Processing maps for hot deformation of rolled AZ31 magnesium alloy plate: Anisotropy of hot workability

Y.V.R.K. Prasad, K.P. Rao*

Department of Manufacturing Engineering and Engineering Management, City University of Hong Kong, Tat Chee Avenue, Kowloon, Hong Kong SAR, China

Received 30 November 2006; received in revised form 5 October 2007; accepted 10 October 2007

Abstract

Processing maps on rolled AZ31 magnesium plate have been developed in the range 300–550 °C and 0.0003–10 s⁻¹ by hot compression of specimens parallel to the rolling direction (RD), the transverse direction (TD), or the normal direction (ND) with a view to examine whether the hot workability is anisotropic. The processing map for RD specimens exhibited a single wide domain of workability in the temperature range 350–550 °C and strain rate range 0.0003–0.3 s⁻¹ in which dynamic recrystallization (DRX) occurs. The apparent activation energy in this domain is estimated to be 143 kJ/mole, which is close to that for self-diffusion in magnesium. On the other hand, the maps for the TD and ND specimens exhibited two DRX domains, one at lower strain rates (<0.01 s⁻¹) and the other at higher strain rates (>1 s⁻¹). The apparent activation energies estimated in the lower strain rate DRX domain of TD and ND maps are 180 and 168 kJ/mole, respectively, suggesting cross-slip occurs in this domain. In the higher strain rate domain, the apparent activation energy estimated in both TD and ND specimens is 105 kJ/mole, which is close to that for grain-boundary diffusion. The results suggest that the hot deformation behavior is anisotropic and a simple correlation with the known primary rolling texture (⟨0002⟩⟨10 10⟩) revealed that both the first order and second order pyramidal slip systems are favored in the RD orientation resulting in higher workability. The other two orientations may be considered “harder” as regards hot workability since either the first order or the second order pyramidal slip is likely to dominate. In all the three orientations and within the DRX domains, the grain size varies linearly with the Zener-Hollomon parameter.

© 2007 Elsevier B.V. All rights reserved.

Keywords: Processing maps; Magnesium alloy; Hot deformation; Texture; Dynamic recrystallization; Pyramidal slip

1. Introduction

In view of the low density and high specific strength, magnesium alloys are being considered for structural components in aerospace and automobile industries [1]. Out of the several alloys being developed [2], Mg–Al–Zn system offers a wide range for both as-cast and wrought products [3] and Mg–3Al–1Zn (AZ31) is a workhorse of all the wrought Mg alloys [4]. Because of the industrial importance of AZ31, its microstructure, mechanical properties and formability are being evaluated more extensively [5–13]. Since the alloy has limited workability at room temperature, bulk forming involving extrusion, rolling or forging is done at elevated temperatures [14–21]. The hot working behavior of Mg–Al–Zn alloys has been reviewed recently [22] and the apparent activation energy estimated by different investigators [23–28] has shown some variations and is generally higher than that for self-diffusion. During mechanical processing of magnesium materials, strong crystallographic texture gets developed [29–32] which depends on the processing conditions, the geometry and state-of-stress in the deformation zone. Although the textures cannot be described by simple “ideal” orientations due to the occurrence of non-basal slip and twinning [33,34], the primary components may be identified in each case. For example, hot extrusion of a rod results in a fiber texture with c-axis perpendicular to the extrusion direction [6,9,35,36] while the predominant rolling texture is (⟨0 0 0 2⟩⟨1 0 1 0⟩) [33,37]. However, cold working intensifies the basal texture along with a strong ⟨1 1 2 0⟩ component in the primary flow direction [34,38,39]. The development of texture during hot deformation of AZ31 alloy by compression [40] and tension [41] has been studied and the role of prismatic and pyramidal slip and cross-slip in weakening basal texture has been recognized. Although the influence of texture on the mechanical properties, formability and high temperature flow stress of magnesium alloys is well documented, there are only limited reports on the anisotropic behavior during hot deformation.

