DEFORMATION MECHANISMS IN Mg-3Al-1Zn ALLOY

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Abstract

The influence of temperature on the deformation behavior of Mg-3Al-1Zn alloy has been investigated over a wide temperature range from room temperature to 300 °C. Stress relaxation tests were used to estimate the activation volume and the components of the applied stress: internal stress and thermal stress. The value of the internal stress at a testing temperature increases with increasing strain. The internal stress estimated at a given strain decreases with increasing temperature. The activation volume decreases with increasing thermal stress independently of testing temperature.

Keywords: Magnesium alloy; tension testing; stress relaxation; thermally activated process

1. INTRODUCTION

The deformation behavior of Mg-3Al-1Zn (AZ31) alloy samples prepared using different techniques was investigated over a wide temperature range [1-3]. The strain to failure in tension at room temperature is much lower than that in some Al alloys. This can be ascribed to the fact that the von Mises criterion requiring five independent slip modes cannot be met. Slip in Mg (and its alloys) occurs on the basal plane at room temperature (the lowest critical resolved shear stress). Improvements in ductility are achieved at higher temperatures and/or with finer grain sizes. The improvement is owing to the activity of non-basal slip systems and/or deformation twinning. The second-order pyramidal slip systems, \{0211\}2311, that have dislocations with the Burgers vector <c+a>, are considered as significant. The critical resolved shear stress (CRSS) for non-basal slip at room temperature is much higher than that for basal slip. However, it decreases very rapidly with increasing temperature.

At a certain temperature, the dislocations move over local obstacles with the aid of thermal activation. The plastic strain rate is then

\[ \dot{\varepsilon} = \dot{\varepsilon}_0 \exp \left( -\frac{\Delta G}{kT} \right) \]

where \( k \) is the Boltzmann constant, \( T \) is the absolute temperature, \( \dot{\varepsilon}_0 \) is a pre-exponential factor and the change in the Gibbs free energy depends on the effective (thermal) stress \( \sigma^* \) as \( \Delta G = \Delta G_0 - V\sigma^* \). Here \( \Delta G_0 \) is the Gibbs free energy necessary for overcoming a short-range obstacle in the absence of the applied effective stress and \( V = b d L \) is the activation volume where \( b \) is the Burgers vector, \( d \) is the obstacle width and \( L \) is the mean length of dislocation segments between obstacle. It should be mentioned that the applied stress \( \sigma \) may be divided into two components \( \sigma = \sigma_i + \sigma^* \), where \( \sigma_i \) is an athermal (internal) stress [4]. The stress relaxation technique has been demonstrated to be a useful method for estimating the activation volume and the both components of the applied stress. In a stress relaxation (SR) test, the specimen is deformed to a certain stress (strain), the testing machine is stopped and the stress acting on the specimen is allowed to relax. The specimen can be again reloaded to a higher stress (strain) and the SR test can be
repeated [5]. The stress decreases with time $t$. The stress decrease with time during the SR test can be described by the following equation [6]

$$\Delta \sigma(t) = \sigma(0) - \sigma(t) = \alpha \ln(\beta t + 1)$$  \hspace{1cm} (2)

Here $\sigma(0) \equiv \sigma_0$ is the stress at the beginning of the SR (at time $t = 0$, $\beta$ is a constant and $\alpha = kT/V$.

Components of the applied stress can be estimated using the method of Li [7, 8]. The stress relaxation curve should be fitted to the power law function in the following form

$$\sigma - \sigma_i = \left[ a(m - 1) \right] t^{-m} \left( t + t_0 \right)^{-\frac{1}{m}}$$  \hspace{1cm} (3)

Here $a$, $t_0$ and $m$ are fitting parameters. This relation was derived based on dislocation dynamics (a power law relation between dislocation velocity and stress) assuming both the internal stress and the density of mobile dislocations are constant during the SR.

The aim of this work is to use the SR tests in squeeze cast AZ31 alloys in order to obtain the values of the activation volume and both stress components and to estimate the deformation mechanism. The results will be compared with those for AZ31 samples processed by different techniques.

