3 Mechanical Properties

The term superplasticity, in the strictest sense, is defined as the elongation in excess of 500% without necking and failure due to cavitation coalescence. The dominate deformation mechanism is based on diffusion controlled grain boundary sliding (GBS) [1], i.e. the sliding of adjacent grains with respect to each other. The materials normally have grain size smaller than 10 µm and are customarily deformed at temperatures, \( T \), close to 0.9 \( T_m \) and at strain rates, \( \dot{\varepsilon} \), typically around \( 10^{-4} \) s\(^{-1}\).

Aluminum based materials that rely solely on grain boundary sliding may achieve a maximum elongation to failure above 550% [2] associated with \( \dot{\varepsilon} \) approximately equal to \( 10^{-1} \) s\(^{-1}\), which requires the reduction of the grain size in the sub-micrometer or nanometer scale, by applying severe plastic deformation (i.e. Equal-Channel Angular Pressing, or ECAP) [3]. These processes, however, are currently not capable of producing low cost material on the industrial scale.

For volume component production, the most widely used aluminum alloy is the high purity fine grained AA5083 (with grain size equal, or less than 10 µm), which is deformed not only by grain boundary sliding but also by a second process of viscous glide or solute drag creep (SDC) of dislocations. The automotive industry requires alloys with a uniform elongation between 200 and 300%. Since rolled products are preferred for volume component production, “engineering superplasticity” is concerned mostly with materials that do not follow the definition of superplasticity in the strictest sense and for that reason their deformation is often characterized as “enhanced ductility” or “quasi-superplasticity” [4,5]. Since the present study is concerned with this type of materials, the
term “superplasticity” or “coarse-grained superplasticity”, even though it is not exactly applicable, will still be maintained.

A further decrease in the cost of the primary material was found to be possible with the use of coarse-grained AA5083 with an initial grain size of 70 µm, exhibited a maximum elongation to failure in excess of 300% at 440°C and at 10^{-2} s^{-1}. The operation of solute drag creep alone, in coarse-grained AA5083 was found to promote dynamic recovery leading to a significant grain refinement of the microstructure (i.e. the average grain size decreased to 43 µm), and thus, an enhancement of plasticity [6]. Consequently, coarse-grained materials may well fulfill the industrial requirements and within this scope, the use of the low purity coarse-grained AA5182 would constitute the next step for further cost reduction.

In this chapter we present a full account of the mechanical properties of the coarse-grained AA5182 aluminum alloy deformed under uniaxial extension. The anisotropy of the materials was determined by extending at strains, \( \varepsilon \), of up to 250% and measuring regions of the gauge which exhibited the most uniform area reduction. By employing transmission electron microscopy (TEM), an investigation in the dislocation microstructures revealed the deformation behaviour of this material and finally, we discuss our results outlining the potential of this alloy for future automotive applications.

### 3.1 Experimental Procedure

The materials used in the present study were commercial aluminum AA5182 alloys with composition Al – 5.0 wt.% Mg – 0.3 wt.% Mn – 0.1 wt.% Cu, minor impurities of Si, Fe, and Cr and an as-received grain size of approximately 21 µm (20.7
µm) and 37 µm (37.1 µm). In at.% their exact composition was Al – 5.80 Mg – 0.12 Mn – 0.05 Cu – 0.08 Fe – 0.77 Si – 0.01 Cr. Based on their as-received grain size they will be denoted, henceforth, as 21G and 37G. Specimens for tensile tests were produced by Electrical Discharge Machining (EDM) from the 2 and 1.5 mm thick rolled and shortly annealed metallic sheets, for the 21G and the 37G material, respectively. Their gauge direction was parallel, at 45° and at 90° to the rolling direction (RD). To determine the conditions for maximum elongation prior to failure, tensile tests were performed, first, at constant cross-head speed (CCS) at temperatures between 350 and 500°C and at initial \( \dot{\varepsilon} \) between 10^{-1} and 10^{-3} s^{-1}. Subsequently, tests measuring the maximum elongation were carried out at constant true \( \dot{\varepsilon} \) (TSR) of 10^{-1} and 10^{-2} s^{-1} and at \( T \), between 400 and 450°C for the material 21G and between 425 and 500°C for the material 37G. Test up to 50, 100, 150, 200, 250 and wherever possible up to 300% \( \varepsilon \), were performed at 425°C and at true strain rate of 10^{-2} s^{-1} for both materials, since at these conditions the average value of the maximum elongation to failure achieved for the material 21G was the highest. All tensile tests were carried out in air using a computer controlled 810 Material Testing System (MTS) equipped with a three-zone split furnace. Each specimen was placed between two stainless steel grips and fastened with two plates, the furnace was closed and the target temperature was set in the computer control of the facility, so as to achieve the fastest heating rate possible. The target temperature was reached in approximately 15 minutes and the test commenced after a very short stabilization period of about 90 seconds. The short heating cycle was chosen so as to approximate as much as possible, the conditions of blow forming in the industry and to restrict static recrystallization. During the test, the temperature variation in the three zones was less than ± 3°C. Upon failure, the furnace
was immediately opened and the specimens were quenched by water spray prior to removal. The temperature decrease during quenching was larger than $50^\circ$C s$^{-1}$. Because this rapid quenching produced very high contraction, upon reaching the predetermined value of $\varepsilon$, the computer control was set so as to immediately unload the specimens. The $\varepsilon$ distributions over the gauge length were determined from optical photographs (grayscale pixel uncompressed bitmaps) from the face and the profile of each side of the failed specimens, using the GOM – ARAMIS system. For the specimens deformed up to specific $\varepsilon$, the $\varepsilon$ distribution was calculated using one picture for the face and one for the profile of the gauge. The images were combined, aligned, converted into black and white and processed by a MATLAB (Ver. R2007b) program, which enabled the calculation of the area reduction per pixel along the extended gauge. Subsequently, the specimens were sectioned and mechanically polished according to the Struers method with diamond suspensions of decreasing grade sizes and finished with a colloidal alumina suspension, OPU, having a particle size of 30nm. For the investigations of microstructure, electron backscattered diffraction (EBSD) was carried out in an XL-30 Scanning Electron Microscope equipped with a Field Emission Gun (FEG-SEM) and a fluorescent screen operating at 20 kV. The data from the orientation imaging microscopy (OIM) were collected and analyzed using the TSL application developed by AMETEC Inc.—EDAX. The step size of 1$\mu$m was selected so as to compromise between scanning appreciably large areas in certain time and being able to collect more than 40 data points for the majority of the grains. Electron dispersive X-ray analysis (EDS) was carried out on an XL-30 Environmental FEG-SEM. Subsequently, characteristic areas of the gauge were punched, ground to a thickness of 60 $\mu$m, mounted on a copper ring and thinned to
perforation using the GATAN precision ion polishing (PIPS) facility. The dislocation microstructures were observed and recorded in a Jeol 2010F transmission electron microscope (TEM) equipped with a Field Emission Gun (FEGTEM), operating at 200 kV.

