axis during forming. This orientation is similar to the orientation describing the qualitative stress distribution used by Xu et al. [1], as shown in Fig. 1.15

At the start of forming, the equivalent stress is highly localized about the centerline of the workpiece, and close to being symmetric about the centerline of the workpiece. The bulk of the forming zone, identified by the regions of elevated stress (greater than 125 MPa), is directed circumferentially along the path of the roller, with the maximum contact stress appearing slightly ahead of the centerline. At the midpoint of forming, the stress state has evolved such that the majority of the forming zone remains ahead of the roller, however, significant stresses have evolved elsewhere on the surface. The region ahead of the roller through to the end of the workpiece shows significant load directed axially. At this stage in forming, the overall bounds of the circumferential region carrying an elevated load has increased dramatically from the start of forming. At the end of forming, the stress state returns to being highly localized immediately beneath the roller, ahead of the centerline.

Fig. 5.16 shows an oblique view of the workpiece at the same stages during forming presented in Fig. 5.15. Also shown in this figure are the nodes that are reported as being in contact with the roller at each stage. At the start of forming, a faint ridge can be seen due to the initial contact of the roller, with a relatively small roller contact footprint. At the midpoint of forming, this ridge is more visible and contact has extended to include a small pileup of material ahead of the roller, which results in a larger contact region. The region ahead of the roller in the axial direction demonstrates
Figure 5.15: Predicted equivalent stress state ($\sigma_{VM}$) on the surface of the least-deformed workpiece during forming.

Slight diametral growth, coinciding with the high axial stress levels as compared to behind the roller. At the end of forming, the contact patch has diminished in accordance with the dissipation of material pileup. In this last stage, the formed regions behind the roller exhibit ridges attributed to non-uniform pileup dissipation as the roller location progressed.

Orthogonal views of the deformation and stress states occurring in the mid-deformed workpiece at similar stages during forming are presented in Fig. 5.17. Overall, the stress magnitude is significantly higher than seen with the least-deformed workpiece, coinciding with a more aggressive forming profile. Like the least-deformed workpiece, the peak stresses on the surface of the work-
Figure 5.16: Oblique views of the simulated equivalent stress state immediate to the roller on the surface of the least-deformed workpiece. Inset shows nodes in contact with the roller.
piece occur slightly ahead of the centerline at all forming stages. At the start of forming, the stress state is similar to the least-deformed workpiece in that the contact stress is approximately axially symmetric about the centerline of the roller. However, this stress state embodies a much larger region of stress projected towards the fixed end of the workpiece, which is attributed to bending stress caused by the roller contact. Midway through forming, the stress state has changed dramatically to be quite disparate from the least-deformed workpiece. This is attributed to the workpiece buckling and forming a convex flange ahead of the roller. Two regions of elevated stress appear on the circumference. One aligned with the centerline of the roller, and the other appearing on the flange, behind the roller circumferentially and ahead of the roller axially. At the end of forming, the stress returns to being localized to the vicinity of the roller, however, the axial length shows a large degree of irregularity around the circumference than that seen in the least-deformed workpiece.

Oblique views of the mid-deformed workpiece at the same forming stages are shown in Fig. 5.18 in conjunction with the nodes in contact. This shows that the same number of elements are in contact with the roller as the least-deformed workpiece at comparable forming stages. However, this is the sole similarity. The initial ridge formed by roller contact at the start of forming is much more pronounced. Beyond the formation of the flange midway through forming, the surface previously encountering the roller is much more irregular and a small region of pileup is seen at the edge of the flange closest to the roller. At the end of forming, the flange has collapsed, and the edge of the workpiece shows localized irregularities. The ridges seen in the least-deformed workpiece that were attributed non-uniform pileup dissipation are much more exaggerated in this forming case. However, the ridges in the mid-deformed case are much less radially consistent than in the least-deformed workpiece. Clearly, the simulation predicts significantly less uniform deformation than observed in the least-deformed workpiece, with the workpiece wrinkling as the flange buckles.

Two orthogonal views of the mid-deformed workpiece at the last stage of forming is provided in Fig. 5.19 with contours showing the predicted axial displacement. This result highlights the non-uniform distribution of axial deformation, in particular the formation of lobes that are apparent at the end of forming. This phenomena is discussed further in the next section. To further examine the evolution of other process variables beyond stress and displacement, cross-section views of the simulated workpieces immediately underneath the roller have been extracted at each stage of the forming process. An example of one of these locations is shown in Fig. 5.19.
Figure 5.17: Simulated equivalent stress state ($\sigma_{VM}$) on the surface of the mid-deformed workpiece during forming.

Contours of equivalent stress, equivalent plastic strain, strain rate, and temperature on the cross-sectional planes at each of the three stages in forming of the least-deformed workpiece are shown in Figs. 5.20 - 5.23, respectively. The equivalent stress in the least-deformed workpiece (Fig. 5.20) at the start of forming shows that the region of elevated stress is localized directly beneath the roller and does not extend very far through-thickness. As forming progresses, the stress distribution evolves ahead of the roller, staying predominantly localized to the outer diameter of the workpiece as the workpiece bends to conform with the forming profile. The distribution of plastic strain (Fig. 5.21) mirrors the stress profile, with a peak of 0.7 on the outer diameter and 0.3 on the inner. The strain rate distribution (Fig. 5.20) which encompasses both elastic and plastic strain rates, has a
peak located at the roller interface at the start of forming. Midway through forming, the peak strain rate has shifted to being slightly ahead of the roller and midway through the workpiece thickness. This is attributed to the combined effects of surface deformation and bending. The peak rate stays at approximately 4-4.5 s$^{-1}$ for most of the forming pass, rising to 7 s$^{-1}$ briefly at the end. Reflecting the relatively low amount of strain imparted to the workpiece, the temperature has only increased by approximately 10°C (Fig. 5.23).

Contours of equivalent stress, equivalent plastic strain, strain rate, and temperature on the cross-
sectional planes at each of the three stages in forming of the mid-deformed workpiece are shown in Figs. 5.24 - 5.27, respectively. The equivalent stress state occurring radially in the mid-deformed workpiece (Fig. 5.24) demonstrates a much higher load both on the outer and inner diameter than the least-deformed workpiece. This is likely due to bending stresses developed as soon as the roller contacts the workpiece. As forming progresses, the stress state evolves ahead and to a lesser extent, behind the roller. The highest stress appears in the middle of the workpiece, differing from the least-deformed workpiece as the deformation in mid-deformed workpiece is dominated by bending.

The distribution of equivalent plastic strain (Fig. 5.25) is primarily localized on the outer and inner diameter of the workpiece at all forming stages. This is similar to the least-deformed workpiece, albeit peak strains are an order of magnitude higher. Additionally, at the midpoint of forming, an appreciable plastic zone has developed on the edge of the flange, well ahead of the roller, which is attributed to the start of buckling in the flange. At the end of forming, the flange region exhibits the largest amount of strain caused by the combination of buckling and roller contact during forming.

The strain rate (Fig. 5.26) distribution is similar to that observed in the least-deformed workpiece,
however, the magnitude is approximately 4 times that seen owing to the large amount of buckling. As there was significantly higher amounts of strain applied, the temperature increase is significantly higher, with the peak temperature increasing by 40°C (Fig. 5.23).

5.4.2 Geometric comparison to experimental results

In order to gauge the effectiveness of the model to predict the final shape of the workpiece, the results of the model after cooling the deformed workpiece to room temperature have been compared in two manners. First, predicted workpiece lengths are compared over the entirety of the circumference. Second, the predicted workpiece cross-sections are compared through thickness with corresponding experimental profiles (Fig. 2.14c and 2.14d).

Fig. 5.28 shows variation of the predicted axial length of the workpiece around the circumference for the least-deformed case. The model data shows that there are four discrete minima and maxima appearing about the circumference of the part, corresponding to faint lobes (illustrated in Fig. 5.19). The measured extents of the least-deformed workpiece have been plotted as dashed
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Figure 5.22: Strain rate distribution on cross-sections of the least-deformed workpiece at different forming stages.

Figure 5.23: Temperature distribution on cross-sections of the least-deformed workpiece at different forming stages.

Figure 5.24: Equivalent stress distribution on cross-sections of the mid-deformed workpiece at different forming stages.