* Corresponding author. Tel.: +852 27888409; fax: +852 27888423. E-mail address: mekprao@cityu.edu.hk (K.P. Rao).
AZ31 has been studied in some detail [5,6,10,12,13,35–37], its effect on the hot workability has not been evaluated. The aim of the present study is to investigate whether the hot-working behavior of a rolled AZ31 magnesium alloy plate is anisotropic. In an earlier investigation on the kinetics of hot deformation [42], texture has been found to have significant effect on the apparent activation energy. In this paper, processing maps have been developed on AZ31 specimens deformed in three different orientations, namely the rolling direction (RD), the transverse direction (TD) and along the normal to the rolling plane (ND). This technique has been used earlier to study the hot deformation mechanisms in Mg and its alloys [43–46] including dynamic recrystallization (DRX) and flow instabilities.

The principles and basis for processing maps are described earlier [47,48]. Briefly, the work-piece undergoing hot deformation is considered to be a dissipator of power and the strain rate sensitivity (m) of flow stress is the factor that partitions power between deformation heat and microstructural changes. The efficiency of power dissipation occurring through microstructural changes during deformation as a function of temperature and strain rate is given by

\[ \eta = \frac{2m}{m + 1} \]  \hspace{1cm} (1)

Eq. (1) is derived by comparing the non-linear power dissipation occurring instantaneously in the work-piece with that of a linear dissipater for which the m value is unity. The variation of efficiency of power dissipation with temperature and strain rate represents a power dissipation map which is generally viewed as an iso-efficiency contour map. Further, the extremum principles of irreversible thermodynamics as applied to continuum mechanics of large plastic flow [49] are explored to define a criterion for the onset of flow instability given by the equation for the instability parameter \( \xi(\dot{\varepsilon}) \):

\[ \xi(\dot{\varepsilon}) = \frac{\partial \ln[m/(m + 1)]}{\partial \ln \dot{\varepsilon}} + m \leq 0 \]  \hspace{1cm} (2)

The variation of the instability parameter as a function of temperature and strain rate represents an instability map which delineates regimes of instability where \( \xi \) is negative. A superimposition of the instability map on the power dissipation map gives a processing map which reveals domains (efficiency contours converging towards a peak efficiency) where individual microstructural processes occur and the limiting conditions for the regimes (bounded by a contour for \( \xi = 0 \)) of flow instability. Processing maps help in identifying temperature–strain rate windows for hot working where the intrinsic workability of the material is maximum (e.g. dynamic recrystallization (DRX) or superplasticity) and also in avoiding the regimes of flow instabilities (e.g. adiabatic shear bands or flow localization).

2. Experimental

The starting material was a 25-mm thick commercially rolled plate of AZ31B magnesium alloy (nominal composition (wt. %): Al—3%, Zn—1%, Mn—0.2%, remainder Mg). Cylindrical specimens of 10 mm diameter and 15 mm height were machined with the compression axis parallel to RD, TD, or ND as the case may be. The specimen orientations are schematically shown in Fig. 1. For inserting a thermocouple to measure the specimen temperature as well as the adiabatic temperature rise during deformation, the specimens are provided with a 0.8-mm diameter hole machined at mid-height to reach the centre of the specimen.

Uniaxial compression tests were conducted at constant true strain rates in the range 0.0003–10 s\(^{-1}\) and temperature range 300–550 °C. Details of the test set-up and procedure are described in earlier publication [50]. Constant true strain rates during the tests were achieved using an exponential decay of actuator speed in the servo-hydraulic machine. Graphite powder mixed with grease was used as the lubricant in all the experiments. The specimens were deformed up to a true strain of 0.7 and quenched in water. The load–stroke data were converted into true stress–true strain curves using standard equations. The flow stress values were corrected for the adiabatic temperature rise assuming linear relationship between logarithm of flow stress and inverse of temperature within the intervals of experimental data points. Processing maps were developed using the procedures described earlier [47] and with the flow stress data at different temperatures, strain rates and strains obtained from the above experiments as inputs. The deformed specimens were sectioned in the centre parallel to the compression axis and the cut-surface was mounted, polished and etched for metallographic examination. All the specimens were etched with an aqueous solution containing 10% picric acid. The rolling texture of the starting material was confirmed by X-ray diffractometry using Cu Kα radiation.

