



Hot deformation behaviors and processing maps of Mg–Zn–Er alloys based on Gleeble–1500 hot compression simulation

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Abstract: The hot deformation behaviors of as-solution Mg– x Zn– y Er alloys ($x/y=6$, $x=3.0$, 4.5 and 6.0; $y=0.50$, 0.75 and 1.00) were investigated on Gleeble–1500 thermal simulator in a temperature range of 200–450 °C at a strain rate of 0.001–1 s^{−1}. The true stress–strain curves showed the dynamic competition between the working hardening and working softening mainly due to the dynamic recrystallization (DRX) occurring during hot compression. The constitutive equations were constructed which could accurately predict the peak stress of the alloys. The addition of Zn and/or Er resulted in higher deformation activation energy for Mg–3Zn–0.5Er (alloy A). The processing maps were constructed as function of the temperature and the strain rate, providing the optimum hot working conditions (i.e., at strain of 0.3, Mg–3Zn–0.5Er (alloy A): 380–430 °C, <0.1 s^{−1}; Mg–4.5Zn–0.75Er (alloy B): 380–450 °C, 0.01–0.1 s^{−1}; Mg–6Zn–1Er (alloy C): 390–440 °C, 0.01–0.1 s^{−1}). The as-solution treated Mg–4.5Zn–0.75Er (alloy B) demonstrated more optimum hot working window comparing with Mg–3Zn–0.5Er (alloy A) and Mg–6Zn–1Er (alloy C).

Key words: Mg–Zn–Er alloy; compression behavior; microstructure evolution; constitutive equations; processing map

1 Introduction

As the lightest metallic structural materials, magnesium alloys have been received much attention in the last decade because of their low density, high strength and recycling, especially for the Mg–Zn–RE alloys [1–3]. According to the previous work, three kinds of ternary equilibrium phases in Mg–Zn–RE alloys have been found, i.e., *W*-phase (Mg₃Zn₃RE₂, cubic structure), *I*-phase (Mg₃Zn₆RE, icosahedra quasicrystal structure, quasiperiodically ordered) and Mg₁₂ZnRE long-period stacking order (LPSO) structures (including 6H, 14H and 18R) [4–6]. However, it is reported that the *W*-phase with face-centered cubic structure has weak bonding with Mg matrix [7]. Generally, the *I*-phase and the LPSO structure are used to improve the mechanical properties of alloys effectively. The *I*-phase has many unique properties such as conspicuous hardness, high thermal stability, high corrosion resistance and low surface energy [8]. SINGH et al [2,9] and LE et al [3] have introduced the hot extrusion process to prepare

Mg–Zn–Y alloys and Mg–Zn–Gd alloys strengthened by the *I*-phase, and these alloys exhibited excellent mechanical properties. The Mg–Zn–Y–Zr alloy containing *I*-phase was studied and the yield tensile strength (YTS) reached from 150 to 450 MPa at room temperature [10].

It is well known that the primary *I*-phase formed during solidification process is coarse and uneven distribution, resulting in disablement of improving mechanical properties effectively. Thus, hot deformation methods such as hot extrusion, hot rolling and ECAP are normally used to advance the mechanical properties of magnesium alloys. The mechanical properties of the alloys are greatly improved due to grain refinement, modification of texture and uniform distribution of the broken secondary phase. It should be noted that the nano-scale *I*-phase precipitated during hot deformation plays an important role in further improving mechanical properties of Mg–Zn–RE alloys [2,11–14], because the nano-scale *I*-phase has more uniform distribution and smaller size comparing with the primary *I*-phase.

The aims of the present investigation are to find an

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optimum hot working window in order to obtain a proper hot deformation method to produce a new kind of Mg–Zn–Er alloy reinforced with nano-scale *I*-phase. Lots of works have been reported on the constitutive equation and processing maps of Mg–Zn–RE alloys to understand the thermo-mechanical behaviors [15–18], but few work has been carried out on the as-solution Mg–Zn–Er alloys up to now. Therefore, in the present work, the compressive behaviors and processing maps of the as-solution Mg–*x*Zn–*y*Er (*x*=3.0, 4.5 and 6.0; *y*=0.50, 0.75 and 1.00; *x/y*=6, identified as alloys A, B and C, respectively) will be studied by using a Gleeble–1500 thermal simulator in temperature range of 200–450 °C at a strain rate of 0.001–1 s^{−1}.

