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Kazantzis, A. V.; Chen, Z. G.; De Hosson, J. Th M.

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Deformation mechanism of aluminum–magnesium alloys at elevated temperatures

A. V. Kazantzis · Z. G. Chen · J. Th. M. De Hosson

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Abstract The study concentrates on the formulation of a reliable constitutive equation for plastic forming of Al–Mg-based alloys above 400 °C and at strain rates above 10^{-3} s^{-1} . The deformation mechanisms of two coarse-grained Al–Mg alloys, also known as AA5182, with grain sizes 21 and 37 μm were investigated. They exhibited optimum extension at 10^{-2} s^{-1} and at T equal to 425 °C and above 475 °C, respectively, with uniform elongation above 300 %. The strain-rate sensitivity and the stress exponent were equal to 0.25 and 4, respectively, suggesting that the deformation is controlled by the solute drag of gliding dislocations whereas dislocation climb occurs also in grains whose orientation renders them hard. Grain boundary sliding may contribute to a small extent in the deformation process. The threshold stress was found to be small and the activation energy lies between 144 and 136 kJ mol^{-1} , i.e., that of Al self-diffusion and Mg diffusion in Al. It is concluded that coarse-grained materials may well fulfill the industrial requirements of forming and within this scope, the use of the low purity coarse-grained Al–Mg-based alloys of the AA5182 type would constitute the next step in the course for further cost reduction.

Introduction

The term superplasticity, in the strictest sense, refers to materials which exhibit elongation in excess of 500 % without necking and fail due to cavitation coalescence.

Their grain size is often much smaller than 10 μm and they are customarily deformed at temperatures, T , close to 0.9 T_m (where T_m is their melting point in K) and at strain rates, $\dot{\epsilon}$, much lower than 10^{-3} s^{-1} . Their main deformation mechanism is based on diffusion controlled grain boundary sliding (GBS) [1], i.e., the sliding of adjacent grains with respect to each other in response to an external stress and can be accomplished with, or without significant grain elongation [2]. This mechanism is very $\dot{\epsilon}$ sensitive, demonstrating sensitivities, a strain-rate sensitivity m higher than 0.5.

Aluminum-based materials that rely solely on GBS for their superplastic properties have relatively high purity. In order to acquire their fine-grain size, they require significant thermomechanical processing and therefore are quite costly. Furthermore, the low $\dot{\epsilon}$ results in long forming times, and consequently, their application in the automotive industry is fairly limited. By promoting GBS, a dramatic decrease of the forming time may be obtained, that is associated with $\dot{\epsilon}$ approximately equal to 10^{-1} s^{-1} and a maximum elongation to failure above 550 % [3]. This 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) [4]. These processes, however, are currently not capable of producing low cost material on an industrial scale.

The automotive industry requires alloys with a uniform maximum 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” [5, 6]. Since the present article is

A. V. Kazantzis · Z. G. Chen · J. Th. M. De Hosson (✉)
Department of Applied Physics, Materials Innovation Institute,
University of Groningen, Nijenborgh 4, 9747 AG Groningen,
The Netherlands
e-mail: j.t.m.de.hosson@rug.nl

concerned with this type of materials, the term “coarse-grained superplasticity” will be maintained.

For volume component production, the most widely used alloy was the high purity fine-grained AA5083 (with grain size $\leq 10 \mu\text{m}$). The forming times corresponded to $\dot{\epsilon}$ approximately equal to 10^{-3} s^{-1} and the operating T was lower than 500°C . At these conditions, the material deforms not only by GBS but also by a second process, which is based on viscous glide or solute drag creep (SDC) of dislocations. This mechanism is grain size insensitive and demonstrates $m \approx 0.3$. Indeed, a recently developed constitutive model that incorporates both mechanisms seems to predict accurately the results obtained by gas-pressure bulge tests [7]. A further decrease in the cost of the primary material was possible with the use of coarse-grained AA5083, i.e., it does not require extensive thermomechanical treatment for microstructure modification. These alloys with an initial grain size of $70 \mu\text{m}$, exhibited a maximum elongation to failure in excess of 300 % at 440°C and at 10^{-2} s^{-1} . The operation of SDC 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 $43 \mu\text{m}$), and thus, an enhancement of plasticity [8]. Even though viscous glide and climb of dislocations occurred simultaneously, the former was rate controlling, as the dislocation density was three orders of magnitude larger than the amount necessary for the deformation to be accomplished by dislocation moving faster than the solute diffusion rate. Consequently, coarse-grained materials may well fulfill the industrial requirements and within this scope, the use of the low purity coarse-grained Al–Mg-based alloys of the AA5182 type would constitute the next step in the course for further cost reduction. This is because lower Mn and Cr alloying makes AA5182 cheaper, whereas the higher Mn and Cr used for the grain size stability in AA5083 makes the latter stronger and more difficult to process (requiring higher rolling forces) and thus it is more expensive.