3. Results and analysis

3.1. Microstructure of the starting plate

The microstructures recorded on the three planes of the as-received rolled plate are shown in Fig. 2 which exhibits non-uniform grain structure with fine-grained bands seen prominently on the RD–ND plane. These bands might have been caused by flow localization which is likely to occur if finish rolling is done in a lower temperature–higher strain rate regime. The fine grains in the bands represent recrystallized grains formed during post-deformation cooling. On the
rolling plane (RD–TD plane), the average grain diameter is about 18 μm while that in the fine-grained bands is about 12 μm.

3.2. Deformation parallel to the rolling direction (RD specimens)

The true stress–true strain curves obtained on RD specimens at different strain rates and at 350 and 550 °C are shown in Fig. 3(a) and (b), respectively. These exhibited flow softening at higher strain rates and steady state flow at lower strain rates. At temperatures above 400 °C and strain rates lower than 0.1 s⁻¹, the flow curves were of steady state right from the start of plastic deformation.

The processing map obtained at a strain of 0.5 (steady state plastic flow) is shown in Fig. 4. The numbers against each contour represent the efficiency of power dissipation as percent. The thick line represents the boundary for the regime of flow instability as per the criterion given in Eq. (2). The map exhibits a single domain in the temperature range 375–550 °C and strain rate range 0.0003–0.3 s⁻¹ with a peak efficiency of about 42% occurring at 550 °C and 0.0003 s⁻¹. The map exhibits a narrow regime of instability in vicinity of 425–510 °C and at strain rates higher than about 3.0 s⁻¹.

The variation of average grain diameter measured on the specimens deformed at different temperatures across the domain at a strain rate of 0.0003 s⁻¹ is shown in Fig. 5. The grain size increases gradually up to a temperature of 450 °C followed by grain growth at higher temperatures which may have a large contribution from the thermal effect. The variation of the torsion ductility recorded by several investigators [23–26] exhibited a similar trend and as an example the data obtained by Chabbi and Lehnert [26] is shown in Fig. 5. Also, for the purpose of comparison, the variation of the efficiency of power dissipation is also plotted in this figure. Within the domain, the grain size and the hot ductility increase continuously with increasing temperature similar to that of the efficiency of power dissipation. Such variations are typical of DRX as also observed in the hot deformation of a wide range of materials [47,48]. The microstructure recorded on RD specimen deformed under conditions within this domain (450 °C/0.001 s⁻¹) is shown in Fig. 6 which exhibits
Fig. 4. Processing map obtained on RD specimens of AZ31 at a strain of 0.5. The numbers against the contours represent efficiency in percent. The thick line represents the boundary between stable and unstable regimes (marked INST). The DRX domain is the shaded region.

Fig. 5. The variation of the average grain diameter and efficiency of power dissipation with temperature at the strain rate of 0.0003 s$^{-1}$. The torsional ductility (strain to fracture) variation reported by Chabbi and Lehnert [26] is also shown.

Fig. 6. Microstructure of the RD specimens deformed at a strain rate of 0.001 s$^{-1}$ and at temperatures of 450 °C (DRX domain).

Detailed analysis of the kinetics of deformation for steady-state flow (strain of 0.5) in RD, TD and ND specimens of AZ31 has been done earlier [42] using the standard kinetic equation [51] relating the flow stress ($\sigma$) to strain rate ($\dot{\varepsilon}$) and temperature ($T$), given by

$$\dot{\varepsilon} = A\sigma^n \exp\left(-\frac{Q}{RT}\right)$$

where $A = $ constant, $n = $ stress exponent, $Q = $ activation energy, and $R = $ gas constant. The values of $n$ and $Q$ evaluated for the specimens with the above three orientations are shown in Table 1. For RD specimens, a value of stress exponent $n = 4.5$ was estimated along with an apparent activation energy of 143 kJ/mole. As an example, the Arrhenius plot of $[n(\log \sigma)]$ versus $(1/T)$ for the strain rate of 0.0003 s$^{-1}$ is shown in Fig. 7, which reveals that the data representing conditions within the domain fit well with the rate equation (Eq. (3)). The apparent activation energy estimated for RD specimens (143 kJ/mole) is close to that for lattice self-diffusion in magnesium (135 kJ/mole) [52]. Thus, the rate-controlling process in the DRX domain is lattice self-diffusion controlled.

