Original Article

Dependence of deformation behaviors on temperature for twin-roll casted AZ31 alloy by processing maps

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\section*{ARTICLE INFO}

Article history:
Received 8 June 2019
Accepted 23 August 2019
Available online 21 September 2019

Keywords:
Twin-roll cast AZ31 alloy
Hot compression
Processing map
Deformation mechanism

\section*{ABSTRACT}

Hot deformation behaviors of the Twin-roll Casted (TRCed) AZ31 magnesium alloy were studied by uniaxial compression tests under a temperature range of 473–673 K and strain rate of 0.001–1s\textsuperscript{-1}. The effects of the temperature on activation energy and recrystallization were discussed, and the degree of DRX had also been discussed in detail considering the activation energy and processing maps. The results showed that the activation energy is significantly dependent on temperature, and its value increases gradually with the increasing of temperature. The activation energy increased to 192 kJ/mol when the deformation temperature is within 473–673 K. The values of Q are close to the energy of lattice self-diffusion of magnesium alloy in the region of 473–523 K and 523–623 K. Basal slip and twinning are the main deformation mechanisms in the region of low temperature (473–523 K). The critical shear stress of non-basal slips decreases with the increasing of temperature, so prismatic slips and pyramidal slips are easy to initiate when the deformation temperature raise to 523–623 K. The optimized processing parameters at 523–580 K/0.007–0.018s\textsuperscript{-1} of TRCed AZ31 alloy were obtained from processing maps. Through the verification of the microstructure, it was found that the best performance occurs under the deformation condition of 573 K/0.01s\textsuperscript{-1}. More importantly, the practical rolling shows that the suitable processing fitting of the formed area is correct.

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https://doi.org/10.1016/j.jmrt.2019.08.044
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1. Introduction

Due to the low density, high strength, good mechanical properties, and shock load capacity, magnesium and its alloys are widely used in electronics, automotive and aerospace fields [1,2]. However, Mg-based alloy is also faced with the challenge of poor formability at room temperature (RT) due to the limited numbers of deformation modes to fulfill the Von-Mises Criterion [3,4].

With the rapidly expanding of the application of magnesium plates, twin-roll casting (TRCed) has been widely in consideration, which possesses the advantages of high productivity, simplified procedures and cost-effective [5]. However, TRCed magnesium alloy plates are seldom used directly in the industry due to their poor properties of the casting. Therefore, secondary processing is necessary to improve its mechanical properties. Among the methods, rolling is a deformation way with great advantages.

Recently, the studies on the rolling of magnesium alloy mainly focus on the influence of rolling method [6,7], temperature [8] and speeds [9] on the mechanical properties. M. Ullmann et al. [10] studied the static recrystallization behaviors of TRCed AZ31 during the hot deformation. In this research, the hot deformation behaviors of TRCed AZ31 magnesium alloy were studied by analyzing the relationship between activation energy and deformation mechanism through thermal simulation experiment. However, few works were carried out to investigate the relationships between activation energy and deformation mechanism in the hot deformation behaviors, especially the guidance of processing map on rolling.

The deformation ability of alloys is usually expressed by using a form of true stress-true strain curves under certain conditions [11]. The mechanism of work hardening or recrystallization softening can be significantly characterized by the stress-strain curve and the flow behavior of materials. In addition, these deformation mechanisms are affected by processing parameters, such as temperature (T), strain (ε) and strain rate (D) and so on [12,13]. The effects of deformation parameters on thermal processing properties were summarized in hot processing maps [14], which were contributed to affirming the optimal processing window of the TRCed alloy sheets. The technique of processing map based on the dynamic materials model (DMM) is helpful to identify different deformation mechanisms within different deformation temperature and strain rate ranges and has been successfully used to analyze and optimize the hot processability in a wide range of materials including magnesium alloys [15,16], Titanium alloys [17] and so on. Liu et al. [18] put forward the 3-D processing maps firstly with considering the effect of strain on the formability.