Figure 5.25: Equivalent plastic strain distribution on cross-sections of the mid-deformed workpiece at different forming stages.
Figure 5.26: Strain rate distribution on cross-sections of the mid-deformed workpiece at different forming stages.

Figure 5.27: Temperature distribution on cross-sections of the mid-deformed workpiece at different forming stages.

lines for comparison. The solid line coincides with the overall length of the experimental cross-section (Fig. 2.14c) presented in Chapter 4. The experimental workpiece did not exhibit lobes as the minimum and maximum length were offset circumferentially by 180°. However, with the experimental workpiece measuring 149.25±0.55 and the simulated 148.43±0.62, the agreement between the model and experiment is within the length of an element.

Fig. 5.29 compares shape of the experimental cross-section (solid outline) with that predicted by the model (mesh) at location of the maxima (on lobe) and minima (off lobe). The model predicts the development of a convex shape axially along the outer diameter, whereas experimental profile is slightly concave. Both predicted profiles exhibit approximately the same amount of error in describing the experimental cross-section, with the inner radius being approximately 4.5 mm smaller at the midpoint of the deformed region than found experimentally.

Fig. 5.30 presents the same comparison as Fig. 5.28 for the mid-deformed workpiece. For the mid-deformed conditions, the circumferential variation in length predicted by the model is much more pronounced, demonstrating much larger lobes. The final length of the experimental workpiece was 162.15±0.65 mm, where as the model predicts a length of 155.15±3.15 mm. The model
does not predict the overall part length for the mid-deformed conditions nearly as well as the least deformed forming profile. This is also reflected in the comparison of the experimental cross-section with model predictions at the on and off lobe positions (Fig. 5.31). The on lobe position agrees with the experimental cross section for the majority of the length, with the exception of the very end of the workpiece. The off lobe position departs significantly, predicting an inner radius approximately 9 mm smaller at the midpoint of the deformed region.

The primary cause for the large discrepancy between the model prediction and the experimental
results for the mid-deformed workpiece can be attributed to limitations in the model. The model currently predicts localized deformation, but does not predict material fracture. Experimentally, surface cracks were observed on the outer diameter of the workpiece (documented in Section 4.4). Incorporating the prediction of surface cracking in the model would affect the bending stresses significantly as the overall part stiffness would decrease. This would in turn modify the flange buckling phenomena. Based on the experimental results, this had the effect of diminishing wrinkling and limits the formation of lobes.

However, the model does show reasonable agreement with the experimental geometry in the case of the least deformed workpiece. It also permitted a fair prediction of the radial cross-section of the mid-deformed workpiece on lobes. The lack of agreement elsewhere for the mid-deformed workpiece can be attributed to cracking. This shows that the basic deformation mechanism has been successfully modelled.

5.4.3 Surface defect formation

In a study of rotary forming of cast aluminum, Mori et al. [77] demonstrated experimentally that surface cracks similar to those found in the present study occurred in regions of high levels of strain. These levels of strain were identified with a strain rate independent model. In reviewing the results of torsion testing conducted by McQueen et al. [40], fracture in A356 was observed to occur at strains of 1 and 1.5 at 300 and 400°C, respectively (Fig. 1.7), for strain rates up to 5 s⁻¹. The torsion tests also showed that the equivalent stress at fracture was seen to increase with higher strain rates.

Based on this information, the forming model can be used to explain the lack of surface cracks appearing in the least deformed workpiece. The peak strain predicted by the model for this workpiece was 30% less than the fracture strain identified by McQueen et al., at approximately the same temperature and strain rate conditions. In the case of the mid-deformed workpiece, the model predicts equivalent plastic strains of 1.5 or more during early forming, prior to the flange buckling. Furthermore, the predicted strain rate is significantly higher than the range employed by McQueen et al., which would decrease the fracture strain to a greater extent. However, strain and strain rate alone are not sufficient to predict local failure.

Fig. 5.32 shows a contour plot of the principal stress (σ₁) magnitude occurring on a cross-
Figure 5.32: Simulated $\sigma_1$ magnitude and orientation from the forming direction immediate to the roller on the $z-R$ plane. The depicted state was extracted from the mid-deformed simulation, at a quarter way through the axial travel of the roller.

section of the mid-deformed workpiece immediately below the roller. Overlaid on the contour plot is a quiver plot showing the projection of $\sigma_1$ onto the $z-R$ plane, with origins located at integration points. In this figure, the length of each arrow (quiver) represents the principal stress magnitude. This state reflects 17.5 mm of axial roller travel, or approximately 25% through the forming profile, coinciding with a position halfway between those shown in Fig. 5.24-5.27. The region in contact with the roller demonstrates a high degree of compressive stress, however the element immediately behind the roller shows a nearly zero stress state at the surface. The stress state is increasingly tensile moving towards the inner diameter. The crack morphology shown in Fig. 4.14 matches this stress disparity, as annotated by the black dashed line.

Attributing this precise mechanical state to categorically identify the conditions necessary for crack development during forming is speculative at best. This is because the experimental basis for the material model employed does not extend to encompass the strain and strain rate predicted by this simulation, as well as being limited by the data available regarding fracture conditions. The model does, however, provide a framework to include further data with which forming parameters may be modified to conclusively mitigate this phenomena from occurring.

5.5 Summary

The various thermal and mechanical processing steps involved in EFA processing have been modelled as a coupled thermomechanical process in ABAQUS. This consisted of an implicit submodel for preheating the workpiece to forming temperatures, as well as explicit submodels of the forming
operation at elevated temperatures and final cooling to room temperature. The overall modelling
effort contributes significantly to the overall understanding of rotary forming as a whole; hereto-
fore, strain rate and temperature dependency in modelling efforts have not been considered in rotary
forming, much less for A356.

The model development commenced with validating the extended Ludwik-Hollomon’s expres-
sion within ABAQUS, both implicitly and explicitly. The explicit implementation had thermal boundary
conditions changed to develop a thermal gradient in the deformation model based on unequal rates of heat generation due to inelastic deformation and conduction. This baseline model was then
used as a metric to compare both time and mass scaling strategies. Time scaling was shown to pro-
vide reasonable solutions up to factors of $f_t = 50$, which significantly reduced explicit simulation computational overhead.

A simplified abstraction of the EFA process was then used to validate boundary conditions, ex-
amine minimum mesh requirements and accompanying computation resources. The minimum mesh
requirements identified had elements with lengths less than 2.5 mm, and the computational penalty
for further mesh refinement was excessive. Each bisection of the element length resulted in solution
times increasing by an order of magnitude. Parallelization proved to provide little improvement
to the solution time as the problem size grew with mesh density, resulting in large communication overhead between processors.

The overall model results show that the local effects of the roller interface dominate throughout
forming, and that the strain rates achieved are significantly higher than those used to fit the constitut-
tive material behaviour. Changes in material state, manifesting with the evolution of strain rate and
temperature produce non-uniform deformation in the form of wrinkling and the formation of lobes.
Comparing the geometry of the lightly deformed workpiece with that predicted by the model shows
reasonable agreement with the experimental geometry, demonstrating that the model successfully
describes the basic deformation mechanism. Comparing the results for the mid-deformed workpiece
showed that the model did not predict the geometry nearly as well. It has been speculated that this
was due to the inability of the model to predict cracking. As additional data on cracking of A356 at
high strain rates and strains at elevated temperatures becomes available, this model may provide the
ability to modify forming parameters such that rotary forming defects may be avoided.
As covered in Section 1.4, there are a multitude of factors which affect the High Cycle Fatigue (HCF) resilience of A356–T6. Most studies conducted in this area have focused on uniaxial loading conditions, leaving a knowledge gap with regards to the effects of multiaxial loading. The effects of multiaxial loading on fatigue resilience of A356 will be assessed prior to characterizing the effect of rotary forming on fatigue resilience. Fully reversed tension, tension-torsion and torsion fatigue tests were performed and combined with fracture surface observations to analyze the fatigue mechanisms. Basic fatigue criteria were also evaluated within this loading regimen. Kitagawa-type analysis has been performed for the three loading cases, and an estimate of the critical defect size from a multiaxial loading standpoint is defined. The uniaxial HCF fatigue resilience of heat treated rotary formed material is then presented in light of these investigations.