In hot working studies, it is common to represent the variation of grain size with temperature and strain rate in terms of the temperature compensated strain rate (Zener-Hollomon) parameter.

Table 1

<table>
<thead>
<tr>
<th>Specimen orientation</th>
<th>Temperature (°C) and strain rate (s$^{-1}$) ranges</th>
<th>Kinetic parameters</th>
<th>Grain size ($d$, µm) vs. $Z$ (s$^{-1}$) correlation</th>
</tr>
</thead>
<tbody>
<tr>
<td>RD</td>
<td>350–550 and 0.0003–0.3</td>
<td>$n = 4.50$</td>
<td>$Q = 143$ kJ/mole</td>
</tr>
<tr>
<td></td>
<td>Domain (1) 300–500 and 0.0003–0.01</td>
<td></td>
<td>$\log d = 2.582 - 0.164 \log Z$</td>
</tr>
<tr>
<td></td>
<td>Domain (2) 300–500 and 1–10</td>
<td></td>
<td></td>
</tr>
<tr>
<td>TD</td>
<td>Domain (1) 300–500 and 0.0003–0.01</td>
<td>$n = 5.26$</td>
<td>$Q = 180$ kJ/mole</td>
</tr>
<tr>
<td></td>
<td>Domain (2) 300–500 and 1–10</td>
<td></td>
<td>$\log d = 2.644 - 0.137 \log Z$</td>
</tr>
<tr>
<td>ND</td>
<td>Domain (1) 300–500 and 0.0003–0.01</td>
<td>$n = 4.75$</td>
<td>$Q = 168$ kJ/mole</td>
</tr>
<tr>
<td></td>
<td>Domain (2) 300–500 and 1–10</td>
<td></td>
<td>$\log d = 2.735 - 0.155 \log Z$</td>
</tr>
</tbody>
</table>

Curved grain boundaries typical of DRX. The domain therefore represents DRX of the alloy and the temperature and strain rate conditions at which it occurs with peak efficiency are 550 °C and 0.0003 s$^{-1}$.
Fig. 7. Arrhenius plot showing linear relationship between \( [n \log \sigma] \) and \((1/T)\) at a strain rate of 0.0003 s\(^{-1}\) for RD, TD and ND specimens.

\[ Z, \text{ defined as } [51] \]

\[ Z = \dot{\varepsilon} \exp \left[ \frac{Q}{RT} \right] \]  

In Fig. 8, the variation of \( \log(d) \) with \( \log(Z) \) is shown where \( d \) is the average grain diameter in \( \mu \text{m} \) and \( Z \) is in s\(^{-1}\). It may be noted that a straight line relation is observed for the data corresponding to the domain (350–550°C and 0.0003–0.01 s\(^{-1}\)). The equation describing the grain size correlation with \( Z \) in the DRX domain is given in Table 1.

3.3. Deformation parallel to the transverse direction (TD specimens)

The shapes of the true stress–true strain curves for the TD specimens are similar to those recorded on the RD specimens (Fig. 3). The processing map obtained at a strain of 0.5 (steady state flow) for TD specimens is shown in Fig. 9, which exhibits two domains:

- Domain (1) occurs in the temperature range 300–520°C and strain rate range 0.0003–0.01 s\(^{-1}\) with a peak efficiency of about 46% occurring at 400°C and 0.0003 s\(^{-1}\).
- Domain (2) occurs in the temperature range 300–500°C and strain rate range 1–10 s\(^{-1}\) with a peak efficiency of about 42% occurring at 300°C and 10 s\(^{-1}\).