The activation energy (Q) obtained by kinetic analysis is generally considered as a key index to characterize the difficulty of hot deformation, and help to design a proper process line to obtain the change rule of the activation energy of material with deformation. In some studies, the activation energy of thermal deformation is always treated as a constant [19,20]. While some studies have shown that Q value changes with different hot deformation conditions in recent years [21,22]. Therefore, the influence of different deformation temperatures on the activation energy is fully discussed in this study, so as to reveal the deformation mechanism of material in different temperature regions. In this paper, the relationship between the power dissipation efficiency and the deformation temperature by drawing three-dimensional power dissipation diagrams and hot working diagrams were studied under different strains. The machinable area and instability regions of TRCed AZ31 magnesium alloy were determined by microstructure analysis and verified by practical rolling test.

2. Experimental procedures

Twin-roll Casted (TRCed) AZ31 magnesium alloy plate was picked from the exit of casting roller for research, with a thickness of 6.88 mm. The chemical composition of the alloy is shown in Table 1. The initial microstructure of the TRCed AZ31 includes coarse grains, a large number of twins and a small number of dynamically recrystallized grains, which presents a typical rolling microstructure, as shown in Fig. 1. The Gleeble-3800 thermal simulation test was conducted with a temperature of 473 K, 523 K, 573 K, 623 K, 673 K and the strain rates of 0.001 s⁻¹, 0.01 s⁻¹, 0.1 s⁻¹ and 1 s⁻¹ respectively. The reduction used in the compression test was 50%, as the processing line i shown in Fig. 2. Fig. 3 displays the processing diagram of the sample. The sample of the subscale test was selected to the cylindrical with the size of ø4 × 6 mm, which is because of the thickness of the TRCed alloy is just 6.88 mm that does not meet the processing conditions of gleeble standard sample. Single-pass isothermal rolling with a reduction of 50% was carried out on TRCed AZ31 plate, obtaining a hot rolled plate with a thickness of 3.44 mm, as processing line ii shown in Fig. 2. The experimental procedure was carried out by a four-high mill with ø320 mm × 340 mm work roll and 0.2 m/s rolling speed.

The microstructure and texture under different processing parameters were characterized by optical microscopy (OM) and electron backscattered diffraction (EBSD). The samples were mechanically ground and polished with diamond sandpaper and silica suspension, and then etched with picric acid (3 g), acetic acid (20 ml), distilled water (20 ml) and ethanol (50 ml) solution to prepare the samples and observe the grain structure through optical microscopy. The EBSD observa-

| Table 1 – Chemical elements of AZ31 magnesium alloy sheet content (wt.%). |
|---|---|---|---|---|---|---|---|---|---|
| Al | Zn | Mn | Fe | Si | Cu | Ni | Mg |
| 2.8 | 0.88 | 0.29 | 0.04 | 0.1 | 0.0015 | 0.0047 | Bal |
Fig. 1 – The initial microstructure of Twin-roll Casted AZ31 alloy.

Fig. 2 – Flowchart of the cast-rolling short process: (i) processing of hot compression (ii) process of rolling.

Fig. 3 – The diagram of compressing specimen geometry and the compressive deformation processes.
tion was carried out by using a Zeiss SEM with an acceleration voltage of 20 kV, and a 70° sample tilt angle, after electropolishing with the temperature of 248 K and current of 0.072 A.

3. Results and discussion

3.1. Flow behavior analysis

Fig. 4 shows the true stress-true strain curves of TRCed AZ31 at different strain rates and deformation temperatures. It can be seen that the true stress-true strain curves of the TRCed AZ31 have similar variation trends under different deformation conditions. First of all, the stress increases rapidly under small strain, can be ascribed to work hardening. Dislocation increases and accumulates rapidly in the stage of work hardening, which leads to the rapid increase of flow stress. Then the stress increases slowly at the flow softening stage. The effect of hardening can be partially offset by the occurrence of dynamic softening mechanism in the softening stage, resulting in a slow increase or decrease of flow stress. Finally comes the stabilization phase at high strain, which indicates that dynamic softening mechanisms occur during the thermal deformation of the material, such as dynamic recovery (DRV) and dynamic recrystallization (DRX) [23–25]. The flow stress remains stable due to the dynamic balance between the machining hardening and dynamic softening in the stable stage. It is worth mentioning that the flow stress is significantly affected by the deformation temperature and strain rate. The rheological stress is close to 190 MPa when the deformation temperature is 473 K and the strain rate is 1 s⁻¹. The rheological stress decreased significantly when the deformation temperature was 673 K. The change of stress value at other strain rates showed the same trend which was due to the decrease of interatomic binding force and critical shear stress caused by the increase of deformation temperature.