2. EXPERIMENTAL PROCEDURE

Commercial alloy AZ31 (Mg-3Al-1Zn) prepared by squeeze cast under a protective gas atmosphere was used in this study. Samples for tensile tests having a cylindrical form with a diameter of 5 mm and a gauge length of 25 mm were deformed using an INSTRON machine at a constant cross head speed giving an initial strain rate of $3 \times 10^{-5}$ s$^{-1}$ over a wide temperature range of 23 to 300 °C. Stress relaxation tests were performed at different temperatures along the stress-strain curve at a given temperature. Duration of the SR was 300 s. The activation volume was estimated using Eq. (2).

3. RESULTS AND DISCUSSION

The true stress-true strain curves of AZ31 alloy deformed at different temperatures are shown in Fig.1. It can be seen that temperature has a significant influence on the course of the stress-strain curve. The values of the yield strength, $\sigma_{0.2}$, (estimated as the flow stress at 0.2 of strain) and tensile strength, maximum stress, $\sigma_{\text{max}}$, (estimated as the maximum value of the flow stress) decrease with increasing temperature. The yield strength decreases slowly with increasing temperature, from about 57 MPa to about 34 MPa at 300 °C. The values of the yield strength for as-cast AZ31 alloy are very low, whereas the yield strength of AZ31 processed by hot rolling are much higher; $\sigma_{0.2}$ decreases from about 220 MPa at room temperature to about 30 MPa at 300 °C [1]. AZ31 samples deformed at room temperature after equal channel angular pressing (ECAP) exhibit the yield strength about 140 MPa. It is obvious that the yield strength depends on the processing technique used. Samples prepared by different processing techniques may exhibit different texture. The initial texture influences the value of the yield strength. It should be mention
that the value of the yield strength depends not only on the testing temperature but also on the grain size (that is influenced by the processing technique) and on the strain rate. Rolling and/or ECAP technique cause an increase in the dislocation density, which increases the yield strength. It is obvious (Fig. 1) that the maximum stress decreases rapidly with increasing test temperature between 100 and 300 °C. The difference between the maximum stress and the yield stress decreases with increasing temperature. The strain hardening also decreases. At a temperature of 300 °C the strain hardening rate is close to zero. This indicates a dynamic balance between hardening and recovery; a recovery process leads to softening at higher temperatures.

A part of the true stress-true strain curve for AZ31 alloy sample deformed at room temperature is shown in Fig. 2. Points indicate the stresses at which subsequent SR tests were carried out. Fig. 2 also shows both components of the flow stress, the internal stress, \( \sigma_i \), and the thermal stress, \( \sigma^* \), as a function of strain. Analogous variation of the flow stress, internal stress and thermal stress with strain estimated at 300 °C are shown in Fig. 3. It can be seen that the internal stress form a substantial contribution to the applied stress at room temperature. On the other hand, the internal stress is much lower than the effective stress at 300 °C. The applied stress decreases with increasing testing temperature. This indicates a recovery process. It is known that the internal stress depends on the dislocation density, i.e. \( \sigma_i = A\rho^{1/2} \) (\( \rho \) is the density of the stored dislocations and \( A \) is a constant). The observed decrease in the internal stress indicates a decrease in the dislocation density. The dislocations stored at obstacles contribute to hardening. The moving dislocations can cross slip and after cross slip may meet the stored dislocations and annihilate. The dislocation density decreases, hardening decreases and softening increases. At higher temperatures, dislocations may also climb. A rearrangement and/or annihilation of dislocations may happen owing to climb. The activity of cross slip and climb increases with increasing temperature. It means the stored dislocation density decreases with increasing temperature and therefore the internal stress (and the applied stress) has to decrease, which is observed in experiment. Similar variations of the internal as well as effective stress with temperature were also observed in some Mg-Al-Ca, Mg-Al-Sr and Mg-Al-Nd alloys [9, 10].