3.3.1. Domain (1) of TD map

The variation of grain size with temperature at the strain rate of 0.0003 s\(^{-1}\) (in Domain 1) is shown in Fig. 10. The grain size increased slowly up to the temperature of about 450°C beyond which grain growth has occurred. The variation is similar to that observed in the RD specimens (Fig. 5), although the grain growth at 550°C is somewhat less in the TD specimens. A plot of the efficiency variation (Fig. 10) showed that the efficiency of power dissipation decreases at temperatures where grain growth occurs. The microstructure on the specimen deformed at 400°C and at 0.0003 s\(^{-1}\) is shown in Fig. 11(a), which exhibits curved grain boundaries typical of DRX and is similar to that observed on RD specimens (Fig. 6). Thus, the domain may be interpreted to represent DRX, which occurs in the TD specimens with a peak...
efficiency of 46% at 400 °C and 0.0003 s⁻¹. In Domain 1, kinetic analysis of the data for the TD specimens at the strain of 0.5 (steady-state flow) has been done in detail [42], and a typical plot of \([n \log(\sigma)]\) versus \((1/T)\) at a strain rate of 0.0003 s⁻¹ is shown in Fig. 7. The apparent activation energy estimated is 180 kJ/mole (Table 1) which is much higher than that for lattice self-diffusion in magnesium (135 kJ/mole) [52] and suggests that cross-slip may be the rate controlling process. The relationship between the average grain diameter and the \(Z\) parameter (Eq. (4)) in domain (1) of the processing map for the TD specimens is shown in Fig. 8, and a linear relationship is observed for the data within the domain as per the equation given in Table 1.

### 3.3.2. Domain (2) of TD map

The variation of grain size with temperature at the strain rate of 10 s⁻¹ (Domain 2) of the processing map for TD specimens is shown in Fig. 10 along with that of efficiency of power dissipation. The grain size has gradually increased and the efficiency decreased with increased temperature in this domain. The microstructure of the specimen deformed at 300 °C/10 s⁻¹ (conditions for peak efficiency) is shown in Fig. 11(b) which exhibits fine DRX grain structure. In this domain, the value of stress exponent \(n\) is estimated to be about 4.35 and the Arrhenius plot showing \([n \log(\sigma)]\) versus \((1/T)\) at the strain rate of 10 s⁻¹ (shown in Fig. 12) has yielded an apparent activation energy of 105 kJ/mole, which is lower than that for self-diffusion (135 kJ/mole). At these high strain rates (>1 s⁻¹), recovery mechanisms involving self-diffusion (climb of edge dislocations) or cross-slip of screw dislocations are unlikely to be rate controlling. Alternately, short-circuit diffusion mechanisms like grain-boundary diffusion are to be considered. The activation energy for grain-boundary diffusion in Mg is about 92 kJ/mole [52] and the above apparent activation energy of 105 kJ/mole is close to this value and therefore is the likely mechanism controlling hot deformation in Domain 2. Similar rate-controlling mechanism involving grain boundary diffusion was advocated [50] for hot deformation of electrolytic tough pitch copper deformed at high temperatures and strain rates. The
The shapes of the true stress–true strain curves for the ND specimens are similar to those recorded on the RD and TD specimens. The processing map obtained at a strain of 0.5 (steady state flow) for the ND specimens is shown in Fig. 14. Similar to the map for TD specimens (Fig. 9), it exhibits two domains:

- Domain 1 occurs in the temperature range 300–550°C and strain rate range 0.0003–0.01 s⁻¹ with a peak efficiency of about 48% occurring at 450°C and 0.0003 s⁻¹.
- Domain 2 occurs in the temperature range 300–500°C and strain rate range 1.0–10 s⁻¹ with a peak efficiency of about 36% occurring at 375°C and 10 s⁻¹.

### 3.4.1. Domain (1) of ND map

The variation of grain size and efficiency with temperature at the strain rate of 10 s⁻¹ is shown in Fig. 15, which is similar to that observed in the TD specimens (Fig. 10). The microstructure on specimen deformed at 450°C and at 0.0003 s⁻¹ is shown in Fig. 16(a) which exhibits curved grain boundaries typical of DRX. In a detailed kinetic analysis in this domain [42], the value of stress exponent $n$ of about 4.75 and an apparent activation energy value of 168 kJ/mole have been estimated. Typical Arrhenius plot showing $[n \log(\sigma)]$ versus $(1/T)$ at the strain rate of 0.0003 s⁻¹ is shown in Fig. 7. The apparent activation energy is higher than that for self-diffusion in magnesium (135 kJ/mole) [52] as in the case of TD specimens and again it is likely that cross-slip is the rate controlling mechanism in this domain. The relationship between the average grain diameter and the $Z$ parameter (Eq. (4)) for the ND specimens is shown in Fig. 8 and a linear relationship is observed which may be described by the equation given in Table 1.