In addition, dynamic recrystallization is more likely to occur due to increased softening and has a significant impact on deformation with increasing temperature. Furthermore, the peak stress is only 80 MPa at the strain rate of 0.001 s⁻¹ and the temperature of 473 K. However, the stress value goes up to 190 MPa when the strain rate is 1 s⁻¹, as shown in Fig. 4(d). Similar trends were found at other temperatures. It can be concluded that the flow stress increases with the increase of strain rate at a certain temperature. With the increase of strain rate, the speed of dislocation movement is too quick to provide enough time for the occurrence of dynamic recrystallization. The dynamic recrystallization is one of the main softening mechanisms in the process of hot deformation.

Fig. 4 – The true stress-strain curves of TRCed AZ31 magnesium alloy at different strain rates: (a) 0.001 s⁻¹, (b) 0.01 s⁻¹, (c) 0.1 s⁻¹, (d) 1 s⁻¹.
3.2. Basic constitutive and deformation mechanism analysis

Hot processing map plays an important role in the microstructure and mechanical properties of products in the industry. Thermal deformation of magnesium alloys is a process controlled by thermal activation, which is mainly affected by deformation temperature, strain rate and deformation amount. Therefore, it is very important to have a correct cognition on the deformation activation energy and microstructure evolution under elevated temperature.

The Zener–Holloman parameter is an effective factor used to describe the influence of deformation temperature and strain rate on material flow behavior [26]. Currently, the most widely used method to calculate the activation energy in the literature is associating Zener–Holloman parameter with flow stress so as to model flow stress, as shown in Eq. (1).

\[
Z = \dot{\varepsilon} \exp \left( \frac{Q}{RT} \right) = \begin{cases} 
A' \alpha^n' \\
A'' \exp (\beta \sigma) \\
A[\sinh (\alpha \sigma)]^n
\end{cases}
\]

Where, \(T\) is the deformation temperature, \(K\); \(Q\) is the activation energy of thermal deformation; \(R\) is a constant of gas and its value is 8.314 J/mol \cdot K; \(A', A'', A, n, n', \beta\) and \(\alpha\) are all constants about the material, where exist an equation: \(\alpha = \beta / n'\). The Arrhenius equation, which has been sinusoidally corrected by the hyperbolic curves containing the deformation activation energy and the deformation temperature, can accurately enough to predict the high-temperature flow characteristics of the material in the whole range of deformation temperature and strain rate.

The peak stress in the process of thermal deformation is used to calculate the parameters in the constitutive equation. The logarithms of both sides of Eq. (2) can be obtained:

\[
\ln \dot{\varepsilon} = \begin{cases} 
n' \ln \sigma_p + \ln A' - \frac{Q}{RT} \\
\beta \sigma_p + \ln A'' - \frac{Q}{RT} \\
n \ln[\sinh(\alpha \sigma_p)] + \ln A - \frac{Q}{RT}
\end{cases}
\]

Thus, the calculation formulas of \(n', \beta\) and \(n\) are shown in formulas (3)–(5):

\[
\begin{align*}
\frac{\partial \ln \dot{\varepsilon}}{\partial \ln \sigma_p} & = n' \\
\frac{\partial \ln \dot{\varepsilon}}{\partial \sigma_p} & = \beta \\
\frac{\partial \ln \dot{\varepsilon}}{\partial \ln[\sinh(\alpha \sigma_p)]} & = n
\end{align*}
\]

![Fig. 5 – Relationship between (a) \(\ln \dot{\varepsilon}\) and \(\ln \sigma\), (b) \(\ln \dot{\varepsilon}\) and.](image)

![Fig. 6 – The relationship between \(\beta\) and the temperature.](image)
Table 2 – Various parameter values in different temperature region.

