A crystal plasticity study on the deformation of an AZ31 alloy sheet under elevated temperature

Zihan Li¹, Guowei Zhou¹, Dayong Li¹,² and Yinghong Peng¹
¹. State Key Laboratory of Mechanical Systems and Vibration, Shanghai Jiao Tong University, Shanghai 200240, China

* E-mail: dyli@sjtu.edu.cn

Abstract. In order to investigate the evolution of deformation mechanisms of AZ31B Mg alloy sheet and their correlation with material property development under intermediate temperatures, systematic experimental examination and the crystal plasticity analysis are performed. The mechanical responses of AZ31B Mg alloy sheet are measured under uniaxial tension and compression along rolling direction, transverse direction and normal direction directions and over the temperature range 100-300°C. A series of EBSD experiments is also carried out in order to explore nucleation features in the dynamic recrystallization (DRX) of AZ31 Mg alloy sheets. Furthermore, a phenomenological DRX criterion is introduced into the visco-plastic self-consistent model through a dislocation density based hardening model to analyze the deformation of AZ31 Mg alloy sheets at 200°C.

1. Introduction
Magnesium alloy sheet as a light-weight material offers a high strength to weight ratio that has drawn much attention in recent years. However, its poor formability and high anisotropy due to limited deformation modes at room temperature and strong initial basal texture limit its widespread structural application [1]. To explore the possibility of improving formability, extensive researches have been done on the deformation mechanisms of Mg alloys [1,8,9].

Under elevated temperature, the occurrence of dynamic recrystallization (DRX) brings another challenge, which causes not only softening behavior but also a dramatic change in microstructure. During DRX, the driving forces for both nucleation and growth of new grains are related to the stored energies or strain histories of their mother grains, which generally are varied from grain to grain. Therefore, it is important to take the study on the deformation of an AZ31 alloy sheet under elevated temperature [7].

2. Experiment Analysis
In order to investigate the evolution of deformation mechanisms of AZ31B Mg alloy sheet and their correlation with material property development under intermediate temperatures, systematic experimental examination and the crystal plasticity analysis are performed. The mechanical responses of AZ31B Mg alloy sheet are measured under uniaxial tension and compression along rolling direction (RD), transverse direction (TD) and normal direction (ND) and over the temperature range 100-300°C. The uniaxial tensile flow curves under different test conditions are depicted in figure[5]. The samples oriented in TD and RD have the largest and lowest yield strength and flow stress respectively at all testing temperatures, while the sample of 45° orientation falls between them. At 200°C and 300°C, and at lower strain rate (0.001/s), the softening behavior is observed which is caused by DRX. The AZ31B
sheet has an excellent elongation (failure strain is about 0.6) at low strain rate even at 100°C. Figure 2 shows stress-strain curves in uniaxial compression. For both RD and TD compression, the yield stresses at 100°C and 150°C are almost the same, and they drop slightly at 200°C. The yield stresses of TD compression are higher than those of RD compression, and similar to the uniaxial tension results.

The evolution of r-value [1] from a strain of 0.07 to 0.2 in uniaxial compression is shown in figure 3. At lower temperature (100°C and 150°C), the r-value becomes less sensitive to strain with increasing temperature. However, at higher temperatures (200°C and 300°C) in RD compression, a rather unusual trend is observed: a large initial r-value decreases gradually with increasing strain. The r-value in TD compression also has a similar trend but with a larger magnitude. It needs to be mentioned that specimens with different height to width ratios (i.e., aspect ratios) are tested but the results show the same trend. The reason for this unusual behavior is either activation of multiple deformation modes or dynamic recrystallization. During ND compression, it seems the deformation along RD is much faster than TD, which leads to a higher r-value. In contrast to uniaxial compression and tension, the r-value remains almost constant with increasing strain during ND compression, and only in the earlier stage there is a slight decreasing trend.