3.2 Results

3.3 The As-Received Microstructure

Figure 3.1: Automatic Inverse Pole Figures, obtained by EBSD from the as-received AA5182; (a) material 21G and (b) material 37G

Figure 3.1 presents the inverse pole figures (IPF) of the as-received 21G (a) and 37G (b) material, demonstrating a weak recrystallized texture, comprising equiaxed
grains. Some small non-indexed areas, no more than a few pixels, can be discerned at the top and right of Figure 3.1(a) and middle and bottom of Figure 3.1(b) corresponding, most likely, to precipitates. These were located very often close and/or exactly at the GB interfaces and were identified as being rich in Mn, Cu and Fe and slightly depleted in Al and Mg. Their size most often was between 0.5 – 1.5 \( \mu \text{m} \), but could in rare occasions reach 10 \( \mu \text{m} \) [7]. The average dislocation density as it was determined by TEM was relatively low (i.e. \( 5 \times 10^{10} \pm 10^{10} \text{ m}^{-2} \)).

### 3.4 Mechanical Properties

Representative true stress – engineering strain curves are presented elsewhere [8]. Their shape was completely reproducible up to \( \varepsilon \) where local necking developed prior to the specimen final fracture. Tests performed at \( 10^{-2} \text{ s}^{-1} \) and at \( T \) equal to 425\(^\circ\)C for both materials (see Section 3.3) showed that “visible” local necking was developed well above 250% \( \varepsilon \). Prior to this \( \varepsilon \), the variations of the true stress were recorded and were minimal (mostly much lower than 2 MPa) not only along the stress – strain curve of a particular specimen but also between specimens with the same orientation extended at the same testing conditions. We can assume, therefore, that within a certain degree of accuracy and up to the point where notable local necking has developed, the data are representative of a “steady state” deformation. This \( \varepsilon \), however, at the onset of local necking, could not be predicted for each specimen despite repeated tests (more than ten) at exactly the same conditions. As a result the maximum elongation to failure showed significantly large error margins. Figure 3.2 shows the variation of the maximum elongation to failure with \( T \) for each \( \dot{\varepsilon} \) for the alloy 21G (Figure 3.2a, 3.2c and 3.2e) and 37G (Figure 3.2b, 3.2d and 3.2f). The top two plots represent experiments with specimens cut along the RD, the
Figure 3.2: Maximum elongation to failure against temperature at different strain rates, for the material 21G (a), (c) and (e) and 37G (b), (d) and (f). The areas correspond to experiments at cross-head speed, while the scatter and line plots to those at true strain rate.
middle for tests with specimens at 45° and the last two for tests with specimens perpendicular to the RD. The plots are arranged so as to compare the performance of the two alloys at the same conditions. The colored and patterned areas represent experiments at cross-head speed, blue, red and green for tests at initial strain rate, $\dot{\varepsilon}_0$ of $10^{-1}$, $10^{-2}$ and $10^{-3}$ s$^{-1}$, respectively. Their spread corresponds to the error on the maximum elongation to failure at each temperature. The scatter and line plots represent values from the tests at true strain rate and are colored dark blue and dark red for the $\dot{\varepsilon}$ of $10^{-1}$ and $10^{-2}$ s$^{-1}$, respectively.

It is apparent that at cross-head speed the maximum elongation to failure of the 21G lies consistently above 300% at 425°C and at $10^{-2}$ s$^{-1}$, with values as high as 410% whereas the similar high values at 450°C and at $10^{-1}$ s$^{-1}$ are accompanied by a very large error margin as specimens have demonstrated values as low as 220%. Only for tests with specimens oriented perpendicular to the RD the variation at these conditions was found to be relatively small (i.e. between 340 and 380%). The optimum deformation conditions for the material 21G were reproduced accurately for the experiments carried out at true strain rate, with the specimens along the RD performing slightly better than those cut at 45° and considerably better than those cut perpendicularly to the RD. At cross-head speed and at all the conditions tested, the alloy 37G showed inferior performance at cross-head speed compared with that of the 21G at the optimum conditions. Its only advantage over the 21G was observed at and above 475°C, where its maximum elongation prior to failure was attained systematically at $10^{-1}$ s$^{-1}$ with values close and/or slightly above 300%. At true strain rate conditions along and/or at 45° to the RD the 37G showed the best performance at 500°C and at $10^{-2}$ s$^{-1}$, reaching peak values similar to those of the 21G at
425°C and at $10^{-2} \text{s}^{-1}$ (i.e. 350%). It seems likely that its behavior may be better at these orientations at $T > 500°C$. The use of this material at such conditions, however, may almost certainly require the use of lubricants between the material and the mould. This would increase the forming costs and is, therefore, undesirable by the industry. At $90°$ to the RD, finally, at $500°C$ and at $10^{-2} \text{s}^{-1}$, its performance decreased slightly below 300%, i.e. it exhibited a similar behaviour than that of the specimens of 21G having the same orientation at 425°C and at $10^{-2} \text{s}^{-1}$.

The steady state true flow stress, $\sigma$, was obtained from the stress – strain curves as the average between 5 and 10% $\varepsilon$. These values were compensated by dividing with the Young’s elastic (dynamic) modulus for pure Al [9]:

$$E = 77630 - 12.98 T - 0.03084 T^2$$

(1)

where $E$ is expressed in MPa. The slope on the plots of the logarithm of the strain rate vs. the logarithm of $\sigma/E$ (not shown here, for brevity) is equivalent to the stress exponent, $n$, in the phenomenological equation for creep [1]:

$$\dot{\varepsilon} = A \left( \frac{b}{d} \right)^p \left( \frac{\sigma}{E} \right)^n \exp \left( - \frac{Q_c}{RT} \right)$$

(2)

where $A$, is a constant that depends on the material the stacking fault energy and the deformation mechanism, $b$ is the magnitude of the Burgers vector, $d$ is the grain size, $p$ is the grain-size exponent, $Q_c$, or $Q$ is the activation energy for creep (and/or viscous glide or climb of dislocations) and $R$ is the universal gas constant.

The slope from such plots showed that $n$, does not vary significantly for the strain rates investigated, despite their large differences. The strain rate sensitivity, $m = 1/n$ is plotted as a function of $T$ for experiments carried at cross-head speed (areas) and true strain rate (line and scatter plots) in Figure 3.3. At cross-head speed and at $T$ lower than
Figure 3.3: Plots of the variation of the strain rate sensitivity, $m$ with temperature for the alloys 21G (a) and 37G (b). The areas correspond to experiments at cross-head speed, whereas the scatter and line plots to those at true strain rate.

400°C, $m$ ranges between 0.20 to 0.26 for the alloy 21G and between 0.20 and 0.23 for the 37G. Throughout the entire $T$ regime the error in the variation of $m$ for the former is significantly larger than that of the latter. $m$ may exceed 0.30 at $T > 425°C$ for the 21G but for the coarser 37G that may occur only between 450 and 475°C. At true strain rate, however, $m$ is significantly lower for both alloys. For the coarser 37G, it ranges between 0.20 and 0.26 at 425°C, down to 0.19 to 0.25 at 500°C with a marginally decreasing trend, whereas for the 21G, $m$ demonstrates a significantly larger scatter (e.g. between 0.14 and 0.28, at 400°C, for the specimens perpendicular to the RD) and on average it increases between 400 and 425°C. These low values between 0.20 and 0.25, are most likely indicative that for both alloys, a combination of viscous glide and climb of dislocation occurs during deformation, with the latter mechanism exerting a larger influence, especially for the experiments at true strain rate. The large scattering, however, demonstrated by the alloy 21G at 400°C, resulting in an $m$ value lower than 0.15, most
likely indicates that at this $T$, the deformation at $\dot{\varepsilon}$ of $10^{-2}$ and especially $10^{-1}$ s$^{-1}$ is very close and/or crosses over to the power-law breakdown regime [4,10-11].