2 Experimental

The as-cast alloys were prepared from the pure Mg, pure Zn and Mg–30%Er master alloys in a graphite crucible in an electric resistance furnace under an anti-oxidizing flux. The melt was poured into a steel mold and cooled in air. The chemical compositions of alloys were analyzed by X-ray fluorescence (XRF) analyzer, as shown in Table 1. The as-cast samples were solid-solution treated at 460 °C for 20 h followed by quenching into water. For isothermal compression test, the specimens were machined into a cylindrical sample with a diameter of 10 mm and a height of 15 mm. The longitude direction of the cylinder was parallel to the compression direction.

Table 1 Chemical compositions of Mg–Zn–Er alloys

Alloy	Mass fraction/%			<i>m</i> (Zn)/ <i>m</i> (Er)
	Zn	Er	Mg	
Mg–3Zn–0.5Er	3.06	0.46	Bal.	6.65
Mg–4.5Zn–0.75Er	4.37	0.74	Bal.	5.91
Mg–6Zn–1Er	5.83	1.04	Bal.	5.61

The compression tests were carried out in a temperature range of 200–450 °C at strain rate of 0.001, 0.01, 0.1 and 1 s^{−1} on the Gleeble–1500 thermo-mechanical simulator, respectively. In order to reduce friction, a graphite lubricant was applied between the specimen and crosshead. The specimens were heated to the deformation temperature at heating rate of 5 K/s using induction heating and held for 1 min in order to obtain a stable and uniform temperature prior to deformation. All specimens were deformed to a total true strain of about 0.6 and water-quenched from testing temperature.

The microstructure observation was carried out by an optical microscope (OM, Zeiss-ImagerA2m). The samples for OM were mechanically polished and etched in a solution of 4 mL nitric acid and 96 mL ethanol.

3 Results and discussion

3.1 Flow stress behaviors

Figures 1–3 show the true stress–strain curves of the as-solution treated Mg–3Zn–0.5Er (alloy A), as-solution treated Mg–4.5Zn–0.75Er (alloy B) and as-solution treated Mg–6Zn–1Er (alloy C) at strain rates of 0.001, 0.01, 0.1 and 1 s^{−1}, respectively. They indicated that the flow stress increased to a maximum and decreased to a steady state. The processing was composed of four stages, i.e., working hardening, transition, working softening and steady stages. It was the typical characteristic for hot working accompanied by dynamic recrystallization (DRX), especially when the stress reached the stable state, the curves were not smooth but wave, which was due to the dynamic competition of the strain hardening and DRX softening during hot compression processing [19]. In addition, the peak of flow stress decreased with the increment in temperature and reduction in strain rate. Higher temperature can supply higher dislocation mobility, activate non-basal slip system, advance the DRX and make the dynamical recrystallized grains grown [20]. Meanwhile, lower strain rate can provide longer time for activation energy accumulation [21].

In order to investigate the microstructure evolution of the hot-compressed Mg–Zn–Er alloys, the OM images of the Mg–6Zn–1Er alloys after hot compression under various conditions are shown in Fig. 4. The amount of twins could be found in the alloys and the twin density increased with the increment in strain rate at 200 °C, which led to the work hardening strength with the increase of the strain rate, as shown in Figs. 4(a), (d), (g) and (j). However, a few of recrystallized grains could be observed at the twin boundaries (indicated by arrows in Figs. 4(a), (d), (g) and (j), which suggested that the twins played an important role in improving the ductility of the magnesium alloy at lower temperature (200 °C). When the compression temperature was up to 300 °C, amounts of fine recrystallized grains were developed in necklace along the original grain boundaries in a strain rate range of 0.001–0.01 s^{−1}, as shown in Figs. 4(b) and (e). However, when the strain rate reached up to the range of 0.1–1 s^{−1}, the twins were observed and a large number of fine recrystallized grains were found locating at both original grain boundaries and twin boundaries (as shown in Figs. 4(h) and (k). Although it has been reported in Ref. [22] that the twinning had a great effect on rotating the basal plane towards a more favorable orientation for basal glide at 300 °C, the softening effect of DRX was more obvious, which could explain that the values of the peak stress at strain rates of 0.1 and 1 s^{−1} were higher than those at 0.01 and 0.001 s^{−1}, respectively. When