<table>
<thead>
<tr>
<th>Temperature/K</th>
<th>n′</th>
<th>β</th>
<th>α</th>
<th>n</th>
<th>Q/nR</th>
<th>Q/kJ/mol</th>
</tr>
</thead>
<tbody>
<tr>
<td>LR</td>
<td>473-523</td>
<td>9.024</td>
<td>0.083</td>
<td>0.009</td>
<td>6.87</td>
<td>1833.12</td>
</tr>
<tr>
<td>MR</td>
<td>523-623</td>
<td>8.489</td>
<td>0.112</td>
<td>0.013</td>
<td>6.42</td>
<td>2090.85</td>
</tr>
<tr>
<td>HR</td>
<td>623-673</td>
<td>8.442</td>
<td>0.163</td>
<td>0.019</td>
<td>6.43</td>
<td>3597.23</td>
</tr>
<tr>
<td>Average</td>
<td>473-673</td>
<td>8.456</td>
<td>0.115</td>
<td>0.014</td>
<td>6.07</td>
<td>2496.04</td>
</tr>
</tbody>
</table>

Fig. 7 – The $\ln \dot{\varepsilon} - \ln[\sinh(\alpha)]$ curves in different regions: (a) 473 K–523 K, (b) 523 K–623 K, (c) 623 K–673 K.

$\ln \dot{\varepsilon} - \ln \sigma$ and $\ln \dot{\varepsilon} - \sigma$ curves under different conditions are shown in Fig. 5. As can be seen from Fig. 5(a), the curves obtained at each deformation temperature are parallel to each other, which indicates that $n'$ hardly change with the various temperatures. Therefore, the average slope of the curve at each temperature is the value of $n'$. As indicated in Fig. 5(b), the slopes of curves are significantly different at different deformation temperatures, indicating that the value changes with the various temperatures.

A further observation (Fig. 6) reveals that $\beta$ increases as the temperature increases. However, the value increases differently during different temperature ranges. The $\beta$ increases rapidly with the increase of temperature at high temperatures, indicating that $\beta$ is more sensitive to temperature. Both $\beta$ and $\alpha$ are constants about the material, which are very important parameters in the solution process of deformation activation energy, and $\beta$ is obtained from $\alpha = \beta/n'$. Therefore, the calculation accuracy of deformation activation energy will be seriously affected if the influence of temperature on the value is ignored according to the previous empirical formula. In other words, the whole temperature range is divided into three regions according to different sensitivities to temperature expressed by the slope of the $\beta - T$ relation. The low-temperature region (LR) is defined as the
temperatures from 473 K to 523 K. The medium temperature region (MR) is defined as the temperatures from 523 K to 623 K. The high-temperature region (HR) is defined as the temperatures from 623 K to 673 K. According to the value of $n'$ and $\beta$ obtained from the curves at different temperatures, the corresponding $\alpha$ can be obtained from the formula $\alpha = \beta/n'$, and the results are shown in Table 2.

The $\ln \varepsilon - \ln[\sinh(\alpha)]$ curves in different regions were drawn in Fig. 7 to obtain the average values of each region. Curves of different temperature ranges are shown in Fig. 7. The average $n$ value of LR is 6.87, which is obtained from the $\ln \varepsilon - \ln[\sinh(\alpha)]$ curve in Fig. 7(a). Similarly, the mean $n$ of MR and HR are 6.42 and 6.43 respectively, as shown in Figs. 7(b, c).

The $\ln[\sinh(\alpha)] - 1/T$ curves in different regions were drawn in Fig. 8 to obtain the $Q/nR$ values of each region, through which the deformation activation energy $Q$ of each temperature region can be obtained from Eq. (6). Similarly, the activation energy $Q_a$ over the entire deformation temperature range can also be obtained from Eq. (6), by inputting the average of other parameters in the equation. The results are shown in Table 2. The deformation activation energy of the alloy increases with the increase of deformation temperature, and the increase rate is larger especially in the transition stage from the MR to HR, as shown in Fig. 9. This may be related to the large dislocation consumption caused by DRX. The same results about the trend of activation energy with temperature were obtained from Deng et al [27].

$$Q = \frac{\partial \ln[\sinh(\alpha)]}{\partial(1/T)} \cdot n \cdot R$$

Fig. 8 – The $\ln[\sinh(\alpha)] - 1/T$ curves in different regions: (a) 473 K–523 K, (b) 523 K–623 K, (c) 623 K–673 K.