6.1 Multiaxial fatigue characterization

As described in Section 2.1.3, A356–T6 from a variety of sources has been employed to characterize the multiaxial fatigue behaviour. This included material from wedge and wheel castings with a range of DAS and porosity sizes, with some specimens containing artificial defects.

1Portions of this chapter have been published in:

6.1.1 Fatigue test conditions and results

In total, 32 fatigue samples of A356–T6 with various geometries were tested according to the step methodology described in Section 2.3.6. The specimen name, final step loading condition, extraction location and number of cycles to failure employing the step method for each sample is summarized in Table 6.1. The resulting number of steps to failure for each specimen are also provided.

For example, specimen W1 underwent $10^6$ cycles at $\tau_a = 80$ MPa without failure, as calculated at the surface of the specimen. The stress amplitude was increased by 5 MPa to $\tau_a = 85$ MPa, where it also withstood $N = 10^6$ cycles. The stress amplitude was increased again by 5 MPa to $\tau_a = 90$ MPa and the sample failed after $N_f = 7.22 \times 10^5$ cycles on the 3rd loading step.

While the results obtained from this testing are not endurance limits from a statistical standpoint, step testing is the only technique that permits the evaluation of an ‘endurance limit’ for a natural defect of unknown size. The term ‘endurance limit’ is defined as the stress level at fracture after one million loading cycles. In the majority of the fatigue tests, failure occurred after at least one loading step. There were two instances of specimens arriving at the same endurance limit albeit with different numbers of steps. Samples W4 and M2 failed at $\sigma_a = 90$ and $\tau_a = 52$ MPa after 5 and 2 loading steps, respectively. Specimens T1 and T2, both from the top of the wedge, failed at $\sigma_a, \tau_a = 65$ MPa after 5 and 2 loading steps, respectively. Furthermore, specimen T3 was conducted as a run-out test and failed after 908 kilocycles with the same loading amplitude as T1 and T2.

Owing to the non-ferrous nature of this material, these results indicate that A356–T6 is not sensitive to the coaxing effect [132, 133]. It is asserted that the step testing method was able to ascertain the endurance limit for $10^6$ cycles to within 5 MPa.

6.1.2 Fatigue criteria comparison encompassing natural defects

Grouped according to family, Fig. 6.1a depicts $\sigma_f$ versus $\tau_f$ for each specimens containing natural defects (i.e. specimens with artificial defects, family A have not been plotted). Based on the observed decrease in DAS from top to bottom of the wedge, the general trend identified in Fig. 6.1a is that specimens with the largest DAS exhibit the lowest endurance limit. However, specimens extracted from the wheel (W family) show a lower endurance limit than the wedge material at a tension/torsion ratio slightly below pure torsion. This is in spite of having a smaller overall DAS. This difference is less than the range of results shown by the pure torsion testing of the wheel speci-
Table 6.1: Test history of all A356–T6 multiaxial fatigue specimens.

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Loading (MPa)</th>
<th>Steps</th>
<th>$N_f \times 10^5$</th>
<th>√area (µm)</th>
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</thead>
<tbody>
<tr>
<td>W1 a</td>
<td>0 90</td>
<td>3 5</td>
<td>7.22</td>
<td>59³</td>
</tr>
<tr>
<td>W2 a</td>
<td>0 85</td>
<td>1²</td>
<td>3.00</td>
<td>59³</td>
</tr>
<tr>
<td>W3 a</td>
<td>45 78</td>
<td>1²</td>
<td>8.35</td>
<td>59³</td>
</tr>
<tr>
<td>W4 a</td>
<td>90 52</td>
<td>5 5</td>
<td>1.45</td>
<td>59³</td>
</tr>
<tr>
<td>W5 a</td>
<td>70 70</td>
<td>2 5</td>
<td>1.05</td>
<td>59³</td>
</tr>
<tr>
<td>B1 a</td>
<td>50 87</td>
<td>2 5</td>
<td>1.18</td>
<td>59³</td>
</tr>
<tr>
<td>B2 a</td>
<td>90 52</td>
<td>2 5</td>
<td>2.04</td>
<td>59³</td>
</tr>
<tr>
<td>B3 a</td>
<td>70 70</td>
<td>2 5</td>
<td>0.76</td>
<td>90³</td>
</tr>
<tr>
<td>B4 b</td>
<td>0 70</td>
<td>2 5</td>
<td>3.98</td>
<td>39³</td>
</tr>
<tr>
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<td>0 100</td>
<td>3 10</td>
<td>1.51</td>
<td>30³</td>
</tr>
<tr>
<td>B6 a</td>
<td>0 110</td>
<td>2 10</td>
<td>8.83</td>
<td>38³</td>
</tr>
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<td>M1 a</td>
<td>50 87</td>
<td>2 5</td>
<td>2.31</td>
<td>90³</td>
</tr>
<tr>
<td>M2 a</td>
<td>90 52</td>
<td>2 5</td>
<td>4.01</td>
<td>90³</td>
</tr>
<tr>
<td>M3 a</td>
<td>95 0</td>
<td>3 5</td>
<td>0.79</td>
<td>90³</td>
</tr>
<tr>
<td>M4 a</td>
<td>65 65</td>
<td>2 5</td>
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<td>514</td>
</tr>
<tr>
<td>M5 a</td>
<td>70 70</td>
<td>3 5</td>
<td>5.26</td>
<td>53³</td>
</tr>
<tr>
<td>M6 b</td>
<td>0 60</td>
<td>2 10</td>
<td>3.25</td>
<td>531</td>
</tr>
<tr>
<td>M7 b</td>
<td>0 55</td>
<td>1²</td>
<td>NA</td>
<td>2.27 90³</td>
</tr>
<tr>
<td>M8 b</td>
<td>0 60</td>
<td>4 10</td>
<td>2.61</td>
<td>90³</td>
</tr>
<tr>
<td>T1 a</td>
<td>65 65</td>
<td>5 5</td>
<td>4.24</td>
<td>112³</td>
</tr>
<tr>
<td>T2 a</td>
<td>65 65</td>
<td>2 5</td>
<td>1.29</td>
<td>265</td>
</tr>
<tr>
<td>T3 a</td>
<td>65 65</td>
<td>1²</td>
<td>NA</td>
<td>9.08 300</td>
</tr>
<tr>
<td>T4 a</td>
<td>60 60</td>
<td>1²</td>
<td>NA</td>
<td>4.05 496</td>
</tr>
<tr>
<td>T5 c</td>
<td>90 0</td>
<td>5 10</td>
<td>6.63</td>
<td>310</td>
</tr>
<tr>
<td>T6 b</td>
<td>0 50</td>
<td>1²</td>
<td>NA</td>
<td>7.33 265</td>
</tr>
<tr>
<td>T7 b</td>
<td>0 50</td>
<td>2 10</td>
<td>4.84</td>
<td>372</td>
</tr>
<tr>
<td>A1 c</td>
<td>90 0</td>
<td>1²</td>
<td>NA</td>
<td>4.24 398</td>
</tr>
<tr>
<td>A2 c</td>
<td>90 0</td>
<td>3 10</td>
<td>1.29</td>
<td>514</td>
</tr>
<tr>
<td>A3 c</td>
<td>80 0</td>
<td>4 10</td>
<td>9.08</td>
<td>740</td>
</tr>
<tr>
<td>A4 c</td>
<td>70 0</td>
<td>2 10</td>
<td>5.26</td>
<td>760</td>
</tr>
<tr>
<td>A5 b</td>
<td>0 70</td>
<td>4 10</td>
<td>4.84</td>
<td>465</td>
</tr>
<tr>
<td>A6 b</td>
<td>0 50</td>
<td>2 10</td>
<td>6.63</td>
<td>708</td>
</tr>
</tbody>
</table>

³As calculated on the surface of the specimen.
³Fig. 2.22: Tension-torsion specimens W1, W2, B5 and B6 were tested in pure torsion; M3 was tested in pure tension.
³Failed before $10^6$ cycles during the first step
³Estimated based on the maximum √area in Table 2.2
imens, and therefore the material from the wedge performed on par with that from the wheel. Fig. 6.1b plots the \( \sigma_a \) versus \( \tau_a \) data for all of the loading scenarios (Table 6.1) independent of family type. While the tension and tension-torsion tests are tightly grouped, the specimens tested under pure torsion exhibit a greater range of results with \( \tau_a \) at failure between 50 and 110 MPa. With the exception of pure torsion, comparing the maximum endurance limit at all ratios of tension and torsion, \( \sigma_f \) and \( \tau_f \) for \( R_L = -1 \) are approximately equidistant from the origin for all ratios of tension to torsion.