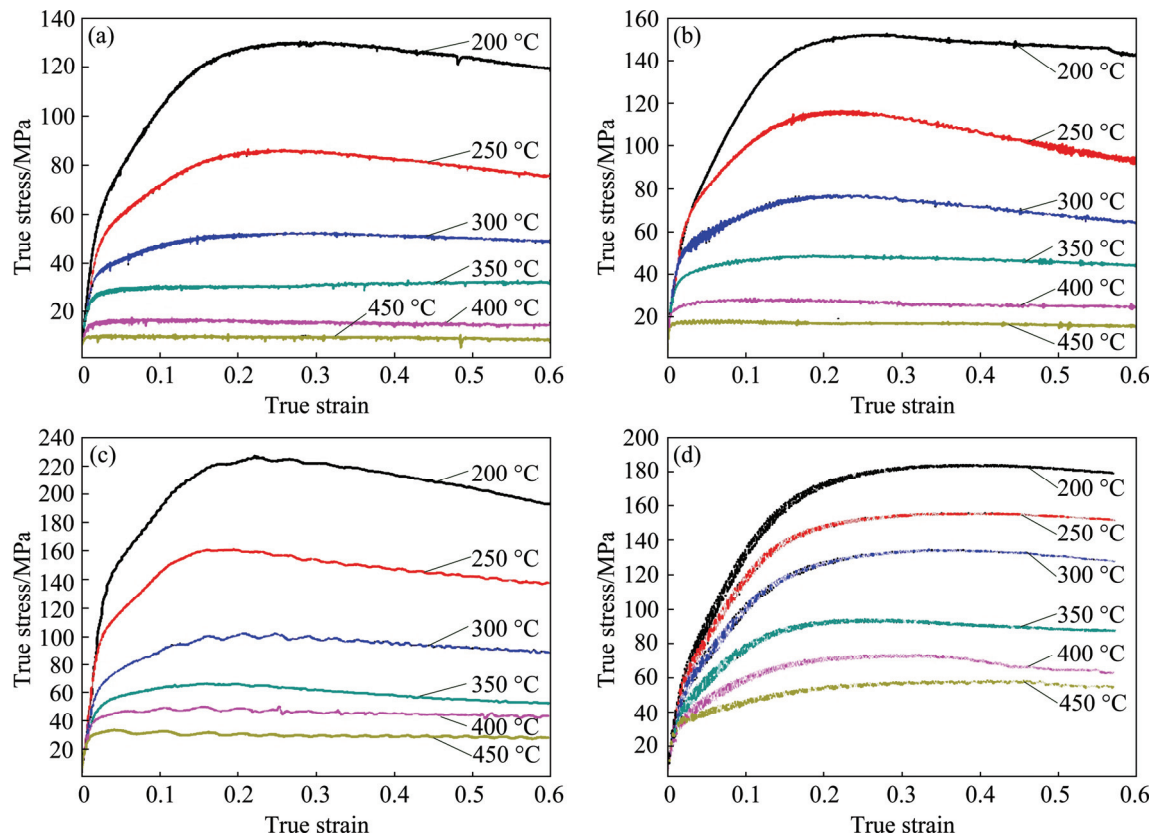


Fig. 1 True stress–strain curves of as-solution treated alloy A during hot compression at different strain rates: (a) 0.001 s^{-1} ; (b) 0.01 s^{-1} ; (c) 0.1 s^{-1} ; (d) 1 s^{-1}