Figure 3.4: Plots of the variation of the activation energy, $Q$ with modulus compensated flow stress, $\sigma/E$ for the alloys 21G (a) and 37G (b). The areas correspond to experiments at cross-head speed, whereas the scatter and line plots to those at true strain rate.

The vertical separation between data at different temperatures for the same alloy (i.e. at constant grain size and constant $\sigma/E$) in the plots of the logarithm of the strain rate vs. the logarithm of $\sigma/E$, is proportional to the activation energy of the deformation [12-14], according to:

$$Q = -R \frac{\partial \ln \dot{\varepsilon}}{\partial (1/T)} \bigg|_{\sigma/E}$$

Hence, the result of the linear regression of the $\dot{\varepsilon}$ logarithm with $1/T$ multiplied by $(-R)$ is equal to $Q$. These values plotted against $\sigma/E$ are shown in Figure 3.4. For the alloy 21G, at cross-head speed, irrespective of specimen orientation and at low values of $\sigma/E$ (i.e. at high $T$ and/or low initial $\dot{\varepsilon}$) $Q$ ranges between 97 and 125 kJ mol$^{-1}$ having an average value of close to 110 kJ mol$^{-1}$, significantly lower than that for Mg diffusion in Al, (widely considered equal to 136 kJ mol$^{-1}$), which is associated with solute drag creep, as
well as viscous glide of dislocations [10,15]. Surprisingly, however, the average value of 110 kJ mol$^{-1}$ coincides with that for grain boundary sliding [14,16–17]. At the intermediate values of $\sigma/E$, $Q$ acquired a range from 116 to 150 kJ mol$^{-1}$ and an average value of 133 kJ mol$^{-1}$ which is very close to that of solute drag creep and, finally, at high $\sigma/E$, $Q$ settles at 130 kJ mol$^{-1}$ with a substantially small error margin. Since $Q$ for the self-diffusion of Al has been considered equal to 143 kJ mol$^{-1}$ but it can vary from 120.4 up to 144.4 kJ mol$^{-1}$ depending on the method of calculation [6,18], it is not surprising that for the material 37G, at low $\sigma/E$ (i.e. at high $T$ and/or low initial $\dot{\varepsilon}$) $Q$ ranges between 140 and 156 kJ mol$^{-1}$, with average values of 142, 153 and 147 kJ mol$^{-1}$ for the specimens along, at 45° and perpendicular to the RD demonstrating that at least at these conditions the self diffusion of Al plays a dominant role in the deformation (i.e. dislocation climb seems to be the rate controlling mechanism of deformation). Due to the high temperatures, however, the presence of some of the alloy species present in the matrix (i.e. presumably Cr, Fe and Cu) may provide obstacles in the dislocation motion, at least in high temperatures. With decreasing $T$, however, (and the corresponding $\sigma$ increase) $Q$ consistently decreases below 140 kJ mol$^{-1}$ indicating that at low $T$ dislocation motion may be affected by the Mg diffusion. At true strain rate (i.e. the data depicted by the line and scatter plots) and for the alloy 21G, at intermediate values of $\sigma/E$, $Q$ has values close and/or slightly lower than that of the Mg diffusion but as the $\sigma/E$ increase (at low $T$ and/or high $\sigma$) $Q$ become very high reaching values close to 180 kJ mol$^{-1}$. The variation of $Q$ with $\sigma/E$ showed similar trends irrespective to the orientation of the specimens and the high values may be indicative of obstacles to the dislocations, either by coherent precipitates or by segregation of alloy species, especially close to the grain boundary.
interface, which inhibit their intragranular motion. Similar high values of $Q$ are observed for the alloy 37G. These, however, were consistently calculated, irrespective of the specimen orientation, only at low to intermediated values of $\sigma/E$ and were close to 170–180 kJ mol$^{-1}$. As the value of $\sigma/E$ increased, $Q$ showed a decrease to values between that of Al self-diffusion, for the specimens oriented perpendicularly to the RD and/or that of the Mg diffusion in the fcc Al, for those at 45° to the RD. For the specimens along the RD, however, $Q$ increased again to an average value $\approx$ 170 kJ mol$^{-1}$, with increasing $\sigma/E$, demonstrating, perhaps, a notable anisotropic response for this alloy at low $T$ and/or at high $\dot{\varepsilon}$.

**Table 1** Lankford Coefficients and the parameters $\bar{r}$ and $\Delta r$ for the planar anisotropy of the alloys 21G and 37G.

<table>
<thead>
<tr>
<th>Alloy / Testing Conditions</th>
<th>$r_0$</th>
<th>$r_{45}$</th>
<th>$r_{90}$</th>
<th>$\bar{r}$</th>
<th>$\Delta r$</th>
</tr>
</thead>
<tbody>
<tr>
<td>21G / 425°C, 10$^{-2}$ s$^{-1}$, 50%</td>
<td>0.935</td>
<td>0.961</td>
<td>1.0069</td>
<td>0.966</td>
<td>0.009</td>
</tr>
<tr>
<td>21G / 425°C, 10$^{-2}$ s$^{-1}$, failure</td>
<td>1.106</td>
<td>1.112</td>
<td>1.071</td>
<td>1.100</td>
<td>-0.024</td>
</tr>
<tr>
<td>21G / 425°C, 10$^{-1}$ s$^{-1}$, failure</td>
<td>1.031</td>
<td>1.085</td>
<td>1.096</td>
<td>1.074</td>
<td>-0.022</td>
</tr>
<tr>
<td>37G / 425°C, 10$^{-2}$ s$^{-1}$, 50%</td>
<td>1.017</td>
<td>0.963</td>
<td>1.036</td>
<td>0.995</td>
<td>0.064</td>
</tr>
<tr>
<td>37G / 475°C, 10$^{-2}$ s$^{-1}$, failure</td>
<td>1.163</td>
<td>1.043</td>
<td>1.185</td>
<td>1.109</td>
<td>0.131</td>
</tr>
<tr>
<td>37G / 500°C, 10$^{-2}$ s$^{-1}$, failure</td>
<td>1.130</td>
<td>1.165</td>
<td>1.066</td>
<td>1.132</td>
<td>-0.067</td>
</tr>
<tr>
<td>37G / 475°C, 10$^{-1}$ s$^{-1}$, failure</td>
<td>1.059</td>
<td>1.161</td>
<td>1.137</td>
<td>1.130</td>
<td>-0.063</td>
</tr>
<tr>
<td>37G / 500°C, 10$^{-1}$ s$^{-1}$, failure</td>
<td>1.025</td>
<td>1.124</td>
<td>1.079</td>
<td>1.088</td>
<td>-0.072</td>
</tr>
</tbody>
</table>

**3.5 Anisotropy and Necking Instabilities**

To calculate the planar anisotropy of Al alloys fabricated in the form of sheets, first specimens along, at 45° and perpendicularly to the RD were extended up to a $\varepsilon$ prior to the development of diffuse necking, i.e. where $\sigma$ shows still an increasing trend with $\varepsilon$, within the uniform plastic deformation range. Study of the stress – strain curves, showed
that the onset of diffuse necking was systematically above 50% $\varepsilon$ for both materials deformed at $10^{-2}$ s$^{-1}$ and thus the straining of selected specimens up to 50% would ensure that the specimen was well within the uniform plastic deformation regime. Even though this value is above 20% (according to ASTM E517), it was considered appropriate for specimens exhibiting superplastic properties, since the planar anisotropy has been also customarily estimated at half the fracture strain and/or after the specimen has failed [20].