Under pure torsion loading, the straight-gauged tension-torsion type ‘a’ specimens exhibited a higher endurance limit compared to the torsion type ‘b’ specimens. The defect population assessment of different positions in the wedge (Table 2.2) showed that the largest range of defect sizes was observed in the middle of the casting. This location is where \( \sim 80\% \) of the samples loaded under pure torsion were extracted from. During the torsion tests, it was also observed that multiple shear cracks were active at the same time, with the dominant crack not appearing until late in the test. The fractographic observations for this loading condition (Section 6.1.3) combined with the porosity measurements preclude specimen configuration being responsible for scatter. As the tension and combined loading cases produced tightly grouped results, the scatter seen with pure torsion

Figure 6.1: Fatigue testing results: endurance limit grouped by material family type and test points compared to Crossland and MPS criteria. Note that only specimens with natural defects were included in this analysis.
appear to be loading-dependent. However, this specific loading condition may be unimportant as the combined effects of loading and geometry on components will create local multiaxial stress states different from pure shear.

As an exploratory effort to ascertain a basic multiaxial fatigue criteria, the maximum endurance limit from all data sets containing natural defects (i.e. each specimen family) at each tension/torsion ratio was determined to characterize the behaviour for A356-T6. The extracted endurance limit data was compared to the Crossland and Maximum Principal Stress (MPS) fatigue criteria. The Crossland criterion was selected owing to its similarity with other stress-based critical plane approaches under constant amplitude [134–136]. The MPS criterion has been applied as it is relatively simplistic and is often used to assess brittle materials. These criteria were applied to the entire dataset of experimental results, since this is representative of the microstructural differences expected in a large cast component.

Crossland [137] proposed that the second invariant of the deviatoric stress and the maximum hydrostatic stress are the main parameters determining fatigue resilience:

$$\sqrt{J_{2,a}} + \rho_f \sigma_{H,max} \leq A_{CL}$$ (6.1)

where $J_{2,a}$ is the second invariant of the deviatoric stress amplitude and $\sigma_{H,max}$ is the maximum hydrostatic stress. The constant $\rho_f$ is a function of the endurance limit in pure tension and torsion and $A_{CL}$ is the endurance limit in pure torsion. Using the mean $\sigma_f = 89$ MPa and $\tau_f = 67$ MPa based on data given in Table 6.1, $\rho_f$ and $A_{CL}$ were determined to be 0.54 and 67 MPa, respectively.

The MPS criterion asserts that the maximum principal stress must be below a critical threshold such that:

$$\sigma_{1,max} \leq C_{MPS}$$ (6.2)

where $C_{MPS}$ is taken to be the average tensile endurance limit, $\sigma_f = 89$ MPa.

These two criteria are plotted versus $\sigma_a$ and $\tau_a$ in Fig. 6.1b. Both the Crossland and the MPS criteria underestimate the measured endurance limit for combined loading. It should be noted that the Crossland criteria approaches the MPS criteria under tension. Depending on the expected variance in the pure torsion, these results may show that the Crossland criterion is overly conservative.
The Crossland criterion is reliant on the determination of an accurate endurance limit under pure torsion, which experimentally showed significant scatter. Therefore, the MPS criterion is the best of the two criteria to describe these results [55, 59].

6.1.3 Fracture surfaces

A representative summary of the fracture surfaces formed for each type of loading for specimens with natural defects is presented in Fig. 6.2. Under pure tension (Fig. 6.2a), the fracture plane was found to always be normal to the direction of applied stress and thus, coincident with the maximum principal stress ($\sigma_{1,\text{max}}$) plane. Specimens from both the wheel and wedge casting exhibited this behaviour indicating that this observation is independent of microstructure. Furthermore, SEM observations on gauge section material far from the initiation site did not reveal any other cracks. Thus, multiple initiation sites did not manifest under pure tension, regardless of microstructure.

Under combined tension-torsion loading, the fracture surface features are similar to pure tension including regular fracture planes with no bifurcation (Fig. 6.2b) regardless of the sample location. SEM observation of the gage section far from the fracture surface reveals only small secondary cracks less than 50 $\mu$m long. The orientation of the fracture surface in Fig. 6.2d is shown by the fracture surface normal. In this combined loading case where $\sigma_a = \tau_a$, the fracture surface normal is oriented $27^\circ$ from the axis of the specimen. With the maximum principal stress acting at $31^\circ$ from the specimen axis, the fracture surface normal is close to being parallel with this direction. This difference of $4^\circ$ is the largest discrepancy observed over the 18 multiaxial specimens tested. Thus, the orientation of the macroscopic fracture plane under combined loading conditions correlates to the loading condition. More specifically, the fracture plane correlates to the maximum principal stress ($\sigma_{1,\text{max}}$) direction. Therefore, the tension component of the loading must play a major role in determining the path of crack propagation.

Under pure torsion, the failures surfaces observed were very different and showed no similarities to that of pure tension or combined tension-torsion loading. Fig. 6.2c is an example of the fracture surface observed. The tortuous fracture surface has two major crack planes activated: one aligned with the axis of the specimen and the other one perpendicular to this axis. This highlights the difficulty in finding a unique initiation site on the fracture surface. Macroscopic observations confirmed by SEM analysis indicate that there are many different initiation sites over the periphery
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Figure 6.2: Multiaxial fatigue failure types: pure tension (a), combined tension-torsion, $\sigma_a = \tau_a$ (b) and pure torsion (c). Macroscopic crack plane orientation for combined tension-torsion (d). Secondary shear crack on the gage section of a torsion sample far from the fracture surface (e).

Remarkably, there is no evidence of macroscopic cracks growing in the direction normal to the $\sigma_{1,max}$ in the pure torsion scenario. For this loading condition, the crack path is governed by shear as opposed to principal stress. A clear demonstration of the dominance of shear is shown in Fig. 6.2e where cracks propagate in shear mode from early in the fatigue life to the final failure. The very long crack observed in Fig. 6.2e was observed to propagate throughout the test under shear mode III and showed no evidence of bifurcation under mode I. When bifurcation did occur on this sample, a new crack plane extended from the original mode III shear plane and linked with another mode III shear crack in the opposing activated shear plane. The sample shown in Fig. 6.2e exhibited
more than ten other cracks similar to the one depicted, and additional smaller cracks observed in both shear mode III planes.

The fractographic observations indicate that the fracture morphology is independent of the sample family. Therefore, the microstructural features and the defect characteristics, such as DAS and average pore diameter (Table 2.2) do not dictate crack paths under multiaxial loading. The overriding observation from the following analysis is that the macroscopic crack path is governed by $\sigma_{1,\text{max}}$ under multiaxial loading except for pure torsion where shear mode III dominates. This observation for pure torsion is at odds with the general mechanical analysis performed in Section 6.1.2 which asserted that the MPS criteria provided the closest description of the multiaxial fatigue results.

### 6.1.4 Initiation site observations

There are various multi-scale microstructural features that can cause fatigue initiation in A356–T6 [54, 55, 138]. Defects such as gas pores, shrinkage pores, oxides and inter-metallic particles can initiate fatigue cracks. At smaller length scales and in the absence of larger-scale defects, the fatigue properties are dominated by the primary $\alpha$-Al and eutectic characteristics. SEM observations were performed on each sample tested in this work to identify fatigue crack initiation sites and to examine the crack propagation surfaces.

The analysis performed on each sample followed a systematic methodology to reproducibly identify initiation sites and crack surface features. The methodology employed was as follows:

- **Optical microscopy** was employed to observe the fracture surface and identify the fatigue and fast-fracture zones. The fast-fracture zone refers to that portion of the crack surface which developed in the last fatigue cycles.

- **Observations of the fatigue zone** were conducted using SEM to determine the initiation site (within 1 mm$^2$) where river marks on the fracture surface converge.