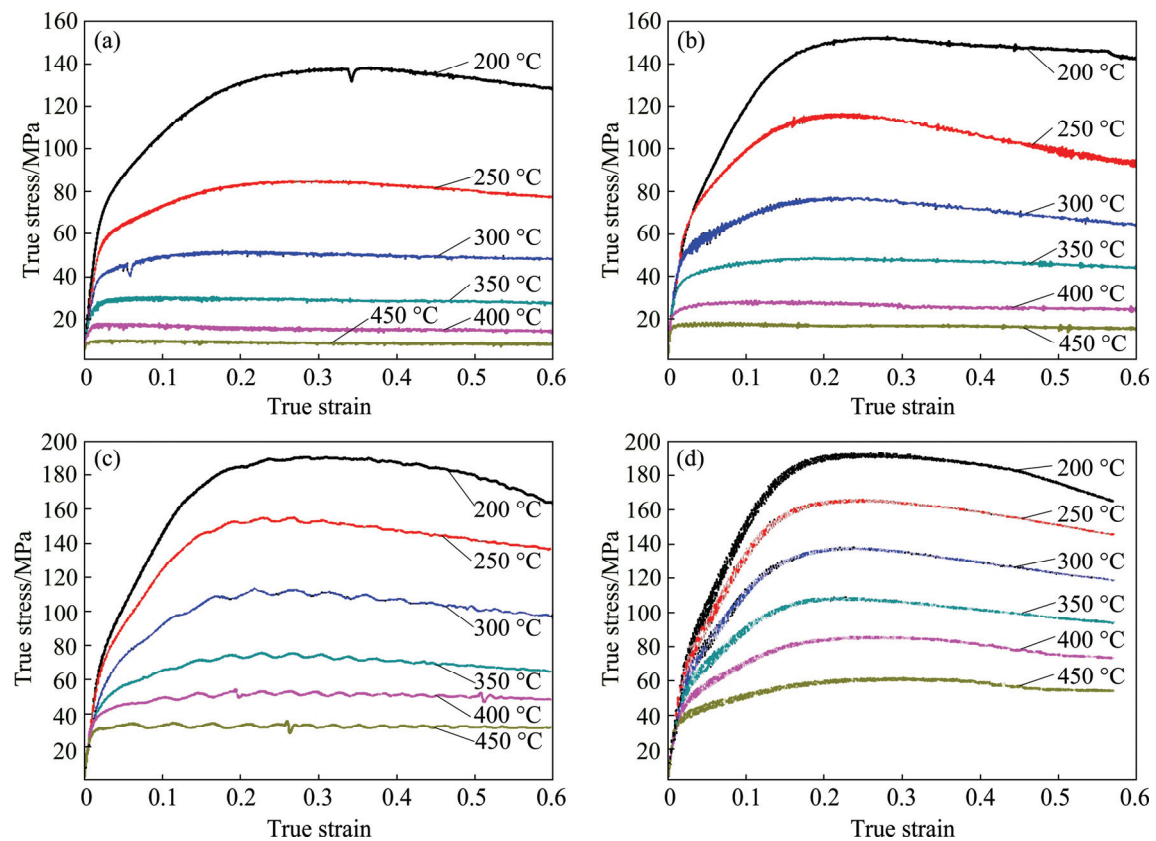


Fig. 2 True stress–strain curves of as-solution treated alloy B during hot compression at different strain rates: (a) 0.001 s^{-1} ; (b) 0.01 s^{-1} ; (c) 0.1 s^{-1} ; (d) 1 s^{-1}