In this study, we present a full account of the mechanical properties of two coarse-grained commercial AA5182 aluminum alloys deformed in uniaxial extension at elevated temperatures. The stress exponent, grain size exponent, and activation energy were calculated in order to clarify the deformation mechanism and a constitutive equation for AA5182 is finally proposed.

Experimental procedure

The materials used in this study were 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 $21 \mu\text{m}$ ($20.7 \mu\text{m}$) and $37 \mu\text{m}$ ($37.1 \mu\text{m}$). 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 21 and the 37G material, respectively. In order to determine the conditions for maximum elongation prior to failure, tensile tests were performed, first, at constant crosshead speed (CCS) at temperatures between 350 and 500°C and at initial $\dot{\epsilon}$ between 10^{-1} and 10^{-3} s^{-1} . Subsequently, tests measuring the maximum elongation were carried out at constant true $\dot{\epsilon}$ (TSR) between 10^{-1} and 10^{-3} s^{-1} and at T , between 400 and 450°C for the material 21G and between 425 and 500°C for the material 37G.

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 min and the test commenced after a very short stabilization period of about 90 s.

The short heating cycle was chosen to approximate 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 $\pm 3^\circ\text{C}$. 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°C s^{-1} . 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 30 nm . 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 results from the orientation imaging microscopy (OIM) were collected and analyzed using the TSL application developed by AMETEC Inc.–EDAX. The step size was selected to be $1 \mu\text{m}$, as a compromise between examining appreciably large areas covering very often the entire thickness of the deformed gage and being able to collect more than 40 data points from the majority of the grains. Electron dispersive X-ray analysis (EDS) was carried out on an XL-30 Environmental FEGSEM. Subsequently, characteristic areas of the gage were punched, ground to a

thickness of 60 μm , 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.

Results

Figure 1 shows inverse pole figures (Auto-IPF) of the as-received 21G (a) and 37G (b) material, demonstrating a weak recrystallized texture, comprising equiaxed grains. Some small non-indexed areas, not more than a few pixels, can be discerned at the top and right of Fig. 1a and middle and bottom of Fig. 1b 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, Fe, and Si and depleted in Al and Mg. Their size most often was between 0.5 and 1.5 μm , but could in

rare occasions reach 10 μm [9]. Size distribution and density of precipitates was done from TEM pictures. The size distribution was fairly monodisperse around 1 μm , and there were maximum of five precipitates per grain.

Representative true stress–true strain curves are shown in Fig. 2. Their shape was completely reproducible up to ϵ where local necking developed prior to the specimen's final fracture. Prior to this ϵ , the variations of the true stress initially were small, i.e., ± 2 MPa but they increased with increasing ϵ , reaching ± 4 MPa just prior to the development of local necking. The curves at and above 10^{-2} s^{-1} showed a yield drop which was more pronounced for the material 21G, indicating that they are both essentially solid solution alloys and the stress required for initial plastic deformation is higher than that required for flow. That is usually, explained in terms of dislocations, at the onset of plasticity, becoming mobile after escaping some of their impurity atmosphere [10].

The influence of $\dot{\epsilon}$ on the flow stress, σ was determined at a fixed ϵ of 0.1, in order to avoid any influence from changes in the microstructure (i.e., grain growth and/or refinement). This is plotted at Fig. 3a and c for the material 21 and 37G, respectively. The flow stress decreased with increasing temperature and with decreasing strain rate. The strain-rate sensitivity, m , was determined by the slope in double logarithmic plots of the flow stress vs. $\dot{\epsilon}$ and varied between 0.24 ± 0.3 and 0.26 ± 0.3 . Hence, the stress exponent $n = 1/m$ was approximately equal to four and agree well with previous reported values and has been explained as indicative of deformation accomplished by SDC [11–13]. For that type of deformation the typical values of n are equal to three and thus it can be argued that AA5182, at these conditions deforms, by a combination of SDC ($m = 0.33$) and five power-law dislocation glide creep, with $m = 0.2$, i.e., the material behaves as a solid solution alloy between those identified as class I and class II [14, 15].