- **If a clear defect was identified**, the size of the defect was measured using the $\sqrt{\text{area}}$ parameter on the fracture surface. This is performed regardless of the position of the defect relative to the gage surface.

In a number of samples, the initiation site could not be accurately identified or characterized. This was true for pure torsion samples with multiple initiation sites, but also for multiaxial tests where
model III shear cracking lead to fretting damage of fracture surfaces.

Due to the readily available tensile fatigue data for A356–T6, only two specimens were tested under these conditions. Each of these specimens exhibited a gas or shrinkage pore as the initiation site for the fatal crack. The first, M3 (Table 6.1), had an endurance limit of 90 MPa and contained a gas pore with an equivalent diameter of 88 \( \mu \text{m} \) at the initiation site. The other specimen, T5, had a shrinkage pore with an ECD of 370 \( \mu \text{m} \) (Fig. 6.3a) at the initiation site and exhibited an endurance limit of 85 MPa. Similar to pure tension, the combined tension-torsion samples exhibited fracture surfaces that were easily characterized. Fig. 6.3b-6.3e are examples of typical defects that initiated the fatal crack. Fig. 6.3b shows the fracture surface of sample B3 that had a \( \sigma_f \) of 68 MPa. The initiation area on this specimen was readily identified, but a root initiating defect could not be found. Fig. 6.3c is the fracture surface for sample W5, which had the same endurance limit as B3. The fracture surface is less clear but it was possible to identify the initiation area where no clear defects were observed. The fracture surface instead shows friction-generated oxide associated with crack propagation. Fig. 6.3d reveals a 500 \( \mu \text{m} \) pore just below the surface of specimen M4. Remarkably, the endurance limit for M4 was 63 MPa, which is close to the highest value observed. This may be caused by the root defect location relative to the surface of the sample. When a propagating crack does not intersect the sample surface, it is not under ambient environmental conditions, but under vacuum instead. This effect has been investigated in a cast Al-Si-Cu alloy [139] where the fatigue life under vacuum when a surface defect is present is the same as when failure initiates from an internal defect of the same size. These observations have also been confirmed with nodular cast iron [140]. Therefore, a fatigue assessment of A356 should account for the position of the defect with respect to the free surface as there is different damage accumulation dependant on whether the defect lies on the surface or within the bulk. Fig. 6.3c shows a typical shrinkage pore at the surface of specimen T4. In this case, a 500 \( \mu \text{m} \) pore decreased the \( \sigma_f \) to 51 MPa. This result suggests that a surface pore is more detrimental to the fatigue life than a subsurface pore of the same size. The environmental effect is not the only factor in this case because the morphology of the defects may be different.

The fracture surfaces of the samples tested in pure torsion were much more difficult to analyze than the pure tension samples. As shown in Fig. 6.2c, the macroscopic topology of the surface is very complex with two perpendicular mode III cracks activated and multiple initiation sites. Initia-
Figure 6.3: Characteristic fatigue fracture surfaces: specimen T5 ($\sigma_f = 85$ MPa), B3 ($\sigma_f$, $\tau_f = 68$ MPa), W5 ($\sigma_f$, $\tau_f = 68$ MPa), M4 ($\sigma_f$, $\tau_f = 63$ MPa), T4 ($\sigma_f$, $\tau_f = 51$ MPa) and T7 ($\tau_f = 45$ MPa).
tion is spatially distributed and as a result, the fracture surface is a combination of different cracks that have coalesced to cause final failure of the sample. Thus, it was difficult to identify a single initiating feature. However, when the fracture surface was less tortuous, there were features that could be analyzed (Fig. 6.3f). For these samples, it was non-trivial to separate the fatigue zone from final failure zone. Fig. 6.3f shows three features that are presumed fatigue initiation sites. The feature closest to the centre of the specimen is clearly porosity with an equivalent diameter of 265 µm, located well away from the surface. The second and third features are possibly oxide film(s), but the features have been destroyed by damage/oxidation of crack surfaces under mode III propagation.

Under cyclic torsional loading, it is apparent that fatigue mechanisms are related to small, diffuse damage. This suggests that samples tested under pure shear conditions are more susceptible to distributed porosity as compared to the other loading scenarios. The pure torsion tests were also the only specimens tested that experienced no hydrostatic stress. For steel, it has been shown that pure torsion leads to small distributed shear cracks on the surface of the sample while the other loading states cause more localized fatigue damage [141, 142]. However, when a crack in steel initiates on a shear plane, it bifurcates into mode I propagation after a relatively short length (100 - 300 µm).

In the present study, the pre-bifurcation crack length was much larger: on the order of millimeters as shown in Fig. 6.2e. This is important as crack deflection and bifurcation increase the overall damage tolerance for a material. It is surmised that there are few eutectic-Si particles of the correct size, shape and orientation that provide the necessary impetus for crack bifurcation in A356-T6.

6.1.5 Kitagawa analysis of natural and artificial defects

While Section 6.1.4 focussed on the methodology of initiation site observation and qualitative observations, this section aims to apply a quantitative assessment of the impact of defects on the endurance limit. This includes specimens with small, natural defects, as well as those with larger artificial defects applied, as described in Section 2.1.3. The experimental fatigue test results are presented in Fig. 6.4 in the form of Kitagawa diagrams for each of the loading cases. These diagrams have both loading amplitude versus defect size at each step, as well as the endurance limit versus defect size corrected for step testing as outlined in Section 2.3.6. For specimens where the initiation site was unidentifiable, the initiating feature was estimated as the maximum $\sqrt{\text{area}}$ of porosity found via metallography. The use of these results in the Kitagawa analysis is thus speculative. For speci-
mens where there were multiple initiation sites, the largest identifiable defect closest to the surface was characterized. The results of the defect size assessment for each specimen are summarized in the final column of Table 6.1.

**Tensile results**

Fig. 6.4a presents the experimental Kitagawa relationship under pure tension. In all samples, the initial defect size was readily identifiable on the fracture plane that was found to be perpendicular to the direction of the maximum principal stress. The primary finding from the tensile Kitagawa curve is that the critical defect size is relatively large: specimens T5, A1 and A2 have very little impact on the endurance limit (8% reduction). These tensile specimens show that the material appears to be sensitive to defects only when $\sqrt{\text{area}}$ is greater than than 500 µm.

Specimen M3 displayed an oxide-related defect at the origin of the crack linked to subsurface porosity. Specimen T5 failed due to a 400 µm pore that intersected with the surface of the sample (6.3a), while specimen A1 failed due to a 398 µm artificial defect. Since specimens T5 and A1 demonstrated the same endurance limit, it is concluded that the area parameter is able to correlate different types of defects, independent of the nature of the defect. Nevertheless, this finding should be verified with larger defects having a greater impact on the endurance limit. In terms of artificial defects, fracture surfaces for specimens A2, A3 and A4 were very similar to A1, showing that the artificial defect was unmistakably the initiation point.

**Combined tension-torsion results**

The combined tension-torsion Kitagawa diagram, presented in Fig. 6.4b, includes results for specimens with natural defects only. The macroscopic fracture surfaces were similar to those of the tensile specimens: a flat surface in a plane perpendicular to the direction of the maximum principal stress with clear, identifiable initiation sites excepting specimen B3 (Fig. 6.3b). For the specimens tested, the Kitagawa diagram suggests there is a small influence by large defects such as in specimen M4 (Fig. 6.3d), exhibiting a large 500 µm subsurface pore. Below this size, there is no apparent influence of defects on the endurance limit.

The origin of the failure on sample T4 was a shrinkage void near the surface of the specimen. It is of interest to highlight that specimen M4 has approximately the same endurance limit (~67 MPa)
and $\sqrt{\text{area}}$ parameter as specimen T4 (514 versus 496 $\mu$m), reinforcing the independence of defect type and dependence of defect size characterized by $\sqrt{\text{area}}$ on the overall fatigue life.

**Torsion results**

The Kitagawa diagram for the torsion specimens, shown in Fig. 6.4c is similar to the others in that there was an observable effect on endurance limits with increasing defect sizes. The experimental points presented below 100 $\mu$m are for specimens that were unsuccessfully classified by fractography. The data for these specimens has been separated along the horizontal axis based on the porosity assessment of their location in the wedge to render individual tests identifiable. As the endurance limits vary from 55 to 95 MPa for specimens with unidentifiable defects, the Kitagawa
diagram under torsion exhibits a large amount of scatter as compared to the tensile results.