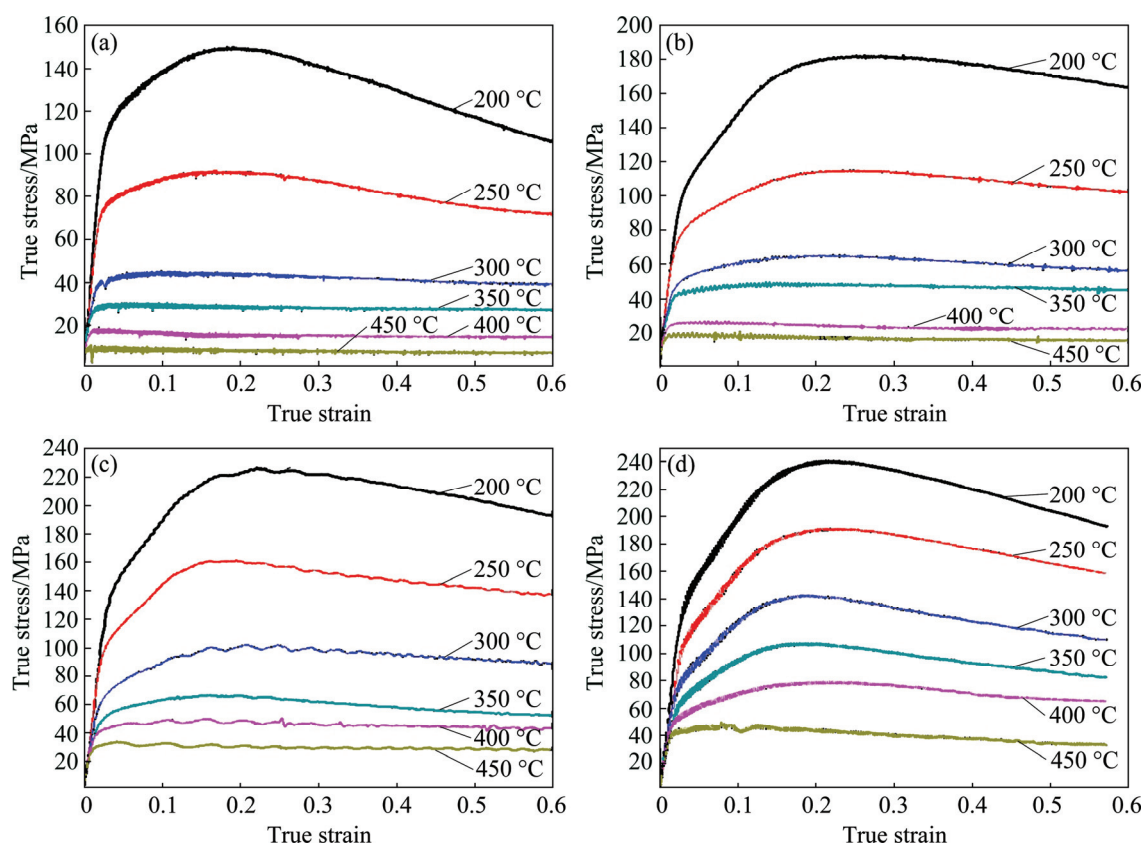


Fig. 3 True stress–strain curves of as-solution treated alloy C during hot compression at different strain rates: (a) 0.001 s^{-1} ; (b) 0.01 s^{-1} ; (c) 0.1 s^{-1} ; (d) 1 s^{-1}

the deformation temperature increased to $450 \text{ }^{\circ}\text{C}$, the amounts of the recrystallized grains could be found in the alloy C which promoted the material work softening (as shown in Figs. 4(c), (f), (i) and (l), and the stress–strain curves could reach the stable state at lower strain. However, it was not difficult to find that the unrecrystallized areas increased when the strain rate increased (as suggested by Figs. 4(c), (f), (i) and (l). Besides, as the strain rate increased, the density of dislocations increased sharply and some dislocations could not offset and integrate, which produced the stress concentration. The stress at stable state increased with the increase in the strain rate.

Figure 5 indicates that the peak stresses of alloys A, B and C vary with different strains rate at constant deformation temperature. It was obvious that the peak stresses increased with increasing the volume fraction of alloying element at $200 \text{ }^{\circ}\text{C}$, which was due to the formation of the twins during hot deformation. The studies have stated that $\{0001\}$ basal slip and $\{10\bar{1}2\}$ twinning were the main deformation mechanisms of magnesium alloys during compression at room and lower temperatures [23,24]. When the deformation temperature was $200 \text{ }^{\circ}\text{C}$, the density of twins in alloy C was higher than those in alloys A and B (as shown in Figs. 6(a), (d), (g)). Especially, twin boundary reduced the mean free

path of dislocation, which resulted in working hardening. However, when the deformation temperature increased to $350 \text{ }^{\circ}\text{C}$, the disparity of peak stress among the three alloys was not obvious, because the occurrence of the DRX resulted in the reduction of stress concentration (as shown in Figs. 6(b), (e), (h)). Moreover, the peak stress of alloy C became the minimum when the temperature was up to $450 \text{ }^{\circ}\text{C}$, which was attributed to the higher volume fraction of DRX grains in alloy C (as shown in Figs. 6(c), (f), (i)). In addition, with increasing in the mass ratio of zinc to yttrium ($m(\text{Zn})/m(\text{Y})$), the stacking fault energy (SFE) of the alloy decreased, which promoted the occurrence of DRX [25].