Optical photography of the specimens using the GOM – ARAMIS system was instrumental in determining the width and the thickness $\varepsilon$, i.e. $\varepsilon_w$ and $\varepsilon_t$, respectively. The Lankford Coefficients as a measure of the plastic anisotropy of a rolled metal sheet, $r_\theta$, the $\bar{r}$ and the $\Delta r$ values for both alloys, are presented in Table 1, as they were determined for specimens deformed up to 50% $\varepsilon$ and from uniformly deformed regions of the gauge in specimens extended up to failure. They were calculated according to the formulae:

$$
\begin{align*}
    r_\theta &= \frac{\varepsilon_w}{\varepsilon_t} \\
    \bar{r} &= \frac{r_0 + 2r_{45} + r_{90}}{4} \\
    \Delta r &= \frac{r_0 - 2r_{45} + r_{90}}{2}
\end{align*}
$$

(4a, b, c)

It becomes immediately apparent that the material 21G is essentially isotropic, whereas the 37G is moderately anisotropic, due to its slightly high $\Delta r$ value. Experiments up to 50% for the alloy 37G at 475 and at 500°C were not carried out since they may produce only marginally better values.

Extensions at 425°C and at $10^{-2}$ s$^{-1}$ up to 200, 250 and 350% $\varepsilon$ showed that the material 21G deformed quite uniformly up to 250% (e.g. Figure 3.5), but profuse local necking developed at 300% $\varepsilon$. The material 37G, however, showed significant local $\varepsilon$ variations even at 200% and, furthermore, not only significant necking but numerous secondary necking instabilities at 250%. Due to the significantly better performance of
the alloy 21G, which can potentially fulfill the industrial criteria, this chapter focuses, henceforth, primarily on this alloy.

Deforming up to failure at the optimum conditions, however, (i.e. at 425°C and at $10^{-2}$ s$^{-1}$) very often resulted in the development of secondary necking instabilities, an unpredicted but frequent behavior, especially when elongations in excess of 400% were achieved [21]. Figure 3.6 shows a specimen of the alloy 21G extended up to almost 500% showing two secondary necking instabilities one to the left and one to the right of the primary failure point. A statistical study of all specimens extended up to failure that
involved photographing each specimen using ARAMIS and measuring the distance
between successive instabilities, including secondary instabilities as well as primary
failure points, from one side of the specimen’s gauge to the other produced Figure 3.7.

Figure 3.6: Specimen from the alloy 21G deformed at 425°C and at 10^{-2} s^{-1} up to 500%
strain showing secondary necking instabilities; the area HT was subjected only to static
recrystallization, SN denotes the secondary necking, UD_T and UD_G are the adjacent
uniformly deformed regions and T is the area tip. A third, but weaker necking instability
can be seen on the left (at approximately 1.5 cm from the left grip).

Figure 3.7: The variation of the distance between successive instabilities with T for each
\dot{\varepsilon}, for the alloy 21G (a) and 37G (b).

For the alloy 21G, the distance between successive instabilities maximizes at the
optimum deformation conditions and at cross-head speed with \dot{\varepsilon}_0 of 10^{-2} s^{-1} it showed a
slightly larger spacing compared with that at true strain rate. At 10^{-1} s^{-1}, largest spacings
were exhibited by the specimens deformed at true strain rate. For the material 37G the
maximum in the distance between successive instabilities was observed at 475°C. The
apparent decrease at 500°C occurred solely because at this $T$, they were far more numerous, i.e. occasionally three and more secondary instabilities were detected in a single specimen, and as a result their apparent spacing decreased. The experiments at cross-head speed systematically showed larger distances between successive necks. For both alloys, finally, they were most likely to appear in specimens along the RD; their presence in specimens at 45° was less frequent and in those perpendicularly to the RD was particularly rare.

**Table 2** Data obtained from the EBSD analysis. The comments refer to the transition from $U_DG$ to secondary necking instabilities SN and to $U_DT$ (i.e. $U_DG\rightarrow SN\rightarrow U_DT$).