The onset of local necking could not be predicted for each specimen despite repeated tests (more than 10) at exactly the same conditions. As a result, the maximum elongation to failure showed significantly large error margins. The elongation to failure as a function of $\dot{\epsilon}$ is presented in Fig. 3b and d for the materials 21 and 37G, respectively, and demonstrates both $\dot{\epsilon}$ and T dependence. The largest elongations for both alloys were observed always at 10^{-2} . For the material 21G the largest elongation was observed at 425 $^{\circ}\text{C}$ with an average value of 360 %, even though elongations as high as 495 % were observed (in two specimens). For the alloy 37G the largest elongation was observed at 500 $^{\circ}\text{C}$, with an average value of 350 % whereas one specimen reached 410 %. A fractured specimen from each one of the two alloys is shown in Fig. 4. Figure 4a and b shows one specimen of the material

Fig. 1 Automatic inverse pole figures, obtained by EBSD from the as-received AA5182; (a) material 21G and (b) material 37G

Fig. 2 True stress–true strain curves at the temperatures where the alloy exhibits the maximum elongation to failure, i.e., at 425 °C for the material 21G, (a) and at 500 °C for the material 37G (b)

Fig. 3 The variation of the flow stress (a) and (c) and elongation to failure (b) and (d) for the materials 21 and 37G, respectively

Fig. 4 **a** Undeformed specimen from the material 21G and **b** after deformation at 425 °C and at 10^{-2} s^{-1} , showing an elongation of 494 %; **c** undeformed specimen from the material 37G and **d** after deformation at 500 °C and at 10^{-2} s^{-1} showing an elongation of 410 %

21G before and after deformation showing the largest elongation observed. Both sides showed secondary mechanical instabilities, which were always present at specimens with the largest elongations. Figure 4c and d shows similarly one specimen before and after deformation from the alloy 37G that reached 410 % elongation. All specimens close to the average values showed uniform elongations without secondary instabilities. The mechanism of failure was dynamic recrystallization upon necking and failure as in 5083. For the failure mechanism reference is made to our previous study on 5083 [8]. No grain boundary cavitation was observed.

EDS in the matrix of the deformed specimens showed that it contained $5.38 \pm 0.33 \text{ at.}\%$ Mg, $0.12 \pm 0.10 \text{ at.}\%$ Mn, $0.08 \pm 0.05 \text{ at.}\%$ Cu, $0.04 \pm 0.04 \text{ at.}\%$ Fe, and $0.01 \pm 0.01 \text{ at.}\%$ Cr and the balance in Al. Figure 5a–d shows characteristic microstructures from specimens, deformed at 10^{-2} s^{-1} and at 425 and 500 °C for the materials 21 and 37G, respectively. Figure 5a shows a subgrain boundary (note that the thickness contrast changes only marginally from the left to the right of the image), and a dislocation density approximately equal to $(5.7 \pm 1.0) \cdot 10^{12} \text{ m}^{-2}$. Figure 5b shows a region with an exceptionally uniform dislocation distribution with density $(2.5 \pm 0.5) \cdot 10^{13} \text{ m}^{-2}$. Most precipitates in the grain interior did not produce dislocation entanglements (i.e., they were successfully cut by dislocations) whereas some successfully pinned the motion of the HAGBs (Fig. 5c). Close to this HAGB, the dislocation density increased marginally to $(2.7 \pm 0.5) \cdot 10^{13} \text{ m}^{-2}$. Most grain microstructure showed uniform dislocation distributions such as that seen in Fig. 5d. The subgrains observed were relatively rare (<10

in an electro-transparent microstructure of several hundred μm^2 and their size varied from $<0.5 \mu\text{m}$ up to tens of μm in size. Hence based on the TEM observations, it was not possible to correlate the subgrain size with the applied stress. Each grain showed between 5 and 10 coherent precipitates (i.e., no size and strain field change, was observed when they were tilted between $\pm 5^\circ$), with diameter ≈ 0.5 and $3.0 \mu\text{m}$.