3.2 Constitutive equations

According to true stress–strain curves, the flow stress was influenced by deformation temperature, strain rate and strain, respectively. In order to understand the relationship among flows stress (σ), deformation temperature (T) and strain rate ($\dot{\epsilon}$), it is necessary to study the constitutive equations. The stress–strain data obtained from isothermal compression tests were used to determine the material constants of the constitutive equation. Especially, alloy C was chosen as an example to illustrate the process of constructing constitutive equations in the present research.

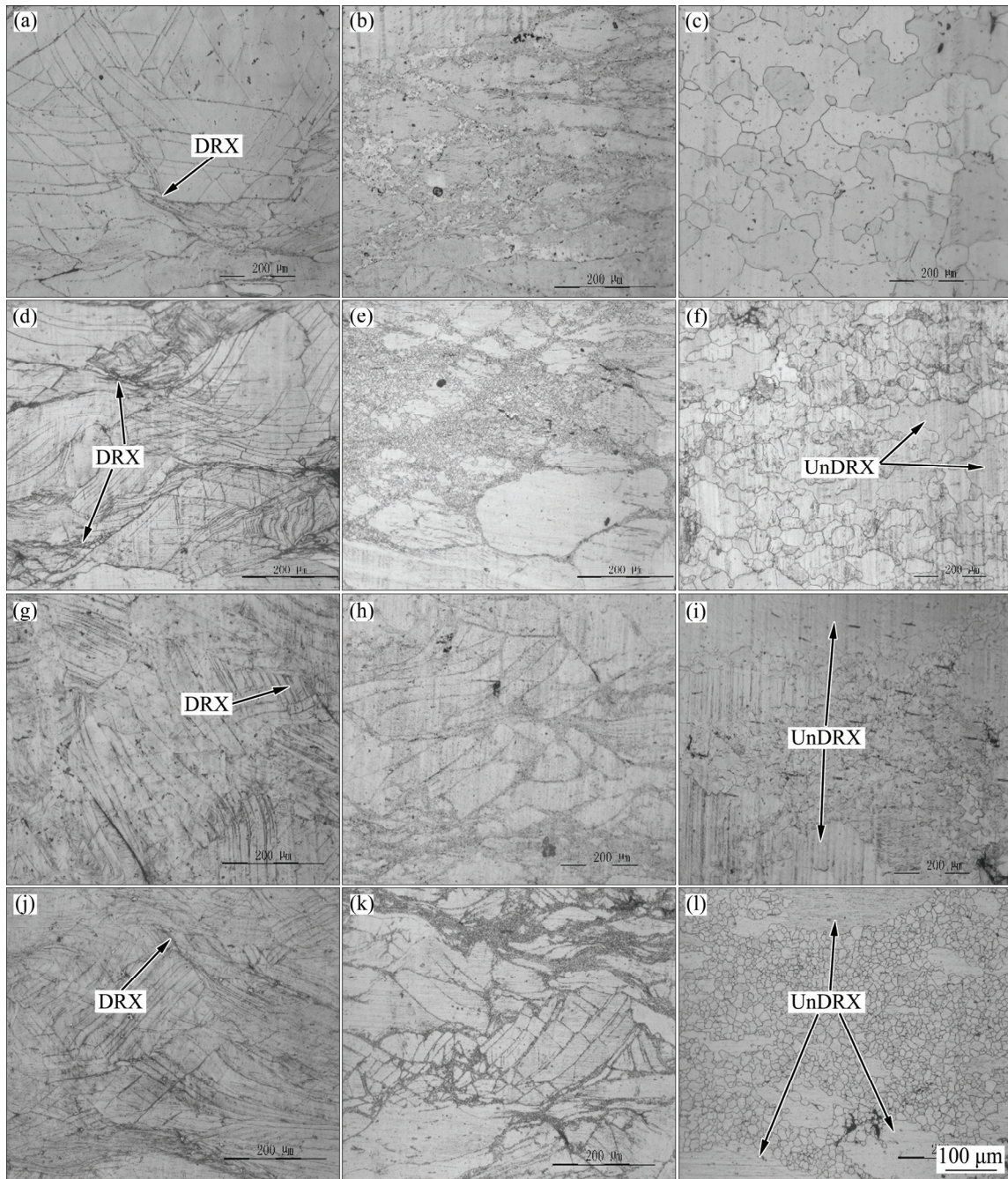


Fig. 4 OM images of as-compressed alloy C: (a) 200 °C, 0.001 s⁻¹; (b) 300 °C, 0.001 s⁻¹; (c) 450 °C, 0.001 s⁻¹; (d) 200 °C, 0.01 s⁻¹; (e) 300 °C, 0.01 s⁻¹; (f) 450 °C, 0.01 s⁻¹; (g) 200 °C, 0.1 s⁻¹; (h) 300 °C, 0.1 s⁻¹; (i) 450 °C, 0.1 s⁻¹; (j) 200 °C, 1 s⁻¹; (k) 300 °C, 1 s⁻¹; (l) 450 °C, 1 s⁻¹

The relationship among strain rate, stress and temperature varied with different materials treated by hot deformation. The flow behavior is usually described in the following three equations at different stress levels [26,27].

1) At low stress levels, the metal flow process can be described by power law:

$$\dot{\varepsilon} = A_1 \sigma^{n_1} \exp[-Q/(RT)], \alpha \sigma < 0.8 \quad (1)$$

2) At high stress levels, the metal flow process can

be described by exponential model:

$$\dot{\varepsilon} = A_2 \exp(\beta \sigma) \exp[Q/(RT)], \alpha \sigma > 1.2 \quad (2)$$

3) At all the stress levels, SELLARS and TEGART introduced a hyperbolic sine constitutive equation:

$$\dot{\varepsilon} = A[\sinh(\alpha \sigma)]^n \exp[Q/(RT)] \quad (3)$$

where $\dot{\varepsilon}$ is the strain rate (s⁻¹), Q is the activation energy for deformation (kJ/mol), R is the mole gas constant