Fig. 9 – The difference between $Q$ and $Q_a$ at different temperature intervals.
The Q of TRCed AZ31 in LR and MR region are 104 kJ/mol and 111 kJ/mol respectively, which is close to the energy of lattice self-diffusion of magnesium alloy. That is to say, DRX is dislocation climb controlled by lattice self-diffusion in this region, and magnesium alloy is dominated by basal slip and a small amount of non-basal slip. The Q of TRCed AZ31 in HR region is 192 kJ/mol, which is related to the fact that the formation of DRX is accompanied with a large amount of dislocation consumption. With the increase of deformation temperature, dynamic recrystallization is more likely to occur, meanwhile, the dislocation consumption is more severe and stress concentration begins to relax, which will result in less potential dislocation sources and more difficulty initiation. Thus, it is reasonable to propose that the deformation activation energy increases rapidly when the deformation temperature reaches 673 K.

When the material is deformed at LR (473–523 K), the basal slip and twinning are the main deformation mechanisms within this temperature range. At low deformation temperature, the dislocation energy of magnesium alloy is low, and the critical shear stress required for a non-basal slip cannot be reached [28], so the base plane slip can only be started with few independent slip systems. It can be seen from the above statement that the activation energy of deformation is proportional to the temperature, and the change of the power dissipation efficiency with the temperature is caused by the different deformation mechanisms under different deformation temperatures, and the value of the activation energy of deformation determines the different deformation mechanisms, thus influencing the power dissipation efficiency. Therefore, the distribution of power dissipation under different deformation parameters and its influence on microstructure will be discussed in the following sections.

3.3. Optimized workability by processing map

3.3.1. The principles of processing map

Based on Dynamic Material Modeling, the thermal processing diagram combines the power dissipation diagram and the
instability diagram to select the appropriate processing area [29,30]. The per unit of input energy $P$ is divided into two parts: content ($G$) and co-content ($J$). Their relationship can be expressed as:

$$P = \sigma \cdot \varepsilon = G + J = \int_0^\varepsilon \sigma \cdot d\varepsilon + \int_0^\sigma \varepsilon \cdot d\sigma$$  \hspace{1cm} (7)

Where, $G$ is the energy consumed by the material when plastic deformation occurs, and $J$ is the energy consumed by the material in the process of tissue evolution during deformation, such as dynamic recovery and dynamic recrystallization. In order to describe the proportion of energy consumed by microstructure evolution during material forming, the strain sensitivity index is introduced:

$$m = \frac{dJ}{dG} = \frac{1}{\sigma \cdot d\varepsilon} = \frac{\partial \ln \sigma}{\partial \ln \varepsilon}$$ \hspace{1cm} (8)

In general, the value is affected by the deformation temperature and strain rate. The dissipation covariance reaches the maximum value when the material is in an ideal linear dissipation state.

$$J_{\text{max}} = \frac{1}{2}P$$ \hspace{1cm} (9)

The ratio of energy dissipated by tissue evolution and linear dissipation in plastic deformation of materials is expressed by the dimensionless parameter, which is called power dissipation factor, and its expression is:

$$\eta = \frac{J}{J_{\text{max}}} = \frac{2m}{m + 1}$$ \hspace{1cm} (10)

The different calculated $\eta$ will be expressed in the form of isolines on the plane of $T - \varepsilon$, which is the power dissipation diagram. Fig. 10 shows the relationship between the power dissipation value and the deformation temperature under different strains. In general, materials have better processability in the case of high power dissipation efficiency. Conversely, areas with low values are always associated with microstructural defects.