In summary, slip and climb occur simultaneously. All the precipitates observed were coherent, and some showed successful pinning of the HAGB motion. The deformation did produce a “core and mantle” mantle grain microstructure, as the subgrain boundaries observed could be small, as well as extensive enough to cut the grains in half. Nearly all grains showed extensive and usually uniform intergranular dislocation motion despite the presence of precipitates, suggesting that viscous glide, or SDC of dislocations is the rate controlling mechanism of the deformation.

Discussion

Mechanical properties

According to the phenomenological equation for creep [16], the strain rate, $\dot{\epsilon}$ at which the alloys deform at elevated temperatures, is given by:

$$\dot{\epsilon} = A \left(\frac{b}{d} \right)^p \left(\frac{\sigma - \sigma_0}{E} \right)^n \exp \left(-\frac{Q_c}{RT} \right) \quad (1)$$

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 (for Al, $b \approx 2.8631 \text{ \AA}$), d is the grain size, p is the grain-size exponent, σ is the applied stress, σ_0 is the threshold stress, E is the Young's elastic (dynamic) modulus, which was considered as that for pure Al and given by [17]:

$$E (\text{MPa}) = 77630 - 12.98 T - 0.03084 T^2 \quad (2)$$

n is the stress exponent, Q_c , is the activation energy for creep, which depends on the rate controlling process, R is the universal gas constant, and T is the absolute temperature.

The three variables of n , p , and Q_c in Eq. (1) are the most important in order to determine the deformation mechanism of the alloy. For SDC, viscous glide, or glide-controlled dislocation creep, $n = 3$, $p = 0$ and Q_c is equal to the activation energy for solute diffusion (considered equal to 136 kJ mol^{-1} , for the diffusion of Mg in Al, but it has been estimated as low as $120.5 \text{ kJ mol}^{-1}$) [16, 18–20]. For GBS, $n = 2$ and $p = 2$ when Q_c is equal to the

Fig. 5 Representative TEM microstructures from the materials 21 and 37G at the optimum deformation conditions; (a) a long SGB away from a precipitate; (b) high density of uniformly distributed

dislocations crossing a precipitate without obvious difficulty; (c) a precipitate pinning a high angle grain boundary; (d) a grain deformed by a high density of uniformly distributed dislocations

activation energy for lattice diffusion (equal to 144 kJ mol^{-1} but it may vary from 120.4 up to $144.4 \text{ kJ mol}^{-1}$, depending on the method of calculation) [8, 21], and $p = 3$ when Q_c is equal to the activation energy for grain boundary diffusion (i.e., 110 kJ mol^{-1}) [1, 12, 22, 23].

The threshold stress is taken into consideration in fine-grained superplastic materials. It is usually attributed to the segregation of solute, or impurity atoms at the grain boundaries and their interaction with the grain boundary dislocations at the onset of two grains sliding against each other [24]. Since GBS in coarse-grained alloys is limited, or very small compared to the magnitude of flow stress, the threshold stress is not considered and it is most commonly not incorporated in the constitutive equation. To confirm the validity of this assumption, the threshold stress for the present alloys was determined by plotting σ vs. $\dot{\epsilon}^m$ for the

different values of m at each temperature. These plots are presented in Fig. 6a and b for the materials 21 and 37G, respectively. It can be readily seen, that the threshold stress is not only small and independent of the temperature, but the errors calculated are such that its values can become negative (i.e., equal to zero). Consequently, σ_0 was set equal to zero for all the calculations that followed.

In order to investigate the grain size dependence on the strain rate, the flow stresses for both materials were plotted as a function of the strain rate (Fig. 7). Preliminary investigations using strain rate change tests showed exactly the same behavior [25], i.e., the flow stresses for both alloys do not coincide at 425 , as well as at 450°C . Attempts were made to produce coarser grain size materials by heat treatment, whose testing would result in obtaining more reliable results for the grain size dependence (i.e., obtaining more points in plots according to

Fig. 6 The variation of the flow stress (obtained at $\varepsilon = 0.1$) as a function of $\dot{\varepsilon}^m$ in AA5182; the m the values of the threshold stress are indicated on the diagrams

Fig. 7 The variation in the flow stress as a function of strain rate for the two different grain size materials at 425 and at 450 °C