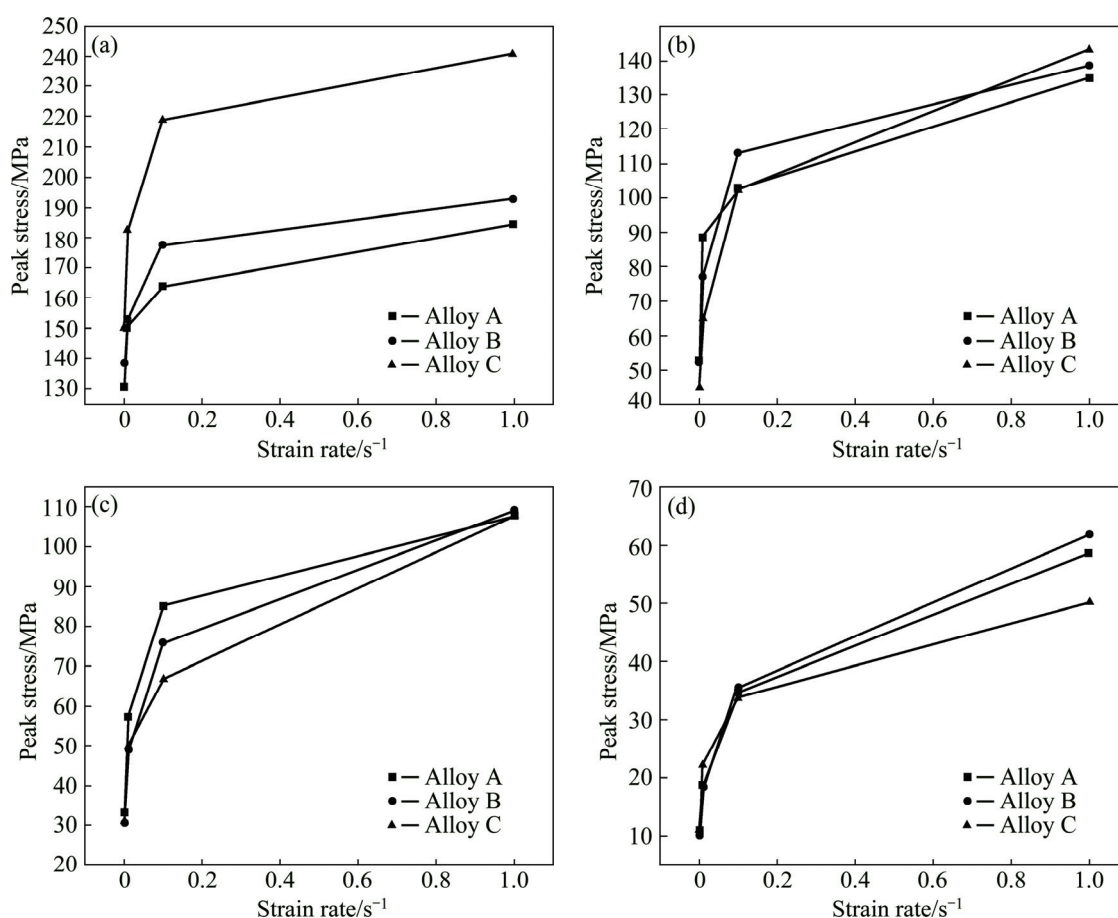


Fig. 5 Peak stresses of alloys A, B and C with different strain rates at different deformation temperatures: (a) 200 °C; (b) 300 °C; (c) 350 °C; (d) 450 °C

(8.314 J/(mol·K)), T is the thermodynamic deformation temperature (K), σ is the peak stress (MPa). The A , α , β and n_1 are the material constants with $\alpha = \beta/n_1$.

Figure 7 shows the linear relationship of $\ln \dot{\epsilon} - \ln \sigma$ and $\ln \dot{\epsilon} - \sigma$ at different temperatures obtained by linear fitting. The values of n_1 and β were determined by measuring the slopes of the $\ln \dot{\epsilon} - \ln \sigma$ and $\ln \dot{\epsilon} - \sigma$ at each temperature, respectively. Subsequently, the average values of n_1 and β were calculated, i.e., $n_1 = 7.329$, $\beta = 0.097$. Finally, the α value of alloy was calculated as $\alpha = \beta/n_1 = 0.013187$.

It is well known that the metal hot working process is controlled by activation energy. The deformation activation energy Q can be determined by Eq. (4).

$$Q = R \left\{ \frac{\partial \ln \dot{\epsilon}}{\partial \ln [\sinh(\alpha\sigma)]} \right\}_T \left\{ \frac{\partial \ln [\sinh(\alpha\sigma)]}{\partial (1/T)} \right\}_{\dot{\epsilon}} = RNS \quad (4)$$