Figure 1. Stress-strain curves in uniaxial tension at different temperatures and specimen orientations. [5]

Figure 2. (a) Stress-strain curves in uniaxial compression along RD/TD, (b) stress-strain curves of compression along ND. [5]

Figure 3. The average r-value evolution at different temperatures. [5]

In order to investigate the misorientation development during the DRX process, two types of grains are selected in the electron backscattered diffraction (EBSD) maps. Type A (red dots in figure 4) represents a parent grain surrounded by some small DRX grains, and in the present work those grains whose sizes are more than 2.5 times of the small grains surrounded are regarded as the parent grains; Type B represents a grain experiencing no DRX, since there are only similarly sized grains and evidently exists no small DRX grains around. Figure 4 shows the EBSD scanned results, taking RD tension as example. At each end point of the lines drawn from each typical grain to its boundary is a unit cell with its orientation presented. The parent grains experiencing DRX and surrounded by some small DRX grains, and the grains without experiencing DRX are recognized in those areas. There exist evident low
angle grain boundaries (white and blue lines) inside the grains, especially in the parent grains around with DRX grains. For each parent grain (Type A grains), the lattices at the core area (with the red dots) are remarkably closer to balance position (basal slip rotates the c-axis toward the sheet normal whereas prismatic slip rotates <10-10> direction with c-axis to the balance position) than other lattices outside the core, i.e. near the DRX grains, because the core area endures more rotation for than the grain boundary areas. More details about misorientation analysis and other loading modes will be further discussed elsewhere.

Figure 4. The EBSD scanned result for RD tension.

3. Polycrystal plasticity analysis

3.1. VPSC model with dislocation density based hardening law

The macroscopic response of the polycrystal aggregate is modeled within the visco-plastic self-consistent (VPSC) framework presented by Lebensohn and Tomé [3], in which each grain is treated as an ellipsoidal visco-plastic inclusion embedded in an effective visco-plastic medium.

Dislocation density (ρ) is thermally-controlled and updated based on dislocation generation and annihilation [2]. The dislocation density increment on a slip mode is calculated as:

$$\frac{\partial \rho^\alpha}{\partial T} = k^\alpha_1 \sqrt{\rho^\alpha} - k^\alpha_2 (\dot{\varepsilon}, T) \rho^\alpha$$

where $k^\alpha_1$ is a material constant, not varying with temperature, and $k^\alpha_2$ is a function of temperature and strain rate. The updated CRSS can be simplified to the following form for each slip mode [6]:

$$\tau^{\alpha}_s = \tau^{\alpha}_0 + 0.9b^\alpha \mu \sqrt{\rho^\alpha}, s \in \alpha$$

where $b^\alpha$ is the length of the Burgers vector associated with the slip mode $\alpha$, and $\mu$ the shear modulus.

3.2. DRX model for DRX nucleation

For a polycrystal with multiple slip system, the total dislocation density of a grain $i$ is calculated by summing up the dislocation densities on all slip modes, i.e., $\rho_i = \sum_\alpha \rho^\alpha_i$. When the dislocation density of a grain reaches the critical value $\rho^c_i$ ($\rho^c_i > \rho^c$), new DRX grains will possibly nucleate.

The expectation of DRX grains nucleation can be evaluated by

$$N = \sum_{i=1}^k C(T) \dot{\varepsilon}^{0.9} S_i \phi_i \Delta t$$

where $\phi_i$ is the rotation angle, which is the parameter of misorientation [4]; $C(T)$ is a temperature related coefficient; $\dot{\varepsilon}$ is strain rate; $S_i$ is the surface area of grain $i$; $\Delta t$ is the time increment of simulation step; $k$ is the number of grains with $\rho^c_i > \rho^c$. 
According to the work of Zhou et al. [6], the driving force for grain boundary migration is calculated based on the difference between the dislocation density of grain $i$ and average dislocation density $\bar{\rho}$ over all the grains, i.e.

$$V_i = M(T)\tau (\bar{\rho} - \rho_i)$$

(4)

$$\bar{\rho} = \sum_i \rho_i WGT(i)$$

(5)

where $V_i$ is grain boundary velocity, $M(T)$ is the grain boundary mobility, $WGT(i)$ is the volume weight of grain $i$.

3.3. Modelling of AZ31B sheet

In the current work, the experimental results on the AZ31B sheet subjected to RD tension, RD compression and ND compression at 200°C and strain rate of $1 \times 10^{-3}$ s$^{-1}$ are used for calibration of the modeling parameters. The input parameters of the polycrystal-DRX modeling are as follows. The initially average grain size is 7 µm, and initial grain number is 300; the initial texture is a strong ellipsoidal basal texture, as shown in Figure 5 where the $c$-axes of grains are slightly tilted away from the normal and toward the RD, while the intensity distribution of (100) planes is remarkably diffuse. The shear modulus $\mu$ at 200°C is 15000 MPa. Table 1 shows the material parameters and fitted parameters relative to hardening and DRX.