The apparent activation volume estimated from the stress relaxation depends on the flow stress and testing temperature. Usually the values of the activation volume are given in \( b^3 \). A thorough analysis has proved that the apparent activation volume is a function of the effective (thermal) stress independently of the testing temperature as shown in Fig. 4. All values of \( V/b^3 \) lie at the curve given by the following equation [11]
Here $p$ and $q$ are parameters reflecting the shape of a resistance profile. Their values are limited by the conditions: $0 < p \leq 1$ and $1 \leq q \leq 2$; it is very often used $p = 1/2$ and $q = 3/2$. The apparent activation volume has values of $290 \, \text{b}^3$ and $20 \, \text{b}^3$ for the thermal stresses equal 5 and 58 MPa, respectively as it is obvious from Fig. 4. The activation volume at a fixed temperature decreases with strain. The values of the activation volume and the activation enthalpy may help to identify a thermally activated process controlling dislocation motion. The double cross slip may be the thermally activated process.

Considering the shape of the true stress-true strain curves observed below 150 °C, one can conclude that strain hardening is caused by multiplication and storage of dislocations. At 200 and 300 °C, there is not only storage of dislocation during straining leading to hardening but also annihilation of dislocations leading to softening and at 300 °C a dynamic balance takes place between hardening and softening. From the dislocation theory point of view, this deformation behavior depending on temperature may be explained assuming changes in deformation modes (changes in deformation mechanisms). The primary (main) dislocation slip system in Mg and in its alloys is the basal slip system, i.e. motion of $\langle a \rangle$ dislocations occurs in the basal plane. Thus, the von Mises requirement for five independent deformation modes to ensure a reasonable deformability of polycrystalline magnesium alloys is not fulfilled. This is reason why the strain to failure in tension at room temperature is lower than that seen, for instance, in Al alloys. Twinning may play an important role, especially in wrought Mg alloys [12]. The number of independent slip systems in the basal plane is only two. The activity of non-basal slip systems is therefore required and ductility can be improved. The critical resolved shear stress (CRSS) for non-basal slip systems at room temperature is higher by about a factor 100 than the CRSS for the basal slip. From the activity of non-basal slip systems, motion of dislocations of the Burgers vector $\langle c+a \rangle$ in the first- and second-order pyramidal slip planes is expected. The von Mises criterion is then fulfilled. The CRSS for the second-order pyramidal slip systems decreases very rapidly with increasing temperature. It is interesting to note that Mátis et al. [13], who studied the evolution of non-basal dislocation with temperature in Mg by X-ray diffraction, found that at higher temperatures the fraction of $\langle c+a \rangle$ dislocations increases at a cost of $\langle a \rangle$ dislocations. The total dislocation density decreases with increasing temperature. Multiplication and storage of $\langle c+a \rangle$ dislocations may contribute to hardening. Different reactions between the mobile $\langle a \rangle$ basal dislocations and $\langle c+a \rangle$ pyramidal can occur [14, 15]. Within the basal plane, immobile, sessile $\langle c \rangle$ and/or $\langle c+a \rangle$, dislocations can arise. They are obstacle for moving dislocations. Production of sessile dislocations may increase the density of forest dislocations and hence an increase in hardening can result. On the other hand, screw components of $\langle c+a \rangle$ dislocations may move to the parallel slip planes by double cross slip and they can annihilate. This leads to softening. The activity of non-basal slip system with $\langle c+a \rangle$ dislocations – the glide and cross slip – increases with

$$V = \frac{\Delta G_0 p q}{\sigma_{0}^*} \left[ 1 - \left( \frac{\sigma_{0}^*}{\sigma_{0}^*} \right)^p \right]^{q-1} \left( \frac{\sigma_{0}^*}{\sigma_{0}^*} \right)^{q-1}$$ (4)
increasing temperature. The scenario described above can help in understanding of the deformation behavior of AZ31 alloy over a wide temperature range.

4. CONCLUSIONS

Testing temperature influences the mechanical properties and deformation behavior of cast AZ31 Mg alloy. A combination of strain hardening and strain softening occurs at temperatures above 200 °C. Stress relaxation tests have been used to reveal the activation volume, internal and thermal stresses and the nature of thermally activated mechanisms.

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REFERENCES