The main complication in identifying defects in the sub 100 µm range is the tortuous nature of the fracture surface, as shown in Fig. 6.2a and described in Section 6.1.4. Cracking was activated on two planes of maximum shear such that the final fracture surface reveals multiple initiation sites. While multiple initiation sites could explain the scatter seen in the Kitagawa plot, careful examination of the fracture surface at each suspected initiation point did not always result in an identifiable defect. An attempt to link the presence of porosity to the multiple initiation sites was made with specimens B4, M7, T6 and T7 by metallography performed on sectioned and polished fracture surfaces. There was little to no deviation found from the porosity measurements given in Table 2.2. In light of these findings, the critical defect size is difficult to assess. Specimens with identifiable defects show a definite decrease in endurance limit beyond 300 µm, which is smaller than under tension. As is the case with the tensile testing, this critical defect size should be clarified by other tests on samples containing 300 µm or larger artificial defects.

6.1.6 Implications for rotary formed material

Up to this point, the fatigue behaviour has been discussed for A356–T6 processed in a standard manner which contained a variety of different defects. The fracture mechanism due to multiaxial loading was seen to be proximate for all loading scenarios aside from pure torsion. This mechanism consisted of cracks initiating at defects, and propagating orthogonal to the direction of maximum principal stress. Kitagawa analysis showed that the endurance limit does not change significantly with defects in the size range of approximately 100–400 µm as characterized by the $\sqrt{\text{area}}$ parameter.

High levels of deformation characteristic of rotary forming has the potential to reduce the mean defect size. In a formed workpiece, defects would be completely eliminated proximate to the outer surface, due to highly localized deformation. Moving away from the roller interface into the bulk of the material, large defects would have their overall size decreased less than the critical defect size identified by the Kitigawa analysis. Smaller defects also have the potential to be mitigated such that crack propagation rates are significantly slowed. Since this crack propagation is largely dependent on MPS, the uniaxial fatigue testing results of rotary formed material discussed in the subsequent section is extensible to multiaxial conditions.
6.2 Fatigue characterization of rotary formed material

As outlined in Section 2.3.6, a series of runout fatigue tests was conducted on fatigue samples extracted from rotary formed materials to investigate the effects of processing on fatigue properties. These tests were conducted on samples from commercially flow formed material, as well as the peak deformed and undeformed EFA material, each with a T6 temper. The load level for these test was selected to achieve a fatigue life of $10^6$ cycles in each sample. Stress-life ($S − N$) data from these tests was augmented with results provided by Bhatnagar [143] for fatigue samples extracted from undeformed-T6 LPDC wheel sections, corresponding to locations similar to ‘AD’ in Fig. 2.6b and having the same DAS as the EFA material. A summary of the tests conducted is presented in Table 6.2 and the resulting $S − N$ data is shown in Fig. 6.5. Geometry for each specimen type is given in Fig. 2.22 with type ‘d’ results plotted in Fig. 6.5a, and type ‘e’ in Fig. 6.5b.

Table 6.2: Runout testing summary: specimen types, condition and quantity employed in the runout fatigue study.

<table>
<thead>
<tr>
<th>Type</th>
<th>Condition</th>
<th>Quantity</th>
<th>Source</th>
</tr>
</thead>
<tbody>
<tr>
<td>d</td>
<td>AC-T6</td>
<td>55</td>
<td>Bhatnagar [143]</td>
</tr>
<tr>
<td></td>
<td>Commercially formed-T6</td>
<td>4 (A)</td>
<td>Fig. 2.6b</td>
</tr>
<tr>
<td></td>
<td></td>
<td>24 (AD)</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>4 (H)</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>4 (HD)</td>
<td></td>
</tr>
<tr>
<td>e</td>
<td>AC-T6</td>
<td>15</td>
<td>Fig. 2.3</td>
</tr>
<tr>
<td></td>
<td>EFA formed-T6</td>
<td>12</td>
<td></td>
</tr>
</tbody>
</table>

In the case of the commercially formed material (Fig. 6.5a), specimens were extracted from four discrete locations, ranging from axial and circumferential/hoop orientations (‘AD’ and ‘HD’, respectively) in a highly deformed region to their lightly deformed counterparts (‘A’, ‘H’). There was negligible difference in the results that could be attributed to orientation as compared to those obtained for the ‘AD’ specimens. At least two out of the four ‘A’ specimens had fatigue lives matching or exceeding ‘AD’ specimens. There is also little difference between heavily deformed circumferential orientations, with an ‘H’ specimen failing before an ‘HD’ specimen at $\sigma_a = 125$ MPa, and the opposite occurring at $\sigma_a = 110$ MPa. However, there were few ‘H’ specimens as compared to ‘A’ specimens and therefore little conclusions can be made regarding the effect of
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Figure 6.5: AC-T6 versus formed-T6 $S - N$ curves. Commercial results employing type ‘d’ specimens in (a), and experimental type ‘e’ in (b). Commercial and experimental results are combined in (c): AC type ‘d’ in I, AC type ‘e’ in II, formed type ‘d’ in III and formed type ‘e’ in IV.

deployment and orientation within the scatter of the results. Regardless, the $S - N$ data of the formed material specimens clearly shows an elevated endurance limit over the undeformed material for all conditions, particularly approaching the low-cycle fatigue regime.

The EFA material (Fig. 6.5b) also shows improved fatigue properties following processing over the undeformed variant. Specimens with $\sigma_u$ between 110 and 120 MPa showed approximately the same range of fatigue lives as the commercially formed material. The same was true for $\sigma_u$ of 160. Other EFA specimens with similar $\sigma_u$ to the commercially formed material show fatigue
lives within 2 orders of magnitude. The undeformed material, however, showed significantly higher endurance limits at similar fatigue lives than that used for the commercially formed comparison. This may be because of the difference in specimen sizes; larger specimens are more likely to contain higher numbers of large defects that affect the endurance limit, and potentially create multiple crack initiation points. The specimens taken from material deformed with the EFA had gauge diameters that were approximately 30% smaller than the commercial.

In order to quantify the impact of forming on the fatigue behaviour, the measured fatigue data was fit to Basquin [99,144] relationships according to:

\[ \hat{\sigma}_f = \sigma'_f \left( 2N_f \right)^{-b} \]  

(6.3)

where \( \hat{\sigma}_f \) is the stress amplitude necessary to arrive at \( N_f \) cycles to failure, \( \sigma'_f \) is a fatigue strength coefficient and \( b \) is the Basquin coefficient. The Basquin coefficient, \( b = 0.17 \), was the same value used to correct the endurance limits acquired via step testing (Section 2.3.6). The results of the fitting procedure are summarized in Table 6.3, including values for \( \sigma'_f \), fitting statistics and the fitted Basquin endurance limit evaluated at \( N_f = 10^6 \) cycles as the key HCF life indicator.

**Table 6.3:** AC-T6 versus formed-T6 Basquin relationships.

| Type | Fig. 6.5c series | Condition | \( \sigma'_f \) (MPa) | \( R^2 \) | RMSE | \( \hat{\sigma}_f \) \( \left| 10^6 \right| \) (MPa) |
|------|------------------|-----------|----------------------|--------|------|------------------|
| d    | I AC-T6          |           | 851.14               | 0.7441 | 21.69| 72.25            |
|      | III Commercially formed-T6 | 1270.02 | 0.8629 | 166.7 | 107.80 |
| e    | II AC-T6         |           | 985.49               | 0.8322 | 15.90| 83.64            |
|      | IV EFA formed-T6 |           | 1215.16              | 0.8850 | 17.32| 103.15           |
| d & e | I & II AC-T6    |           | 880.56               | 0.7454 | 21.49| 74.75            |
|      | III & IV Formed-T6 |       | 1256.70              | 0.8703 | 17.04| 106.67           |

The results of this fitting exercise show that the HCF endurance limit for the commercially flow formed material increases by 33% over the undeformed material, and similarly, a 19% increase is realized for the EFA material. Combining the undeformed and formed sample sets, the predicted endurance limit is increased by 30% after forming. Departure of the data from the Basquin relation-
ship, quantified both in terms of $R^2$ and RMSE, is also greatly diminished for formed materials, for the combined sample sets. The undeformed EFA relationship shows a higher fatigue limit and the RMSE is lower than that of the relationship fitted to Bhatnagar’s data from undeformed-T6 LPDC wheel sections. This is likely due to the larger sample set combined with the increased likelihood of acquiring endurance limiting pores in larger specimens. In the case of the EFA formed material, the RMSE is slightly higher than that of the commercially formed material and has a lower fatigue limit. The commercial forming process imparted significantly higher amounts of deformation than the EFA process, and it is likely that porosity was correspondingly mitigated.