![Figure 5](image)

**Table 1.** The hardening and DRX parameters of AZ31B sheet in the simulation

| Temp(°C) | Mode     | $\tau_0$ (MPa) | $k_2$(m$^{-1}$) | $\rho_c$ (m$^{-2}$) | $C$(s$^{-1}$m$^2$) | $M$(m$^4$J$^{-1}$s$^{-1}$) |
|----------|----------|----------------|----------------|---------------------|-------------------|---------------------------|
| 200      | Basal    | 4.3            | 5.4e7          | 486                 | 0.9e13            | 2.8e10                    | 1e-16                     |
|          | Prismatic| 39             | 1.0e8          | 26                  |                   |                           |                           |
|          | <c+a>    | 75             | 1.5e9          | 7650                |                   |                           |                           |

3.4. Simulation analysis

Figure 6 shows the comparison between the experimental and fitted stress-strain curves, variation of $r$-value of the AZ31B sheets at 200°C. The calculated results of VPSC-DRX model agree well with the experimental results of the RD tension, RD compression and ND compression. Especially, the softening in stress-strain curves is captured due to incorporation of DRX. Figure 7 shows the calculated average grain size varying with strain under the three loading conditions, that simulation captures the experiment points. There is no evident difference in mean grain size evolution among three loading modes. The variation of grain size under the three conditions are similar.

The evolution of texture is calculated by using the established VPSC-DRX model. The comparison of pole figures at strain 0.4 between the calculated and experimental results under the three loading conditions is shown in Figure 8. In RD tension, the dominantly active basal $<a>$ and prismatic $<a>$ slips cause the suppression of basal texture in RD direction in (0002) pole figure and the appearance of six fold maxima distribution in (10-10) pole figure. In RD compression, basal $<a>$ and prismatic $<a>$ slips are the dominant deformation modes. With the activation of the basal slip system, the direction normal to the basal plane rotates towards the compressive direction. Those slips will split the basal texture in (0002) pole figure and cause two maxima distribution in (10-10) pole figure. In ND compression, the
most effective deformation mechanisms basal $<a>$ and pyramidal-II $<c+a>$ slips which rotate $c$-axis to the normal direction of sheets, will enhance the basal texture more stronger in (0002) pole figure than the initial one, and due to low activity of prismatic $<a>$ slip disperse distribution in (10-10) pole figure is similar to initial one. For these three loading modes, the simulated texture can reproduce the experimental observation for both initial parent grains and DRX grains.

**Figure 6.** The experimental and fitted curves: stress-strain curves and r-value curves of the AZ31B sheets at 200°C in different loading modes.

**Figure 7.** The evolution of average grain size with strain at 200°C.

**Figure 8.** Comparison of texture for three loading modes at 200°C and strain 0.4.
4. Conclusion

In this paper, a set of comprehensive experiments along different loading directions are performed on AZ31B Mg alloy sheets over the temperature range 100-300°C to obtain the mechanical responses. The effects of temperature and loading direction on the stress-strain curve and anisotropy are analyzed. EBSD experiments is carried out to explore misorientation development during the DRX process.

The VPSC model employs a dislocation density hardening law to account for the deformation and grain's orientation evolution, and provides the necessary information, like the grain's dislocation density, to the DRX module. The simulated texture can reproduce the experimental observation for both initial parent grains and DRX grains.

Acknowledgement

The authors would like to acknowledge the support of the National Natural Science Foundation of China (No. 51675331), Major Projects of the Ministry of Education (No. 311017), and Natural Science Foundation of Shanghai (No.15ZR1424100).

References

[1] Agnew S R and Duygulu Ö 2005 Int. J. Plasticity 21 1161-93
[2] Beyerlein I J and Tomé C N 2008 Int. J. Plasticity 24 867-95
[3] Lebensohn R A, Tomé C N and Castañeda P P 2007 Philos. Mag. 87 4287-322
[4] Tóth L S, Estrin Y, Lapovok R and Gu C 2010 Acta Mater. 58 1782-94
[5] Zhou G W, Jain M K, Wu P D, Shao Y C, Li D Y and Peng Y H 2016 Int. J. Plasticity 79 19-47.
[6] Zhou G W, Li Z, Li D Y, Peng Y H, Zurob H S and Wu P D 2017 Int. J. Plasticity 91 48-76
[7] Cram D G, Zurob H S, Brechet Y J M and Hutchinson C R 2009 Acta Mater. 57 5218-28
[8] Wang H, Wu P D, Tomé C N and Huang Y 2010 J. Mech. Phys. Solids 58 594-612
[9] Wang H, Raeisinia B, Wu P D, Agnew S R, Tomé C N, 2010 Int. J. Solids Struct. 47 2905-17