### 3.4.2. Domain (2) of ND map

The variation of grain size with temperature in domain (2) of the ND map is shown in Fig. 15 along with the efficiency variation. The grain size has gradually increased with temperature while the efficiency has decreased similar to that observed in Domain 2 of TD map. The microstructure of the specimen deformed at 400°C/10 s⁻¹ is shown in Fig. 16(b) which exhibits typical DRX features like curved boundaries. In this domain, the value of stress exponent $n$ has been estimated to be about 5.39 and the Arrhenius plot showing $[n \log(\sigma)]$ versus $(1/T)$ at the
Fig. 16. Microstructure of ND specimen deformed at (a) 450 °C and 0.0003 s⁻¹ exhibiting dynamic recrystallization in Domain 1, and (b) 400 °C and 10 s⁻¹ exhibiting dynamic recrystallization in Domain 2.

3.5. Change-over region in TD and ND maps

The variation of grain size with strain rate at a temperature of 450 °C in RD, TD and ND specimens is shown in Fig. 17. It may be noted that abnormal grain growth has occurred at 0.1 and 0.01 in the case of TD and ND specimens, respectively. No such change is observed in the case of RD specimens. In the processing maps for TD and ND specimens (Figs. 9 and 14), a change over from Domain 1 to Domain 2 occurs in the strain rate range 0.01–1 s⁻¹ and this region is akin to a bifurcation in dynamical systems terminology [49,53]. Similar bifurcation has also been observed during hot deformation of iron aluminides [54] where dynamic recrystallization and superplastic deformation were the relevant mechanisms. In simple terms, the dissipative energy is highest in the bifurcation region and lowest in the domains on either side and therefore the bifurcation has a saddle-point configuration. Any fluctuations in the applied parameters in the bifurcation region can cause abnormal changes in the dissipative energy state, which in the present case is represented by the microstructure. Thus, the abnormal grain growth observed in the strain rate range 0.01–1.0 s⁻¹ may be attributed to the occurrence of the bifurcations in the processing maps. Microstructural control is difficult if the material is hot worked in these regions where probability plays an important role [49,53].

4. Discussion

A comparison of the processing maps for RD, TD and ND specimens reveals that the hot-working behavior is anisotropic. The features of the maps for TD and ND specimens are similar, although the actual conditions for peak efficiency and the values of apparent activation energy are slightly different. They both exhibited two DRX domains—one at lower strain rates
with cross-slip as likely mechanism and the other at higher strain rates with grain-boundary diffusion as the rate controlling mechanism. On the other hand, the map for RD specimens has totally different features, e.g. it has a single but wide DRX domain in which the rate controlling process is lattice diffusion controlled. The anisotropy of hot workability originates from the differences in the operating slip systems and the associated recovery mechanisms in different directions.

One of the most important mechanisms that provide large scale softening during bulk deformation is DRX which involves nucleation and grain boundary migration processes. Since it is a dynamic process, the relative magnitudes of the rate of nucleation and the rate of grain boundary migration are important in deciding the rate-controlling step [55]. Nucleation involves dynamic recovery either by climb of edge dislocation or cross-slip of screw dislocations and the rate of nucleation \( \dot{N} \) is proportional to the rate of dislocation generation times the rate of their recovery:

\[
\dot{N} \propto \beta \left( \frac{\dot{\epsilon}}{b l} \right) \exp \left( \frac{-Q}{RT} \right) \quad \text{for climb} \tag{5a}
\]

\[
\dot{N} \propto \beta \left( \frac{\dot{\epsilon}}{b l} \right) \exp \left( \frac{-\alpha \mu b^2 d \ln \left( \frac{d}{b} \right)^{1/2}}{kT} \right) \quad \text{for cross-slip} \tag{5b}
\]

where \( \alpha, \beta = \) constants, \( b = \) burgers vector, \( l = \) dislocation length, \( \mu = \) shear modulus, \( d = \) stacking fault width, and \( k = \) Boltzmann’s constant.