However, porosity-driven effects cannot solely explain the significant increase in endurance limit of the flow formed material. The changes in the eutectic structure imparted by forming likely plays a role as well. As previously discussed, larger eutectic particles with elevated aspect ratios induce diminished elongation. The forming process produces eutectic with smaller and rounder particles post heat treatment; this observation combined with $S-N$ data suggests that fatigue characteristics are also affected. In general, brittle materials show greater increases in crack growth rates over wider ranges of stress intensity factors [145], and the additional ductility afforded by the formed material’s eutectic may reduce crack propagation rates. Furthermore, as eutectic particle debonding dominates crack growth in this material [49], eutectic particle refinement increases the number of particles that require debonding, and introduces more chances for path deflection.

6.3 Summary

Undeformed A356–T6 specimens with a wide range of microstructure were tested under multiaxial, full reversed loading conditions. Tests completed with a variety of tension-torsion ratios were used to evaluate endurance limits and were compared to classical fatigue criteria. These mechanical results showed large scatter for pure torsion loading conditions and were compared to classical fatigue criteria. Fatigue cracks were found to initiate either on casting defects (gas or shrinkage based porosity and oxides) or inside the microstructure. Both scales are in competition for the localization of cyclic plastic deformation that induces the initiation of the crack that leads to failure. The distance to the free surface as well as the morphology are important parameters. This study has provided the following conclusions for A356–T6 processed in a standard manner:

- The experimental results demonstrate average endurance limits of $\sigma_f = 89$ MPa and $\tau_f = 67$
MPa.

- Standard criteria like Maximum Principal Stress (MPS) and Crossland provide mostly conservative estimates of the experimental endurance limits. However, MPS provides the closest fit to the mechanical results.

- Cracking mechanisms are very different depending on the loading type. Under pure torsion the material shows multiple initiation sites and long mode III/I shear crack propagation. The bifurcation to mode I is not observed at the end of the fatigue life. Under tension or combined loading the initiation is shorter and more localized such that crack propagation starts directly in mode I.

- The defect size necessary to affect the endurance limit is relatively large. The critical defect size ranges from 300 µm for pure torsion to 500 µm for pure tension and combined tension-torsion. This suggests such that an overall critical defect size necessary to diminish the multiaxial endurance limit is 400 ±100 µm.

In the case of rotary formed material, a comparison of runout $S - N$ data showed the overall HCF endurance limit was improved by 30% over undeformed material. The commercially processed material showed a 33% increase in endurance limit, and the EFA material showed a 19% improvement. As the commercially processed material was deformed to a much higher extent than that with the EFA, this implies that increased levels of rotary forming improves the fatigue properties of the material accordingly. The cause for this increase is primarily attributed to the mitigation of porosity. A secondary cause may be attributed to the refinement of eutectic particles induced by processing, which in turn slows crack propagation.
7.1 Conclusions

The present work served to experimentally investigate rotary forming of A356 at elevated temperatures, and documented the development of a model reciprocating these experiments focussing on predicting workpiece deformation. The development of this model necessitated a constitutive behaviour description that was developed through experimental thermomechanical testing. The through-process microstructural impact of rotary forming on this material was examined, including the implications for final heat treatment. In order to investigate the effect on fatigue resilience imparted by this process, the existing understanding of A356–T6 fatigue behaviour was augmented to account for multiaxial loadings. These findings were then extended to experimental uniaxial fatigue observations of rotary formed material. Specific conclusions for each aspect of this work are presented in the proceeding.

7.1.1 Constitutive behaviour

Through extensive compression testing, the constitutive behaviour of A356 has been experimentally characterized for a range of strains, temperatures and strain rates. The material displayed a diverse range of thermomechanical behaviour characteristic of Al-Si-Mg alloys, with a transition from strain-hardening to strain rate dependent behaviour occurring at approximately 350°C. Above this temperature, the material was found to be entirely rate-dependent for all strain rates tested. Over the experimental data tested, a Ludwik-Hollomon expression was found to better predict flow stresses than with a Kocks-Mecking type expression.
7.1.2 Rotary formed material characterization

The microstructure of three experimental rotary formed workpieces with increasing levels of deformation have been compared to commercially formed and unprocessed material in both the AC and T6 condition. It was found that changes to DAS imposed by forming have little effect on mechanical properties as characterized by macrohardness measurements. Select thermal treatment of AC material inferred that the AC structure is unstable at temperatures beyond 144°C, and with effects accelerated by higher temperatures. This instability is characterized by structural changes in eutectic-Si and Mg$_2$Si precipitate. Rotary forming at elevated temperatures was found to induce fragmentation of eutectic-Si particles prior to heat treatment, resulting in smaller eutectic particle sizes as compared to undeformed material with the same heat treatment. Precipitation strengthening has been found to be relatively unaffected by heat treatment. Eutectic regions were found to be the propagation paths for cracks on the outer diameter during forming which were apparent under moderate to heavy deformation.

7.1.3 Modelling

A comprehensive coupled-thermomechanical model of an experimental rotary forming process was developed, containing submodels to describe changes in the workpiece during initial heating, forming and final cooling of the component. This model was executed with boundary conditions reflecting a workpiece with light and moderate deformation. It was found to predict process characteristics such as the evolution of temperature, strain and strain rate and stress state within the workpiece. The predicted final workpiece geometry provided by the model showed reasonable agreement with the lightly deformed experimental workpiece where no cracking was observed. The geometry predictions of the final moderately deformed workpiece did not agree with experiments to the same degree as the lightly deformed case. As surface cracks were observed in this workpiece, the primary cause for this discrepancy is attributed to the inability of the model to account for damage evolution. Also in this latter case, the predicted strain rates were found to be in excess of those with which the material model was based. However, as there was good agreement with the lightly deformed workpiece, the basic deformation mechanism of this process has been successfully modeled.
7.1.4 Fatigue

Multiaxial HCF step testing of undeformed A356–T6 has found endurance limits of $\sigma_f = 89$ MPa and $\tau_f = 67$ MPa. These endurance limits were not seen to vary with DAS, and were more sensitive to porosity. A Maximum Principal Stress (MPS) fatigue criterion better described the results as compared to a second invariant based criterion, such as Crossland. Fractographic observations showed that the crack growth mechanisms under pure torsion were different as compared to tension or combined loading. In the two latter cases, the fracture plane was orthogonal to the maximum principal stress direction, which reinforces the MPS criterion. While Kitigawa analysis showed inconclusive results for specimens with small defects, larger defect sizes necessary to diminish the endurance limit for this material was found to be 300 $\mu$m for pure torsion to 500 $\mu$m for pure tension and combined tension-torsion.

Runout stress-life ($S - N$) data of experimentally and commercially formed material as compared undeformed material showed significantly improved fatigue resilience. Elevated increases in endurance limit were found with increased levels of deformation imparted by the process, with an overall increase in endurance limit of 30%. The cause for this increase is primarily due to mitigation of porosity. A secondary cause has been proposed to be eutectic-Si morphology changes which slow crack propagation.

7.2 Future work

The current work spans a great deal of fundamental aspects regarding the use of cast aluminum alloys, in particular, A356. The following items have been identified to extend the current work specifically for rotary forming of cast aluminum alloys at elevated temperatures:

- In describing the constitutive behaviour, the effects of holding the casting at elevated temperatures for a long period of time should be included. This is necessary to account for ageing occurring prior to forming.