Fig. 11 - 3D power dissipation diagrams under different strains: (a) 0.1; (b) 0.3; (c) 0.5; (d) 0.7.
However, in the power dissipation diagram, it is not the case that the greater the power dissipation efficiency value is, the better the internal processing performance of the material will be. For example, fracture and hole formation also have high power dissipation coefficients. Prasad established the instability criterion for the material’s plastic deformation based on the irreversible thermodynamic extreme value principle applied to large plastic rheology [30]. The rheological instability criterion is shown in Eq. (11). When the value is negative, the material is considered to be unstable within the range of the corresponding deformation temperature and strain rate.

\[ \varepsilon_i(T) = \frac{a \ln \left( \frac{m \varepsilon^2}{a} \right)}{b \ln(T)} + m < 0 \]  

(11)

3.3.2. Values of the efficiency of power dissipation \( \eta \) under different parameters

It can be seen from Fig. 10 that the power dissipation efficiency fluctuates greatly with the increase of temperature and strain rate under different strains. Power dissipation efficiency with various temperatures exhibits the same trends under the condition of low strain (0.1, 0.3), at the strain rate of 0.001s\(^{-1}\) and 1s\(^{-1}\). The power dissipation efficiency decreases first and then increases with the increase of deformation temperature.

The variation of power dissipation efficiency with temperature is small and the values are less than 30% when the strain rate is 0.01s\(^{-1}\) and 0.1s\(^{-1}\). It indicates that the dynamic recrystallization ability is weak under low strain conditions, which is mainly associated that smaller deformation and dislocation density is not conducive to the dynamic recrystallization nucleation. It can also be seen from Fig. 10 that the effect of machining hardening dominates and the softening effect of dynamic recrystallization are not obvious. When the deformation temperature is 523–573 K, the values remain almost constant under high strain conditions. Therefore, it can be concluded that the variation of the power dissipation coefficient is mainly affected by the strain rate at the temperature of 523 K and 573 K. The values of \( \eta \) decreases with the increase of deformation temperature, when the deformation temperature is 473–523 K and the strain rate is 0.001s\(^{-1}\) and 1s\(^{-1}\). When the strain rate is 0.01s\(^{-1}\) and 0.1s\(^{-1}\), the value increases with the increase of the deformation temperature. On the contrary, the change rule of power dissipation efficiency with temperature is opposite to that at low temperature, when the deformation
temperature is 573–673 K. It is worth noting that the variation of power dissipation efficiency at high temperatures is just opposite to that at low temperatures.

Taking the influence of deformation temperature, strain rate and strain on power dissipation factor into consideration, the three-dimensional power dissipation diagrams under different strains were drawn, as shown in Fig. 11. Different colors represent a different percentage of power dissipation factor. As shown in Fig. 11, the power dissipation efficiency increases significantly with the increase of strain, resulting in the degree of dynamic recrystallization increases with the increase of strain in the deformation process. The increase of power dissipation efficiency was caused by the continued increases of dislocation density resulting from the increasing deformation reduction, which would accelerate the rate of dynamic recrystallization nucleation and increase the number of recrystallized grains, consumes more deformation energy and increase the dissipation covariance.

The regions with higher power dissipation value are mostly concentrated at higher deformation temperatures or lower strain rates. This phenomenon is related to the fact that the main softening mechanism of magnesium alloy in the deformation process is dynamic recrystallization, which is usually accompanied by dislocation recombination, climbing and other behaviors that related to the diffusion of atoms. The rate of dynamic recrystallization increases with the increase of temperature, which leads to the high power dissipation in the high-temperature region. The increase in temperature increases the diffusion ability of atoms and the proportion of energy consumed by the evolution of microstructure during the deformation process. When the strain rate is small, the nucleation and grain growth time of dynamic recrystallization are sufficient, so the power dissipation value is large in the region with low strain rates.

3.3.3. Processing map analysis

Fig. 12 shows the hot working diagram of the cast-rolled AZ31 magnesium alloy under the true strains of 0.3, 0.5 and 0.7. Pink areas represent the flow instability regions, and black contours mean the efficiency of power dissipation, as shown in the processing map. It can be seen from Fig. 12 that contour shapes of the instability regions are similar, especially at high strain rates, and the proportion of the instability region increases with the increase of strain. Three focus areas of the hot processing map were marked as I, II, III, respectively, according to the law of the instability distribution. The low temperature-high strain rate deformation region is named as "Region I", and the proportion decrease with the increase of strain, but the change is not obvious. With the increase of strain, a new instability region II appeared in the area of medium temperature-high strain rate. However, this instability region is disappeared with the increase of deformation temperature. The range of region III significantly reduced with the increase of strain, which occurred in high temperature-high strain rate region. By analyzing the processing map, we can find that the most appropriate area for TRCd AZ31 magnesium alloy is the deformation temperature at 523–593 K and the favorable strain rate at 0.007–0.018s⁻¹; the power dissipation coefficient is greater than 30% and the instability parameter is greater than 0; dynamic recrystallization is more likely to occur than dynamic recovery when the power dissipation efficiency is greater than 30% for low fault-energy materials, such as magnesium alloys [31]; therefore, the dynamic recrystallization of the alloy is easy to occur during the plastic deformation.
The macroscopic morphology of the cast-rolled AZ31 magnesium alloy in the instability zone and the partially compressed sample in the machinable zone within the experimental range is shown in Fig. 13. It can be seen that the samples deformed in the instability zone are seriously cracked, while the samples deformed in the machinable zone are well-surfaced and have no obvious cracks. Therefore, the processing diagram can be used to evaluate the processing performance of TRCed AZ31 magnesium alloy in the whole processing range.