Eq. 3, below). The grain growth, however, even after 15 min at 500 °C was so excessive and/or abnormal (i.e., grains with size varying from 50 to more than 300 μm on the same specimen) that suggested that the microstructure was metastable and there was no possibility to obtain meaningful σ vs. $\dot{\varepsilon}$ data. The values of p , however, were calculated according to the method outlined in [14], for the materials 21 and 37G. At constant temperature, Eq. 1 can be written as:

$$\ln \dot{\varepsilon} = \ln B + p \ln \left(\frac{b}{d} \right) + n \ln \left(\frac{\sigma}{E} \right) \quad (3)$$

where B is a temperature-dependent coefficient. Hence, it is possible to find p , finding the negative slope from a diagram of $\ln(\sigma/E)^n$ against $\ln(b/d)$. The values obtained were 0.25 ± 0.04 at 425 °C and 0.43 ± 0.04 at 450 °C. Even

though these can be interpreted in terms of GBS being a secondary deformation mechanism, whose contribution increases with temperature, the fact that these were calculated from only two different grain sizes does not inspire any confidence in the values for p . Many studies of coarse-grained aluminum alloys of the 5000 series, as well as AA5182 with similar grain sizes have concluded that $p = 0$ [5, 13]. Among those experiments, however, those conducted at 450 °C, were carried out only at 10^{-1} s^{-1} and showed elongations that did not exceed 100 %, whereas those conducted at 400 °C and at $\dot{\varepsilon} \approx 10^{-2} \text{ s}^{-1}$ showed maximum elongations between 250 and 300 % as the 21G material in this study (Fig. 3b) [11]. Consequently, we will consider that $p = 0$, but it will be necessary to conduct tests in AA5182 with at least four different grain sizes in the region between 15 and 30 μm , at temperatures above 400 °C and $\dot{\varepsilon}$ lower than 10^{-1} in order to investigate the contribution, if any, of GBS in the deformation mechanism.

In order to estimate the Q_c of the deformation mechanism, the data of flow stress vs. strain rate presented in Fig. 3a, c were converted into diagrams of strain rate vs. σ/E where E is the Young's elastic (dynamic) modulus calculated for each temperature according to Eq. (2). The vertical separation between data at different temperatures for the same alloy (i.e., at the same grain size) and at constant σ/E , is proportional to the activation energy, which is given by [12, 26, 27]:

$$Q_c = -R \left. \frac{\partial \ln \dot{\varepsilon}}{\partial 1/T} \right|_{\sigma/E} \quad (4)$$

The slopes of the linear regression of $\ln \dot{\varepsilon}$ with the reciprocal of the absolute temperature multiplied by $(-R)$ and are presented in Fig. 8a, b for the materials 21 and 37G, respectively. For the former, at high strain rates Q_c

shows the smaller value of $129.5 \pm 12.5 \text{ kJ mol}^{-1}$ which increases to $137.1 \pm 14.4 \text{ kJ mol}^{-1}$ at intermediate strain rates and at $147.9 \pm 17.2 \text{ kJ mol}^{-1}$ at low strain rates. Despite the large error margins (due to data in only three temperatures) the values at high and intermediate strain rates lie quite close to that of Mg diffusion in Al, i.e., 136 kJ mol^{-1} indicating that at these strain rates SDC is the dominant deformation mechanism. At low strain rates, however, the activation energy is slightly higher than that of Al self-diffusion (144 kJ mol^{-1}) suggesting that at these conditions the deformation is more likely to be accomplished also by the glide and/or climb of dislocations according to five power-law creep. The opposite trend is observed for the material with the higher grain size (i.e., Q_c decreases with decreasing strain rates). Q_c varies between $142.4 \pm 2.6 \text{ kJ mol}^{-1}$ at high and $139.7 \pm 5.8 \text{ kJ mol}^{-1}$ at the lower strain rates indicating that a combination of five power-law dislocation creep and SDC is the deformation mechanism with the former having a contribution larger than that in the 21G material, especially with increasing strain rate.