<table>
<thead>
<tr>
<th>Parameters / Spec. Orientation</th>
<th>Grain Partition</th>
<th>$U_DG$</th>
<th>SN</th>
<th>$U_DT$</th>
<th>Comments</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Grain Size (µm)</strong> / All Orientations</td>
<td>Average</td>
<td>26.6</td>
<td>24.2</td>
<td>25.6</td>
<td>Grain Refinement: Change $U_DG\rightarrow SN\rightarrow U_DT$ Decrease by 8%, Increase by 4%</td>
</tr>
<tr>
<td></td>
<td>Deformation</td>
<td>35.3</td>
<td>32.5</td>
<td>33.9</td>
<td>Decrease by 15%, Increase by 9%</td>
</tr>
<tr>
<td></td>
<td>Recovery</td>
<td>24.0</td>
<td>20.3</td>
<td>22.2</td>
<td>Decrease by 6%, Increase by 8%</td>
</tr>
<tr>
<td></td>
<td>Recrystallization</td>
<td>15.5</td>
<td>14.5</td>
<td>15.7</td>
<td></td>
</tr>
<tr>
<td><strong>Cube Texture Volume Fraction (%)</strong> / All Orientations</td>
<td>Deformation</td>
<td>26.0</td>
<td>30.1</td>
<td>27.9</td>
<td>Local Maximum on SN Increase $\rightarrow SN$ and constant $\rightarrow U_DT$</td>
</tr>
<tr>
<td></td>
<td>Recovery</td>
<td>16.2</td>
<td>18.7</td>
<td>18.7</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Recrystallization</td>
<td>14.9</td>
<td>20.4</td>
<td>22.2</td>
<td></td>
</tr>
<tr>
<td><strong>Goss Texture Volume Fraction (%)</strong> / All Orientations</td>
<td>Deformation</td>
<td>17.9</td>
<td>19.8</td>
<td>16.2</td>
<td>Local Maximum on SN Increase $\rightarrow SN$ and increase $\rightarrow U_DT$</td>
</tr>
<tr>
<td></td>
<td>Recovery</td>
<td>14.5</td>
<td>18.0</td>
<td>17.8</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Recrystallization</td>
<td>8.6</td>
<td>11.7</td>
<td>10.0</td>
<td></td>
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<tr>
<td><strong>Grain Volume Fraction (%)</strong> / Along the RD</td>
<td>Deformation</td>
<td>6.2</td>
<td>7.4</td>
<td>6.9</td>
<td>Small increase $\rightarrow SN$ very small decrease $\rightarrow U_DT$</td>
</tr>
<tr>
<td></td>
<td>Recovery</td>
<td>11.8</td>
<td>11.4</td>
<td>13.8</td>
<td>Slight decrease $\rightarrow SN$ and increase $\rightarrow U_DT$</td>
</tr>
<tr>
<td></td>
<td>Recrystallization</td>
<td>82.1</td>
<td>81.3</td>
<td>79.4</td>
<td></td>
</tr>
<tr>
<td><strong>Grain Volume Fraction (%)</strong> / At 45° to the RD</td>
<td>Deformation</td>
<td>6.3</td>
<td>11.9</td>
<td>7.5</td>
<td>Large increase $\rightarrow SN$ and large decrease $\rightarrow U_DT$</td>
</tr>
<tr>
<td></td>
<td>Recovery</td>
<td>14.9</td>
<td>19.4</td>
<td>15.3</td>
<td>Large decrease $\rightarrow SN$ and large decrease $\rightarrow U_DT$</td>
</tr>
<tr>
<td></td>
<td>Recrystallization</td>
<td>78.8</td>
<td>68.7</td>
<td>77.2</td>
<td></td>
</tr>
<tr>
<td><strong>Grain Volume Fraction (%)</strong> / Perpendicular to RD</td>
<td>Deformation</td>
<td>6.5</td>
<td>5.1</td>
<td>10.2</td>
<td>Marginal decrease $\rightarrow SN$, large increase $\rightarrow U_DT$</td>
</tr>
<tr>
<td></td>
<td>Recovery</td>
<td>14.8</td>
<td>11.3</td>
<td>15.8</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Recrystallization</td>
<td>78.8</td>
<td>83.6</td>
<td>74.0</td>
<td></td>
</tr>
<tr>
<td><strong>Specimen Orientation</strong></td>
<td>GBs Vol. Fr. (%)</td>
<td>$U_DG$</td>
<td>SN</td>
<td>$U_DT$</td>
<td>Behaviour on SN</td>
</tr>
<tr>
<td>-------------------------------</td>
<td>----------------</td>
<td>-------</td>
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<td>-------</td>
<td>----------</td>
</tr>
<tr>
<td><strong>Along the RD</strong></td>
<td>HAGBs</td>
<td>56</td>
<td>61</td>
<td>56</td>
<td>Local Maximum Recrystallization Dominant</td>
</tr>
<tr>
<td></td>
<td>SGBs</td>
<td>32</td>
<td>25</td>
<td>31</td>
<td>Local Minimum</td>
</tr>
<tr>
<td></td>
<td>LAGBs</td>
<td>12</td>
<td>14</td>
<td>12</td>
<td>Small Local Maximum</td>
</tr>
<tr>
<td><strong>At 45° to the RD</strong></td>
<td>HAGBs</td>
<td>58</td>
<td>55</td>
<td>57</td>
<td>Small Local Minimum</td>
</tr>
<tr>
<td></td>
<td>SGBs</td>
<td>32</td>
<td>28</td>
<td>30</td>
<td>Small Local Minimum</td>
</tr>
<tr>
<td></td>
<td>LAGBs</td>
<td>10</td>
<td>17</td>
<td>13</td>
<td>Sharp Local Maximum Recovery Dominant</td>
</tr>
<tr>
<td><strong>Perpendicular to RD</strong></td>
<td>HAGBs</td>
<td>69</td>
<td>60</td>
<td>66</td>
<td>Sharp Local Minimum</td>
</tr>
<tr>
<td></td>
<td>SGBs</td>
<td>23</td>
<td>34</td>
<td>21</td>
<td>Sharp local Maximum Deformation Dominant</td>
</tr>
<tr>
<td></td>
<td>LAGBs</td>
<td>8</td>
<td>6</td>
<td>13</td>
<td>Constant increase Extensive Recovery in $U_DT$</td>
</tr>
</tbody>
</table>
The cumulative results of the EBSD analysis on the secondary necking instabilities (SN) and on those of the adjacent uniformly deformed sections (i.e. UD₉ and UD₈, for closer to the grip and tip, respectively), with the microstructure partitioned in deformed, recovered and recrystallized grains, according to the criteria outlined in [8], are presented in Table 2.

In summary, Table 2 shows that grain refinement occurred at the secondary necking instabilities. The Cube and Goss texture of the deformed grains maximized. For the specimens along the RD, where the appearance of secondary necking instabilities were most frequently observed, the variations on the volume fractions of the three partitions were minimal with recrystallization being extensive, suggesting that the internal structure and orientation of the grains is mostly responsible for the necking development (i.e. the necking most likely indicates the conclusion of significant internal grain deformation that was followed by recovery and then significant recrystallization). For those at 45° to the RD, deformation is followed by extensive recovery, which results in a comparatively increase of the low-angle grain boundaries at the expense of sub-grain boundaries (SGBs), whereas for the specimens oriented perpendicularly to the RD the apparent increase in the volume fraction of the sub-grain boundaries at the secondary instabilities (deformation) serves as a starting point for more extensive recovery closer to the tip, where a higher volume fraction of low-angle grain boundaries was observed.

3.6 Deformation Microstructures

EDS of the matrix of the deformed specimens showed that it contained 5.38±0.33 at.% Mg, 0.12±0.01 at.% Mn, 0.08±0.01 at.% Cu, 0.04±0.01 at.% Fe and 0.01±0.01
at% Cr and the balance in Al. Figure 3.8, 3.9 and 3.10, show the characteristic microstructures of the areas denoted UDG, UDT and SN, in Figure 3.6. The local engineering $\varepsilon$ in these areas was, 150, 250 and 500%, respectively. Figure 3.8a and 3.8b, obtained in strong beam condition close to the [001] pole. Figure 3.8a shows a bimodal deformation often encountered within a single grain, with two sub-grain boundaries on the left and a dislocation density approximately equal to $4 \times 10^{11} \pm 10^{11}$ m$^{-2}$ and a very high dislocation density on the right, i.e. $4 \times 10^{15} \pm 10^{15}$ m$^{-2}$. Figure 3.8b shows three sub-grain boundaries close to a triple junction in an area relatively free of dislocations. Figures 3.9a, 3.9b and 3.9c, obtained also at [001] beam direction, show microstructures with more uniform dislocation distribution and densities between $10^{14}$ and $10^{15}$ m$^{-2}$. The microstructure of each grain showed between 6 and 10 coherent precipitates (i.e. no size and strain field change, was observed when they were tilted between $\pm g$), with diameter $\approx 0.5$ and 1.0 $\mu$m (e.g. Figure 3.9a) and very often were observed close and/or exactly at the grain boundaries. Some of them acted as sub-grain boundary initiation sites (e.g. Figure 3.9a and 3.9c), which further subdivided the grain structure into sub-grains (Figure 3.9b). In Figure 3.10, the number of sub-grain boundaries decreased markedly. Only a few were observed relatively away from precipitates (e.g. Figure 3.10a). Most grains demonstrated exceptionally uniform dislocation distribution with density $\approx 10^{15}$ m$^{-2}$. Precipitates in the grain interior did not produce dislocation entanglements (i.e. they were successfully cut by dislocations, e.g. Figure 3.10b), whereas they successfully pinned the motion of the low-angle grain boundaries / high-angle grain boundaries (Figure 3.10c).