- As the model predicted elevated strain rates for a lightly deformed workpiece, commercial applications with higher levels of deformation may encompass strain rates approaching machining. Therefore, material characterization should extend to examining the constitutive behaviour to account for strain rates beyond $10 \text{ s}^{-1}$ at elevated temperatures.
• Continued development of the EFA to allow for controlled tooling movement such that forming passes can be undertaken such that simultaneous axial and radial tool movements are possible. This could be possible through the augmentation of the apparatus with a numerical control system similar to that employed to perform the hardness profiles with the VHTM. This augmentation would closer reflect a commercial process.

• In order to more accurately predict the final geometry of a rotary formed workpiece by FEA, fracture behaviour at elevated temperatures should be investigated. This could be extended to allow for the determination of forming limits for this process.

• It is recommended a more accurate description of thermal boundary conditions be implemented to improve the process model. This could be arrived at by further instrumentation installed on the EFA. For further modelling validation efforts from a mechanical perspective, the EFA roller assembly and mandrel could be instrumented with strain gauges to measure forming loads.

• FEA forming models would be ameliorated with a greater domain resolution than employed in the current study. However, the computational penalties will have to be addressed either through customized code, or specialized hardware to account for the degree of coupling inherent in the process.

• Harnessing the model to evaluate the effects of this process on the in-service performance at the component level, such as the impact of retained stress.

The computational challenges posed by this process with current FEA techniques are the most poignant. If these challenges are addressed, then a large barrier will be removed for more widespread adoption of this process by industry.


This appendix serves to outline the methodology and experiment considerations needed for accurate hardness measurements. These are the practices that were followed in setting up and calibrating all hardness measurement related equipment to deliver results found in this thesis, as well as the conversion from other hardness values reported in the literature to Vickers measurements.

The ASTM E384-11e1 [146] specification applies to both Vickers and Knoop pyramidal indentation, although there is a pre-existing, Vickers-only specification that is available as well [147].

A.1 Relevant formulas

A hardness value is measured by the ratio of force $F$ required to produce an indent with a given surface area of $A$. The $H_V$ number has the force given in kg/mm$^2$, with DPH (Diamond Pyramidal Hardness) numbers being reported in N/mm$^2$. The area of an indent is given as a function of the diagonal lengths and the included angle of the indenter tip, $\alpha$, which is $136 \pm 0.5^\circ$:

$$A = \frac{d^2}{2\sin(\alpha/2)} = \frac{d^2}{1.8544} \quad (A.1)$$

Therefore, the Vickers hardness, $H_V$ in kg/mm$^2$ is equal to

$$H_V = \frac{F}{A} = \frac{2P\sin(\alpha/2)}{d^2} = \frac{1.8544F}{d^2} \quad (A.2)$$

where $P$ is the load in kg$f$. In base SI units, or the DPH number:

$$H_V = \frac{F}{A} = \frac{0.1891F}{d^2} \quad (A.3)$$
where \( F \) is given in Newtons and \( d \) in millimeters.

### A.2 Hardness test methodology

The specimen was placed on a holder that did not allow any rocking or lateral movement to prevent erroneous readings and potentially damaging the diamond. The diamond was periodically checked for damage via a microscope focused on the bottom of an indentation and measuring the apparent tip of the indent with a filar micrometer. The diamond was not used if the apparent tip was larger than 5\% of the overall diagonal \([147]\). The microscope/measurement apparatus was verified to be able to measure indent diagonals within ±5\% or half a \( \mu \)m, whichever was bigger. The indenter tip was periodically cleaned with ethyl alcohol and lens paper.

In all cases, the unmounted thickness of the specimen was \( 1.5 \times \) the diagonal of the indent. This was done to avoid an intersection of the plastically affected zone of the indentation site with the support, or the ‘anvil effect’.

The analysis surface was be finished such that the diagonals were clearly defined/ Cold-finishing was employed exclusively as to not temper or work harden the surface during preparation. In all cases the specimen was supported such that the indentation surface was parallel within 1\°.

In all cases, the indentation centers were be placed at least \( 2.5 \times \) a diagonal apart. This was done to avoid the indentation-affected zone that surrounds each indent from interfering with results. This was verified by creating a grid of indents that were slightly less than \( 5 \times \) apart, and then indents were placed directly between them to see if the hardness was affected. This was validated via statistical methods.

### A.3 Hardness measurement validation

This section summarizes \textit{ASTM} E92-82(2003) “Standard Test Method for Vickers Hardness of Metallic Materials”, Section B, “Verification of Vickers Hardness Testing Machines” \([147]\). This particular methodology only applies for hardness machines that are used for routine testing using standardized gauge blocks of known hardness, the specification also covers machines that are used continually in a laboratory setting. Also note that this standard applies to machines applying loads of \( 1 \text{ kg}_f \) to \( 120 \text{ kg}_f \). For loads less than this, this section only serves as a general guideline.

- A minimum of five Vickers hardness readings shall be taken on at least three blocks having
different levels of hardness with the test force applied with a dwell of 12 seconds.

- For each block, let $d_1, d_2, \ldots, d_5$ be the arithmetic means of each indentation diagonal arranged from smallest to largest.

- The repeatability is given as $R = d_5 - d_1$.

- The error for each block is given as $\bar{d} - d$ where $d = (d_1 + d_2 + \ldots + d_5)/5$ and $\bar{d}$ is the reported mean diagonal on the gauge block.

The machine is considered calibrated if the repeatability and error are within the following limits. The repeatability must be satisfied by the conditions in Table A.1, and $\bar{d}$ and $d$ should not differ by more than 2% or 0.5 µm, whichever is greater. If the repeatability of the machine or error are outside of the prescribed tolerances, then the machine must be calibrated before further use. If the repeatability of the machine is acceptable, but the error is outside of the prescribed limit then statistical methods may be used to correct the data. Note that under optimal conditions, the accuracy that can be expected is the equivalent of 4% of the Vickers hardness number of a standardized reference test block. As a last resort, the reduction in accuracy can be established via statistical methods.

<table>
<thead>
<tr>
<th>Block Hardness Range</th>
<th>$R$ must be</th>
</tr>
</thead>
<tbody>
<tr>
<td>100 to 240, inclusive</td>
<td>&gt; 4% of $d$</td>
</tr>
<tr>
<td>240 to 600, inclusive</td>
<td>&gt; 3% of $d$</td>
</tr>
<tr>
<td>&gt; 600</td>
<td>&gt; 2% of $d$</td>
</tr>
</tbody>
</table>
CONVERSION OF BALL-TYPE TO PYRAMID HARDNESS VALUES

Ball-type indentation to measure the hardness of material is load-dependent. This means that the result obtained with a ball-type indentor with a given load will not be the same if the load is increased or decreased. Furthermore, different load scales are necessary to test both soft and hard materials. The ASTM E140-07 standard [148] provides hardness values spanning Brinell, Rockwell and Vickers hardness measurements for both hard and soft materials. While the load used for Vickers measurements for hard materials is given, the load is not provided for soft materials. It is assumed that the load used for Vickers measurements in this case were in-line with those for Rockwell and Brinell. The conversion for Rockwell-F ($H_{RF}$ 60 kg, 1.59 mm diameter steel sphere) to $H_V$ for cartridge brass is given as:

$$H_V = \left( a + b \cdot H_{RF} + c \cdot H_{RF}^2 + d \cdot H_{RF}^3 \right)^{-1}$$  \hspace{1cm} (B.1)

For the conversion from a 500 kg, 10 mm ball diameter Brinell hardness ($H_B$ 10/500/15) to $H_V$:

$$H_V = a + b \cdot H_{BS}$$  \hspace{1cm} (B.2)

Conversion coefficients for each of the two preceding expressions are given in the following table.

<table>
<thead>
<tr>
<th>Coefficient</th>
<th>$H_{RF}$</th>
<th>$H_B$</th>
</tr>
</thead>
<tbody>
<tr>
<td>$a$</td>
<td>$2.95966 \times 10^{-2}$</td>
<td>-5.60725</td>
</tr>
<tr>
<td>$b$</td>
<td>$-1.03725 \times 10^{-4}$</td>
<td>1.19007</td>
</tr>
<tr>
<td>$c$</td>
<td>$-2.31669 \times 10^{-6}$</td>
<td>-</td>
</tr>
<tr>
<td>$d$</td>
<td>$1.12203 \times 10^{-8}$</td>
<td>-</td>
</tr>
</tbody>
</table>