3.4. Microstructural evolution

Fig. 14 shows the microstructure under four different processing conditions. Fig. 14(a) demonstrates many slip bands and twins and few twin-recrystallized grains in the microstructure under the deformation condition of 473K/0.01s$^{-1}$. This phenomenon is mainly due to the fact that the deformation temperatures in this region are low and the critical shear stresses of non-basal slips are much larger than that of base slip at low temperatures. Therefore, non-base slips are difficult to start and the base slip is easy to occur at low temperatures.

The base slip can only supply the strain perpendicular to C axis, while the pyramidal slip and twinning can provide the strains of C axis to coordinate deformation. It can be seen from Fig. 14(a) that closely arranged slip bands and few dynamic recrystallization grains can be found in the twin intersecting regions and twin boundaries.

Fig. 15 shows the dynamic recrystallization distribution maps under corresponding processing conditions. Different colors represent the different states of grains, “blue” represents the grains that have undergone completely dynamic recrystallization, “yellow” represents the sub-grain states, and “red” represents the deformed grains. Where, grain boundary with an orientation angle of $2^\circ$–$15^\circ$ is defined as the low-angle grain boundary (LAGB), which is represented by a solid green line as shown in the figure. Grain boundary with orientation angle greater than $15^\circ$ is defined as the high-angle grain boundary (HAGB), which is represented by a solid black line as shown in Fig. 15. As anticipated, dynamic recrystallization hardly occurs and most of the grains have existed as sub-structure under the condition of 473K/0.01s$^{-1}$, as shown in Fig. 15(a). The deformation resistance increases with the decrease of temperature, leading to poor plastic deformation.
Fig. 15 – Recrystallization distribution maps of compressed TRCed AZ31 alloys under various conditions: (a) 473 K/0.01s$^{-1}$, (b) 623 K/0.01s$^{-1}$, (c) 573 K/1s$^{-1}$, (d) 573 K/0.01s$^{-1}$.

ability, and further the workpiece cracking and instability, which can be verified by Fig. 13(a).

Fig. 14(b) shows the microstructure of deformation in the region with a deformation temperature of 623 K and strain rate of 0.01s$^{-1}$. The power dissipation efficiency decreases from 40% to 9.7% as the strain rate increases, when the strain rate is less than 0.1s$^{-1}$. However, the power dissipation efficiency increases to 36% as the strain rate increases when the strain rate is large than 0.1s$^{-1}$, indicating that dynamic recrystallization will occur in this region. The microstructure after deformation is extremely uneven and shows a significant necklace structure [32,33], as shown in Fig. 14(b), resulting from the significant difference between recrystallized grains and non-recrystallized grains. As shown in the analysis of the deformation activation energy, the critical shear stress of material non-basal slip is low in the medium temperature region. The critical shear stress of non-basal slips decreases with the increasing of temperature. However, the thermal motion rate of atoms increases with the increase of temperature. Dynamic recrystallization occurs in TRCed AZ31 magnesium alloy when deformation occurs within this temperature range, as shown in Fig. 14(b). “Necklace” shaped dynamic recrystallization grains appeared in the microstructural, indicating that continuous dynamic recrystallization occurred in the magnesium alloy [34]. At the same time, macroscopic cracks also appear in the deformed samples. It also can be seen from the processing map shown in Fig. 13(b) that rheological instability occurs within the range of HR.