In summary, slip and climb occur simultaneously with bimodal deformation in the UDG and more uniform distribution in UDT. In those regions, most of the sub-grain
boundaries develop in the vicinity and/or emanate from submicrometer-sized coherent precipitates. In the secondary necking instabilities, the deformation seems to deviate from a “core and mantle” microstructure. The intergranular dislocation motion occurs with relative ease despite the presence of precipitates. Sub-grain boundaries are few and isolated from those precipitates inside the matrix, whereas the precipitates at the GB interfaces act as efficient pinning agents of the low-angle grain boundary and/or high-angle grain boundary migration.

Figure 3.8: Representative microstructures in section UDG: (a) A bimodal deformation showing sub-grain boundaries (left) and high dislocations density(right); (b)Three sub-grain boundaries around a triple junction in a microstructure relatively free of dislocations.
Figure 3.9: Representative microstructures in section UD₁: (a) A precipitate initiating a sub-grain boundary (top) and two more sub-grain boundaries, emanating from triple junctions with a sub-grain (shown at the bottom part of the image); (b) a sub-grain boundary by sub-grain boundaries (white arrowheads) with two precipitates at the top-right corner (black arrowheads) (c) a long sub-grain boundary formed at a junction next to a precipitate.
Figure 3.10: Representative microstructures in the secondary necking section: (a) A long sub-grain boundary far from a precipitate; (b) A high density of uniformly distributed dislocations crosses a precipitate and (c) a precipitate pinning a high angle grain boundary.
3.7 Maximum Elongation, Anisotropy and Failure

Previous studies on the mechanical properties of AA5182, at the same $\dot{\varepsilon}$ (cross-head speed) showed a maximum elongation to failure that was consistently below 150%, even though the post-deformation grain size measured at the grip of the specimens was between 19 and 24 $\mu$m [22], i.e. similar with the as-received alloy 21G. Most tests, however, were carried out below 400°C. A test at 450°C and at $10^{-1}$ s$^{-1}$ showed a maximum elongation to failure of only 95% and this was explained by the fact that the parameter $\dot{\varepsilon}_0/D$ was larger than the value $10^{11}$ m$^2$, which indicates the onset of power-law breakdown (PLB) [23]. $D$ was the lattice self-diffusivity of Al (estimated by using $D_0 = 1.7 \times 10^{-4}$ m$^2$ s$^{-1}$ and $Q = 142$ kJ mol$^{-1}$). Even though comparison in the previous section was carried out using this value for the Al self-diffusivity, an average value from experimental and theoretical parameters recently reported [18,19], produced $D_0 = 8.675 \times 10^{-5}$ m$^2$ s$^{-1}$ and $Q = 128.6$ kJ mol$^{-1}$. Using the latter, the parameter $\dot{\varepsilon}_0/D$ becomes equal to $2.25 \times 10^{12}$ m$^2$, i.e. lower than that indicating power-law breakdown. This, as well as the low values reported for the maximum elongation to failure for the Al-Mg-Mn ternary alloys (even though the latter can be attributed to exceptionally large grain size variations [4]) cannot be easily reconciled with the large values in the present study. In the first study [22], the samples were allowed to stabilize in the testing $T$ for over 1 hour. A heat treatment of the material 21G at 450°C for 1 hour showed a significant grain growth with the average grain size reaching 50 $\mu$m, indicating that the present alloys with the weakly recrystallized texture is essentially in a metastable state (i.e. significant internal energy is still stored resulting in a large driving force for grain boundary motion).
Thus, the good properties demonstrated by the material 21G are most likely due to the fabrication methods.

Examination of the large elongations of the specimens along to the RD using EBSD showed that continuous dynamic recrystallization (CDRX) produced an almost stable average grain size with increasing local ε (with values between 22 and 30 µm in most of the gauge of the post-mortem specimen). The specimens oriented perpendicularly to the RD, however, demonstrated an almost continuous grain refinement from 28 down to 17 µm, which most likely triggered the onset of discontinuous dynamic recrystallization and resulted in premature final fracture [8]. This is the main reason for the specimens that were oriented perpendicularly to the RD exhibiting lower values of maximum elongation prior to failure. In this picture, the larger values demonstrated when the specimens were extended at cross-head speed can be explained in terms of a combination of a continuously decreasing true $\dot{\varepsilon}$ and heating, which resulted in an increased recrystallization rate, reduced grain refinement (i.e. the rate of progress of the continuous dynamic recrystallization was lessened), rendering the average grain size more stable. This is, most likely, the main reason for the similarity in the values of the maximum elongation to failure at cross-head speed irrespective of specimen’s orientation. Failure occurred after the onset of discontinuous dynamic recrystallization which resulted in rapid local necking and final fracture.

In summary, the metastable microstructure of the alloy 21G, most likely responsible for its isotropic response, its fairly uniform elongation up to 250% and its maximum elongation to failure between 300 and 400%, indicates that this alloy is capable of fulfilling the criteria for automotive applications (i.e. uniform elongation between 200
and 300% at high $\varepsilon$). On the other hand, the large grain size of the alloy 37G showed good maximum elongation at higher $T$, but its anisotropic nature and the numerous necking instabilities render it quite unfavorable for industrial use. The large variations in the maximum elongation to failure, however, suggest that proper pressure profiles have to be investigated during blow forming, if the 21G AA5182 is to be utilized successfully for automotive products.

### 3.8 Flow Instabilities and Dislocation Microstructure

Secondary necking instabilities produced by unstable material flow (i.e. flow localization) have been observed when subjecting sub-micrometer sized aluminum alloys in high $\dot{\varepsilon}$ elongations [24], but they were considerably more frequent for coarse grain alloys [4,21]. Shorter gauge length has been documented that it is instrumental in attaining more uniform elongations [25], whereas cavity interlinkage (and not cavity nucleation and growth) limits the uniformity in elongation leading to “quasi-superplastic” behavior [26]. In the present case, the gauge length was only 16 mm, and no significant cavitation interlinkage (not even nucleation and growth) was observed in the regions where secondary necking instabilities instabilities developed. It is more likely that lack of cavitation interlinkage are associated with the decrease of the number of grains in the specimen’s cross-section and/or the increase in the probability of cross-sections containing an uneven volume fraction of soft and hard grains. Thus, cross-sections with more soft grains, i.e. grains in orientations where many slip systems can operate easily and accomplish the imposed deformation, will deform (thin down) preferentially. Indeed, the texture analysis showed that the Cube and Goss component of the deformed grains
maximized in these regions. Grains of material with fcc lattice which possess Cube and Goss orientation have both eight (out of the twelve) active slip systems. Given the fact that the grains with Goss orientation are rotated by 45° around an axis parallel to the RD with respect to that of grains with Cube orientation, their coherency, may probably favour the rotation of deformed grains into alternating Cube and Goss orientations, so as to allow for the operation of multiple slip systems across the grain boundary interfaces. Thus microstructures with a more evenly balanced Cube and Goss texture, are most likely softer and lead to large local elongations. This behavior would be detected in the stress–strain curves only by using an extensometer (an optical device with similarities to the GOM–ARAMIS system), which is capable of dividing the gauge into sections, measure all local elongations across all sections comprising the specimen’s gauge, and record one stress–strain curve for each section. As detected by this method, the development of a secondary necking, a sudden increase in the applied stress would be recorded for that section (larger local elongation would result in a thinner cross section).