In order to compare the present experimental data with other authors having worked in coarse-grained AA5182, Fig. 9 shows the Zener-Hollomon parameter $Z = \dot{\epsilon} \exp(Q_c/RT)$ plotted against σ/E on dual-logarithmic scales. The results of the present work (blue symbols) are compared with the data by Taleff et al. [13] and Chang et al. [11]. The main difference arises from the fact that the latter authors used $Q_c = 136 \text{ kJ mol}^{-1}$, i.e., that of Mg diffusion in Al assuming that only one deformation mechanism is present, i.e., SDC. In the present work, $Q_c = 139.6 \pm 9.4 \text{ kJ mol}^{-1}$ (i.e., the average of the values presented in Fig. 7) indicating the contribution of glide/climb according to five

power-law dislocation creep. A good agreement is observed with that published previously; a marginal difference in the stress exponent does not need further mention and the restriction of the published data below the regime of power-law breakdown, i.e., below $Z \approx 5 \times 10^8$ showed similar proportionality constants. The main difference is that a Garofalo type curve (black) was used to fit all the already published results [16], whereas in the present work, even though $Z > 5 \times 10^8$ for some of the data, no deviation from the linearity was detected. From Fig. 9, the constitutive equation for the alloys examined in this study is given as:

Fig. 9 The Zener-Hollomon parameter assuming activation energy of $139.6 \pm 9.4 \text{ kJ mol}^{-1}$ as a function of modulus compensated flow stress. Data for AA5182 investigated by other authors are also included

Fig. 8 The variation of the strain rate $\dot{\epsilon}$, as a function of the reciprocal absolute temperature at the modulus compensated flow stress values indicated for the material 21G, (a) and 37G, (b), respectively

$$\dot{\epsilon} = 9.5 \cdot 10^{20} \left(\frac{\sigma}{E} \right)^4 \exp \left(- \frac{140. \text{ kJ} \cdot \text{mol}^{-1}}{RT} \right) \quad (5)$$

with the proportionality constant having units s^{-1} and corresponds to temperatures between 400 and 500 °C and strain rates between 10^{-3} and 10^{-1} s^{-1} for AA5182 with grain sizes 21 and 37 μm .

Deformation response

Due to the presence of subgrain boundaries in the microstructure, the influence of the activation energy from that of the Al self-diffusion and the proximity of the strain-rate sensitivity to that of the five power-law, one may consider that dislocation climb and not viscous glide is the rate controlling mechanism of the deformation. If that was the case, i.e., the deformation occurred in the high stress creep region then the subgrain size should be decreasing with increased stress [16]. No correlation, however, was possible in the observed microstructure since their size varied from <0.5 μm up to tens of μm in the same specimen. Under control of dislocation climb, furthermore, the elongation prior to failure rarely exceeds 150 %. Therefore, it must be concluded that dislocation climb occurs only to grains oriented so, as to present limited slip systems for dislocation glide (i.e., they appear to be quite hard at some time during plastic deformation). As a result, a higher proportion of the applied load is exerted on their boundaries and their deformation crosses over to the climb-controlled deformation. As soon as their orientation changes and sufficient slip systems activate, their deformation proceeds by viscous glide of dislocations according to SDC.

The ambiguity between dislocation glide and climb has been discussed in studies of Al–Mg–Cu/Mn alloys whose deformation yielded activation energy equal to 153 kJ mol^{-1} . It was concluded that despite solute drag of gliding dislocation being the principal deformation mechanism, a combined mechanism with dislocation climb could not be overruled. Similar calculations would help solidify our perception regarding the supremacy the solute drag glide of dislocation over climb. The average (critical) velocity above which moving dislocations move free of solute atmospheres, $\bar{v}_c$ is given by [8]:

$$\bar{v}_c = \frac{3 + 2\sqrt{2}}{2} \frac{A D_s}{b^2 k_B T} \quad (6)$$

where A is the elastic energy interaction equal to $3\mu b\Omega\epsilon_a/\pi$. μ is the temperature-dependent shear modulus, b is the magnitude of the Burgers vector, and Ω is the atomic volume of a particular solute in Al. ϵ_a is the absolute value of the misfit strain, i.e., the absolute value of the lattice constant change after replacing an Al with a solute atom, or the solute–solvent