Fig. 15(b) shows the recrystallization distribution corresponding to different deformation conditions. It can be seen from the figures that some dynamic recrystallization occurred at the grain boundaries of the sub-grains and the deformed grains. The most grains were still in the form of the substructure, and the deformed grains reduced to some extent.
The microstructure of TRCed AZ31 deformed at the medium temperature-high strain rate is characterized by fewer deformation grains and subgrains due to greater recrystallization, as shown in Fig. 14(c). When the material deforms in the high-temperature region, the main deformation mechanisms of TRCed AZ31 magnesium alloy are non-basal slips, containing prismatic slips, pyramidal slips and cross slips. Fine DRX grains appear at the grain boundaries and within the twins, as shown in Fig. 15(c). Due to the various local strain incompatibility, the crack initiation occurred in the region with weak binding force between grains that not easily deformed. Therefore, processing conditions leading to incompatible strain are considered to be unfavorable. Fig. 15(d) shows the recrystallization distribution of TRCed AZ31 in the stable region. It can be seen that a large proportion of dynamically recrystallized grains (nearly 50%) and sub-grains have appeared in the selected areas. As is known to all, DRX is conducive to improving the inherent formability of the magnesium alloys. In summary, the optimal parameters of Twin-roll Casted AZ31 alloy during hot processing can be defined by the processing map and the results of the microstructural observation. The most suitable processing parameters for this sheet should be in the region of medium temperatures (523–580 K) and low strains (0.007–0.018 s⁻¹).

3.5. Practical hot rolling

TRCed AZ31 magnesium alloy sheets with a thickness of 6.88 mm were isothermally rolled by a four-high rolling mill. The macroscopic morphology and microstructures of rolled sheets are shown in Fig. 16. The morphology of the rolled plate under the condition of 573 K/0.01 s⁻¹, which belongs to the suitable processing area of the hot processing map, is shown in Fig. 16(a). It can be seen that after rolling with a large reduction of 50%, the surface of the hot-rolled plate is smooth and exists a few cracks on the edge. The microstructure exhibited uniform and fine grains with an average grain size less than 10 μm. However, serious cracks appeared in the plate after rolling with large deformation. The microstructure is extremely non-uniform and even has several cross cracks in the width direction of the plate, as shown in Fig. 16(c). The processing parameters of the hot rolling were 473 K/0.01 s⁻¹, which is attached to the instability region of the hot processing map. Along the rheological stress zone of metal deformation, there are many coarse original grains without DRX and cracks appeared in some areas, as shown in Fig. 16(d). The results on the practical rolling test showed that the established processing map played a scientific guiding role in the selection of subsequent deformation parameters of TRCed AZ31 magnesium alloy.

4. Conclusion

According to the processing maps and microstructure analysis of the alloy, the optimal processing ranges of TRCed AZ31 magnesium alloy are proposed as follows: 533–583 K/0.007–0.018 s⁻¹ and 653–673 K/0.6–1 s⁻¹. The highest power dissipation efficiency is 36%. The activation energy of TRCed AZ31 in LR and MR region is closed to the.
energy of lattice self-diffusion of magnesium alloy, which means that DRX in this region is dislocation climb controlled by lattice self-diffusion, and dominated by basal slip. With the increase of deformation temperature, the deformation activation energy increased rapidly which is caused by severe dynamic recrystallization. In the HR, the main deformation mechanism is the non-basal slip including prismatic slip, pyramidal slip, and cross slip. The optimum parameters of TRCed AZ31 alloy are determined at 523−580 K/0.007−0.018 s−1, according to processing maps. Furthermore, the actual rolling also verified that the best performance can be obtained at the condition of 573 K/0.015 s−1, which indicate that the hot processing maps have a great significance to the rolling prediction and Magnesium alloy sheet production.

Acknowledgments

This work was supported by the National Natural Science Foundation of China (Nos. U1610253 and 51604181), The Key Research and Development Program of Shanxi Province (No. 201603D111004), Taiyuan University of Science and Technology Scientific Research Initial Funding (Nos. 20192002 and 20182034), The Fund for Shanxi “1311 Project” Key Subjects Construction (1331KSC).

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