In secondary necking local area, continuous dynamic recrystallization has obviously progressed further leading to grain refinement (i.e. a decrease in the average grain size by 5 – 15%). This is accompanied by a slight coarsening in the adjoining uniformly deformed sections. However, the uniformly deformed region that lies closer to the tip is not expected to exhibit grain coarsening up to the same value as that closer to grip, since the former is more likely to have experienced more deformation (i.e. the continuous dynamic recrystallization progressed further). This refinement was observed mostly for the grains exhibiting deformation and especially recovery but their coarsening closer to the tip was about 4 and 9%, respectively. Only the large volume fraction of the
grains that showed recrystallization coarsened up to a size that was slightly higher than that prior to the appearance of the secondary necking instabilities (i.e. coarsened by 8% after the secondary necking instabilities, as opposed to a 6% refinement). Increased recrystallization, portrayed as the conclusion of internal grain deformation and recovery, characterizes the secondary necking instabilities of the specimens along the RD, whereas extensive recovery, associated with the sharp increase of the volume fraction of the low-angle grain boundaries [8], characterizes the secondary necking instabilities of the specimens oriented at 45° to the RD. Consistent with the continuous grain refinement of the specimens oriented perpendicularly to the RD, the instable region demonstrates extensive deformation and serves as a transition area to a more gradual recovery closer to the tip, as the abrupt increase of the sub-grain boundaries in the secondary necking instabilities was succeeded by their conversion to low-angle grain boundaries in the UD_{T}.

This behavior at the optimum deformation conditions (i.e. at 425°C and 10^{-2} s^{-1} for the alloy 21G and at and above 475°C and at 10^{-2} s^{-1} for the 37G) may potentially decrease the applicability of the coarse grained AA5182. Unstable flow is highly undesirable, since it may lead to components of uneven thickness, in regions where forming progressed further than an equivalent extension of 250%. The study of the TEM microstructures showed sub-grain boundaries in all three regions of the gauge. They were far more numerous in the UD_{T}, most appeared to originate from coherent precipitates, and subdivided the microstructure into sub-grains. They were less numerous in the UD_{G} region; the dislocation density in their vicinity was 4 \times 10^{11} \pm 10^{11} m^{-2} and they appeared to be relatively isolated from regions of extensive dislocation glide (with density as high as 4 \times 10^{15} \pm 10^{15} m^{-2}) and, finally, only a couple were observed in the secondary necking
instabilities. Their presence in the uniformly deformed regions suggests that, at these conditions, the material follows the high stress “five-power-low” creep, where recovery dominates and dislocation climb, as manifested by the sub-grain boundary formation, is rate controlling [10,27-29]. Thus in the secondary necking instabilities, where the cross-section has decreased (and the effective applied stress has increased) the material crosses over to the power-law-break-down regime, where thermally activated glide presumably dominates [11,30], with the coherent precipitates pinning effectively the drastic grain modification of the microstructure [31,32].

3.9 Mechanical Properties and Deformation Behavior

The fact that the material response lies between the power five law and power-law-break-down regime is reflected in the low values of $m$. These lie between 0.20 and 0.30 at cross-head speed and for the alloy 37G at true strain rate, but can decrease significantly below 0.20 for the material 21G at true strain rate, around the optimum deformation conditions, especially for the experiments carried out along the rolling direction (the thickest olive green line and scatter plot in Figure 3.4a). At 425°C, the specimens oriented perpendicularly to the RD exhibit the largest scatter in the value of $m$, presumably due to extensive continuous dynamic recrystallization which produces an increasingly finer grain size, thus leading to premature failure (i.e. a low maximum elongation). Admittedly, these values were calculated from the values of $\sigma$ between 5 and 10% $\varepsilon$, i.e. prior to the development of diffuse necking and/or the onset of secondary necking instabilities, but there seems to be an agreement with the TEM observations of the post-mortem microstructures as well as with the interpretation of the macroscopic
cross-section decrease (with the corresponding increase in the $\sigma$) during the development of secondary necking instabilities. One further evidence supporting the response type of the present materials is provided by the values of $\sigma/\mu$, which at 425°C and at $10^{-2}$ s$^{-1}$ for the material 21G is equal to $1.8 \times 10^{-3}$. For the alloy 37G at the same $\dot{\varepsilon}$ and at 475 and 500°C are equal to $1.3-1.4 \times 10^{-3}$ and $1.1 \times 10^{-3}$. All these values are very close to $10^{-3}$ (i.e. marginally higher), which characterizes the transition to power low behavior [33]. All tests at $\dot{\varepsilon}$ equal to $10^{-1}$ s$^{-1}$, however, even at cross-head speed, lie well within the power-law break down regime. These affect mostly $Q$ at high $\sigma/E$ (low $T$ and/or high $\sigma$), which for the material 21G increase consistently up to 185 kJ mol$^{-1}$, and for the material 37G reach 170–180 kJ mol$^{-1}$ at low to intermediate values of $\sigma/E$ (low $T$) and 170 kJ mol$^{-1}$ at high $\sigma$, for specimens along the RD.

One important issue that is necessary to clarify concerns the validity of the thermally activated glide and obstacle controlled glide [11,34-36] in the region of power-law-break-down. These studies, showed that the former is associated with $Q$ between 171 and 175 kJ mol$^{-1}$ and that the high $\sigma$ associated with this regime indicate that the dislocations have broken away from their solute atmosphere. The latter, on the other hand, is associated with $Q$ lower than that of self-diffusion. These mechanisms are essentially the same [37], but in the present case some considerations regarding the nature of these obstacles may prove quite useful in explaining the high dislocation densities encountered in the TEM microstructures and the deviation from the core and mantle model (i.e. the unexpected uniform dislocation distribution) at the secondary necking instabilities. The average critical velocity, $\bar{\varepsilon}_c$, above which solute atmospheres cannot be formed around the dislocations is given by the relationship [6]:

\[ 63 \]
where \( A \) is the elastic energy interaction equal to \( 3\mu b\Omega \varepsilon_a / \pi \), \( \mu \) the temperature dependent shear modulus, \( b \) the magnitude of the Burgers vector (\( b_{\text{Al}} = 2.8635 \, \text{Å} \)), \( \Omega \) the atomic volume of a particular solute in Al, \( \varepsilon_a \) the absolute value of the misfit strain i.e. the absolute value of the change of the lattice constant after replacing an Al with a solute atom (or the solute-solvent size difference [29]) and \( D_s \) is the diffusivity of the solute. Here, the volumes of all solutes in fcc Al and the absolute value of the linear size factor (lsf) were used for \( \Omega \) and \( \varepsilon_a \), respectively [38]. For the diffusivity of each solute, the average from experimental and theoretical values were used (i.e. \( D_{0-\text{Mg}} = 4.521 \times 10^{-5} \, \text{m}^2 \text{s}^{-1} \), \( Q_{\text{Mg}} = 125.7 \, \text{kJ mol}^{-1} \), \( D_{0-\text{Mn}} = 3.051 \times 10^{-2} \, \text{m}^2 \text{s}^{-1} \), \( Q_{\text{Mn}} = 219.8 \, \text{kJ mol}^{-1} \), \( D_{0-\text{Cu}} = 2.11 \times 10^{-4} \, \text{m}^2 \text{s}^{-1} \), \( Q_{\text{Cu}} = 127.2 \, \text{kJ mol}^{-1} \), \( D_{0-\text{Fe}} = 4.178 \times 10^{-1} \, \text{m}^2 \text{s}^{-1} \), \( Q_{\text{Fe}} = 213.2 \, \text{kJ mol}^{-1} \), \( D_{0-\text{Si}} = 4.604 \times 10^{-5} \, \text{m}^2 \text{s}^{-1} \), \( Q_{\text{Si}} = 121.8 \, \text{kJ mol}^{-1} \), \( D_{0-\text{Cr}} = 3.534 \times 10^{-1} \, \text{m}^2 \text{s}^{-1} \), \( Q_{\text{Cr}} = 251.8 \, \text{kJ mol}^{-1} \) and \( D_{0-\text{Ti}} = 1.12 \times 10^{-1} \, \text{m}^2 \text{s}^{-1} \), \( Q_{\text{Ti}} = 241.4 \, \text{kJ mol}^{-1} \) [18,19]. Figure 3.11, shows the \( \bar{v} \), as a function of \( T \). For \( \dot{\varepsilon} \) equal to \( 10^2 \, \text{s}^{-1} \) an average dislocation velocity \( \bar{v} \) of \( 10^{-4} \, \text{m s}^{-1} \) requires a mobile dislocation density of \( 3.5 \times 10^{11} \, \text{m}^{-2} \), i.e. four orders of magnitude lower that the total dislocation density observed. Since Mg concentrations more than 2 at.% result in super-saturation of the dislocation cores, it can be concluded that the solute drag of mobile dislocations by Mg solutes dominates in AA5182 [23]. Even though Mn, Cu, Fe, Si and Cr have been mostly detected in the precipitates, a possible minute content in the matrix of both alloys is expected, but it is well below their solubility limit in Al. These atoms are approximately 46, 38, 37, 16% and 57% smaller than the Al atoms, and
since Cu and Si can also reach the dislocation cores they may well contribute to the solute drag, or even pin the dislocation motion.

![Critical dislocation velocity for the formation of solute atmospheres in AA5182. The gray shaded area corresponds to the T regime used in the present study. The velocities above the curve of a particular solute correspond to values where that solute is unable to reach the dislocation cores, whereas below solute drag and/or dislocation pinning occurs by that solute.](image)

In summary, solute drag by Mg atoms dominates at $\dot{\varepsilon} = 10^{-2}$. Cu and Si atoms present in the matrix inhibit the dislocation motion. Dislocations that encounter occasional Fe, Mn, Cr or Ti atoms may be arrested and rendered immobile requiring dislocation multiplication mechanisms for continued deformation. Dislocation glide and climb occur concurrently with the latter being the rate controlling mechanism. The coherent precipitates inside the grain microstructure can be cut successfully by dislocations especially when the applied stress locally increases. The local $\varepsilon$ difference by 250% between secondary necking instabilities and UD$_T$ of Figure 3.6, corresponds to an increase of approximately 2.5 times in the effective stress and, consequently, by a maximum of $2.5^5 \approx 100$ times in the average dislocation velocity. Hence, if the mobile
dislocations move with a $\bar{\nu} \approx 10^{-4}$ m s$^{-1}$ in the uniformly deformed regions, they will move with a maximum $\bar{\nu} \approx 10^{-2}$ m s$^{-1}$ in the secondary necking instabilities and thus, most will escape their solute atmospheres. In such cases, their motion will be controlled strictly by thermal activation processes that exhibit high $Q$ values. The contribution of the tests at $10^{-1}$ s$^{-1}$, which lie well into the power-law breakdown regime is, most likely, the reason for the increase of $Q$, from that of the Al self-diffusion and Mg diffusion in the fcc Al, considered in this study (i.e. 128.6 and 125.7 kJ mol$^{-1}$, respectively) to those calculated. Finally, the average value of 110 kJ mol$^{-1}$ for the material 21G at low $\sigma/E$ shows that at these conditions grain boundary sliding becomes important. But due to the presence of coherent precipitates grain boundary sliding is inhibited and thus, failure occurs rapidly with a very low maximum elongation.

### 3.10 Conclusions

- The coarse-grained AA5182 aluminum alloys 21G and 37G, exhibited optimum deformation conditions at $\dot{\varepsilon}$ equal to $10^{-2}$ s$^{-1}$ and at $T$ equal to 425°C and above 475°C, respectively, with maximum elongations to failure between 300 and 400% along, and at 45° to the RD and approximately equal to 300% perpendicularly to the RD. These large values are mostly due to their metastable, weakly recrystallized as-received microstructure.
- The alloy 21G is essentially isotropic exhibiting consistently uniform deformation at all orientations up to an elongation of 250%, at 425°C and at $10^{-2}$ s$^{-1}$, whereas the 37G is slightly anisotropic and demonstrated significant deviations from uniformity.
at 200%, as well as significant necking and numerous secondary necking instabilities at 250%.

- Secondary necking instabilities were most likely to appear at the optimum deformation conditions, especially when elongations in excess of 400% were achieved, but their development could not be very well predicted. For both alloys they were most likely to appear in specimens along the RD; their presence in specimens at 45° was less frequent and in those perpendicularly to the RD was particularly rare. They are associated with regions of the gauge that contain a large volume fraction of soft grains and produce microstructures that exhibit maxima in the Cube and Goss component of the deformed grains and slight grain refinement compared with the adjoining thicker and more uniformly deformed regions.

- The deformation of both alloys at $10^{-2}$ s$^{-1}$ lies at the border between the high stress power five law regime, where dislocation climb is the rate controlling mechanism and the power-law breakdown phenomenon.

- The microstructure of the secondary necking instabilities is indicative of power-law breakdown deformation, where the dislocations have acquired high average velocity and thus have managed to break away form their solute atmospheres. Hence thermal activation processes with high $Q$ control the deformation.

- The high activation energies calculated in this study are most likely due to the influence of the tests at $10^{-1}$ s$^{-1}$ which lay well in the power-law breakdown regime.

- At high $T$ and low $\dot{\varepsilon}$, $Q$ is close to that of grain boundary sliding but the coherent precipitates at the grain boundary interfaces pin effectively their motion and as a result the maximum elongation of both alloys is very low.
References

[26] Langdon TG. Metal Sc 1982;16:175.