size difference [15] and D_s is the diffusivity of the solute. In these calculations, the volumes of all solutes in fcc Al and the absolute value of the linear size factor (lsf) were used for Ω and ϵ_a , respectively [28]. More details of the physical background behind Eq.(6) can be found in [29–32]. The diffusivity of each solute considered was the average taken from experimental and theoretical values. These were equal to Q $D_0(\text{Mg}) = 4.521 \times 10^{-5} \text{ m}^2 \text{ s}^{-1}$, $Q(\text{Mg}) = 125.7 \text{ kJ mol}^{-1}$, $D_0(\text{Mn}) = 3.051 \times 10^{-2} \text{ m}^2 \text{ s}^{-1}$, $Q(\text{Mn}) = 219.8 \text{ kJ mol}^{-1}$, $D_0(\text{Cu}) = 2.11 \times 10^{-4} \text{ m}^2 \text{ s}^{-1}$, $Q(\text{Cu}) = 127.2 \text{ kJ mol}^{-1}$, $D_0(\text{Fe}) = 4.178 \times 10^{-1} \text{ m}^2 \text{ s}^{-1}$, $Q(\text{Fe}) = 213.2 \text{ kJ mol}^{-1}$, $D_0(\text{Si}) = 4.604 \times 10^{-5} \text{ m}^2 \text{ s}^{-1}$, $Q(\text{Si}) = 121.8 \text{ kJ mol}^{-1}$, $D_0(\text{Cr}) = 3.534 \times 10^{-1} \text{ m}^2 \text{ s}^{-1}$, $Q(\text{Cr}) = 251.8 \text{ kJ mol}^{-1}$ and $D_0(\text{Ti}) = 1.12 \times 10^{-1} \text{ m}^2 \text{ s}^{-1}$, $Q(\text{Ti}) = 241.4 \text{ kJ mol}^{-1}$ [20, 21]. Figure 10 shows the $\bar{v}_c$, as a function of T . 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 particular solute. In other words, the critical velocity represents an upper limit to the strain rates at which solute drag takes place. Considering Orowan's equation ($\dot{\epsilon} = \phi \cdot b \cdot \rho_m \cdot \bar{v}$, where ϕ is the Schmid factor, b is the magnitude of the Burgers vector, ρ_m is the mobile dislocation density, and $\bar{v}$ is the average dislocation velocity), for $\dot{\epsilon}$ equal to 10^{-2} s^{-1} and an average dislocation velocity $\bar{v}$ of 10^{-4} m s^{-1} , a mobile dislocation density of $7 \times 10^{11} \text{ m}^{-2}$ is required. That is approximately equal to one or two orders of magnitude lower than the total dislocation density observed. Since Mg concentrations more than 2 at.% result in supersaturation of the dislocation cores, it can be concluded that the solute drag of mobile dislocations in AA5182 by Mg solutes dominates [33]. 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 since their

Fig. 10 Critical dislocation velocity for the formation of solute atmospheres in AA5182. The gray shaded area corresponds to the T regime investigated in the present study

content is well below their solubility limit in Al. These atoms are approximately 46, 38, 37, 16, and 57 % smaller than the Al atoms, respectively, 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.

In summary, solute drag by Mg atoms dominates at $\dot{\epsilon} = 10^{-2}$. Cu and Si atoms present in the matrix may also 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 may occur concurrently with the former 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 higher dislocation densities that are expected for the tests at 10^{-1} s^{-1} , most likely result in complex interactions/configurations and may well prove to result in an increase in the driving force for dynamic recrystallization (i.e., the reason for the lower maximum elongations shown at this strain rate). Finally, at 10^{-3} s^{-1} the coarsening of the metastable microstructure most likely produces large grains where dislocations with lower average velocity are rapidly absorbed at the grain boundaries, but due to the presence of coherent precipitates GBS is inhibited and thus, failure occurs rapidly with a very low maximum elongation.

Conclusions

- A constitutive equation was proposed for tests above 400 °C and at strain rates higher than 10^{-3} s^{-1} of the coarse-grained Al–Mg-based alloys (AA5182) with grain sizes of 21 and 37 μm . The material exhibited optimum deformation conditions at $\dot{\epsilon}$ 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 %.
- The strain-rate sensitivity and the stress exponent were equal to 0.25 and 4, respectively. It suggests that the deformation mechanism is controlled by the solute drag of gliding of dislocations but from the observation that few grains show subgrain boundary formation it is concluded that dislocation climb according to the five power-law can also take place.
- In the material investigated the threshold stress is small with respect to the applied stress.
- It is concluded that GBS contributes only to a small extent to the deformation at temperatures above 400 °C and at strain rates equal or lower than 10^{-2} s^{-1} .
- The activation energy lies between 144 and 136 kJ mol^{-1} , i.e., that of Al self-diffusion and Mg diffusion in Al.

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