In low stacking fault energy metals, \( \dot{N} \) is higher for climb than cross-slip and hence Eq. (5a) is applicable while it is vice versa in the case of high stacking fault energy metals. The rate of grain boundary migration is proportional to the diffusion coefficient \( D \) and the driving force \( \Gamma \) as given by

\[
M \propto \frac{D \Gamma}{kTb} \tag{6}
\]

The rate-controlling step in DRX is decided by the lower one of the two rates \( \dot{N} \) and \( M \) and in low stacking fault energy metals, \( \dot{N} \) is much lower than \( M \) and in high stacking fault energy metals it is vice versa. In magnesium, the operating slip systems and the recovery mechanisms depend on the starting texture as well as the temperature of deformation as discussed below.

In magnesium materials, four different slip systems operate if their critical resolved shear stress (CRSS) is exceeded and these are: (1) basal slip \( \{0 0 0 2\}\{1 1 2 0\} \), (2) prismatic slip \( \{1 0 0 \bar{1} 0\}\{1 1 2 0\} \), (3) first order pyramidal slip \( \{1 0 \bar{1} 0\}\{1 1 \bar{2} 0\} \) and \( \{1 0 1 2\}\{1 1 \bar{2} 0\} \), and (4) second order pyramidal slip \( \{1 1 2 2\}\{1 1 \bar{2} 3\} \). The CRSS for these slip systems is approximately in the ratio of 1:38:50:100. It is well known that slip on any slip system is most favored when the resolved shear stress reaches a maximum which occurs when the slip plane and the slip direction are at 45° orientation with respect to the compression axis.

In the deformation of magnesium at room temperature, basal slip has the lowest CRSS but non-basal slip systems and twinning also contribute to a small extent. Prismatic slip becomes extensive at temperatures higher than about 225 °C and combines the basal slip to enhance the ductility of magnesium. The pyramidal slip systems start contributing significantly to deformation at temperatures higher than about 350 °C. The fundamental parameter that influences recovery is the stacking fault energy (SFE). On basal slip systems, Mg exhibits low SFE (60–78 mJ/m²) [56] while on pyramidal slip systems the SFE is estimated to be high (173 mJ/m²) [57] which promotes cross-slip [58,59].

The primary rolling texture in the starting plate has been identified to consist of the rolling plane parallel to \( \{0 0 0 2\} \) and the rolling direction is \( \{1 0 \bar{1} 0\} \), which is schematically represented in Fig. 18. To confirm this texture, X-ray diffractograms were recorded as-cast material which represents random texture as observed by other investigators [12,35] and also on the rolled material on the rolling surface and on the plane perpendicular to the rolling direction. These are respectively shown in Fig. 19(a–c), which confirm that the primary rolling texture is of the type \( \{0 0 0 2\}\{1 0 \bar{1} 0\} \) similar to that reported by others [30,33,37]. The basal pole intensity on the rolling plane is about 5.5 times the random and the \( \{1 0 \bar{1} 0\} \) pole intensity is about 3.3 times the random.

On the basis of the above primary texture existing in the starting rolled plate, a qualitative correlation of the hot deformation behavior with the texture may be attempted. For specimens compressed parallel to the rolling direction (RD specimens), the \( \{0 0 0 2\} \) planes are oriented parallel to the compression axis and this reduces the basal slip considerably. The prismatic planes \( \{1 0 \bar{1} 0\} \) are either parallel or at 60° with respect to the compression axis, the former orientation reduces prismatic slip and the latter helps to a limited extent. Moreover, for this orientation, slip is favorable on many of the first order and second order pyramidal systems, which are activated at higher temperatures (>350 °C). Also, cross-slip can occur on these pyramidal slip systems extensively [6,39] and cause dynamic recovery which nucleates DRX. Since many of the pyramidal systems are involved, the rate of nucleation will be much higher than...
the rate of grain boundary migration and hence DRX will be controlled by the grain boundary migration rate which depends on self-diffusion. The apparent activation energy estimated for RD specimens (143 kJ/mole) is therefore close to that for self-diffusion (135 kJ/mole).

In the case of specimens deformed parallel to the transverse direction (TD specimens), \{0002\} planes are parallel to the compression axis and this orientation reduces basal slip. The prismatic planes \{10\overline{1}0\} are either perpendicular or at 30° with respect to the compression axis and the latter orientation may contribute to a limited extent to the prismatic slip. The first order pyramidal slip planes \{10\overline{1}1\} and \{10\overline{1}2\} are highly favorable for slip since they are oriented at about 40° and 54°. The contribution from the second order pyramidal system is limited for this orientation since the \{1\overline{1}2\overline{2}\} planes are oriented at about 32° or 60°. Thus, for the TD orientation, the first order pyramidal slip plays a dominant role in the hot deformation. At lower strain rates (Domain 1), there can be considerable cross-slip associated with the most preferred first order pyramidal slip [58,59] causing dynamic recovery and the rate of nucleation will be lower than the rate of grain boundary migration since only one type of slip system is involved. DRX in Domain 1 will then be controlled by the rate of nucleation involving dynamic recovery by cross-slip and the activation energy for which is higher than that for self-diffusion as observed in this case.

As regards the specimens deformed perpendicular to the rolling plane (ND specimens), \{0002\} slip is very much reduced since the basal planes are perpendicular to the compression axis and so is the prismatic slip since the prismatic planes \{10\overline{1}0\} are parallel. The slip direction \langle\overline{1}1\overline{2}0\rangle is also unfavorable since it is at 90° with respect to the compression axis. Thus, the contributions of basal and prismatic slip to plastic deformation are very much reduced. Although, the first order pyramidal planes \{10\overline{1}1\} and \{10\overline{1}2\} are oriented favorably at 62° or 44°, respectively, the slip direction \langle\overline{1}1\overline{2}0\rangle is normal to the compression axis making their activation difficult. On the other hand, the second order pyramidal planes \{1\overline{1}2\overline{2}\} are oriented at 58° with the slip direction \langle\overline{1}1\overline{2}3\rangle at 50°, both of which favor the activation of second order pyramidal slip. However, these slip systems have higher CRSS and require higher temperatures than the first order pyramidal slip. At lower strain rates, and in view of the high stacking fault energy (173 mJ/m²) [57], there is scope for considerable cross-slip [58,59] to cause dynamic recovery which nucleates DRX. Again, since only one type of slip system is involved, the rate of nucleation will be lower than the rate of grain boundary migration and controls DRX. Since the second order pyramidal slip gets activated at higher temperatures than first order pyramidal slip, the DRX domain moves to higher temperatures than in TD specimens.

5. Conclusions

On the basis of characteristics of processing maps and kinetics of hot deformation of specimens compressed parallel to the rolling direction, the transverse direction, or the normal direction of a rolled AZ31 plate in the temperature range 300–550°C...
and strain rate range 0.0003–10 s⁻¹, the following conclusions are drawn:

(1) RD specimens exhibit a wide domain of workability extending through the temperature range 350–550 °C and strain rate range 0.0003–0.3 s⁻¹, in which DRX occurs.

(2) Maps for TD and ND specimens exhibited two DRX domains at strain rates lower than 0.01 s⁻¹ and the other at strain rates higher than 1 s⁻¹.

(3) The apparent activation energies for hot deformation of RD, TD and ND specimens are 143, 180 and 168 kJ/mole, respectively, and the value for RD specimens is closer to that for self-diffusion.

(4) Dynamic recrystallization in the RD specimens is controlled by the grain boundary migration rate, being slower than the nucleation rate, and depends on self-diffusion.

(5) In TD and ND specimens where DRX is controlled by the rate of nucleation in which recovery occurs by cross-slip of screw dislocations at lower strain rates and climb controlled by grain boundary diffusion at higher strain rates.

(6) The hot workability of rolled AZ31 alloy plate is anisotropic and deformation in the rolling direction gives higher workability in comparison with the transverse or normal direction.

Acknowledgement

The work presented in this paper has been fully supported by a strategic research grant from City University of Hong Kong (Project Ref. no. 7001605).

References