The values of N and S can be obtained from the slopes of $\ln \dot{\epsilon} - \ln [\sinh(\alpha\sigma)]$ and $\ln [\sinh(\alpha\sigma)] - 10^3/T$ plots (as seen in Fig. 8). Then, substituting the values of both N and S into Eq. (4), it was therefore to get the average value of the deformation activation energy (Q) for alloy

C, i.e., 154.36 kJ/mol.

The combined effect of temperature and strain rate on the flow stress behavior can be also described by the Zener–Hollomon parameter at all stress levels [27] (as Eq. (5)).

$$Z = \dot{\epsilon} \exp[Q/(RT)] = A [\sinh(\alpha\sigma)]^n \quad (5)$$

Z values were calculated by substituting the values of Q under different hot deformation conditions into Eq. (5). The linear relationship of $\ln Z - \ln [\sinh(\alpha\sigma)]$ is shown in Fig. 9. The values of n and A were determined, i.e., 4.56595 and 5.9233×10^{11} , respectively.

The material constants of alloys A and B were also obtained by the same methods, and the values of which are listed in Table 2. The deformation activation energy is not only an important physical parameter, standing for the difficulty degree of the deformation during plasticity working, but also reasonably correlated with contents of solute elements [28]. Table 2 demonstrates that the values of Q for alloys A, B and C are 159.98, 151.52 and 154.36 kJ/mol, respectively, which suggests that alloys B and C are easier to be deformed than alloy A. The magnesium alloy is the material with a low SFE, which

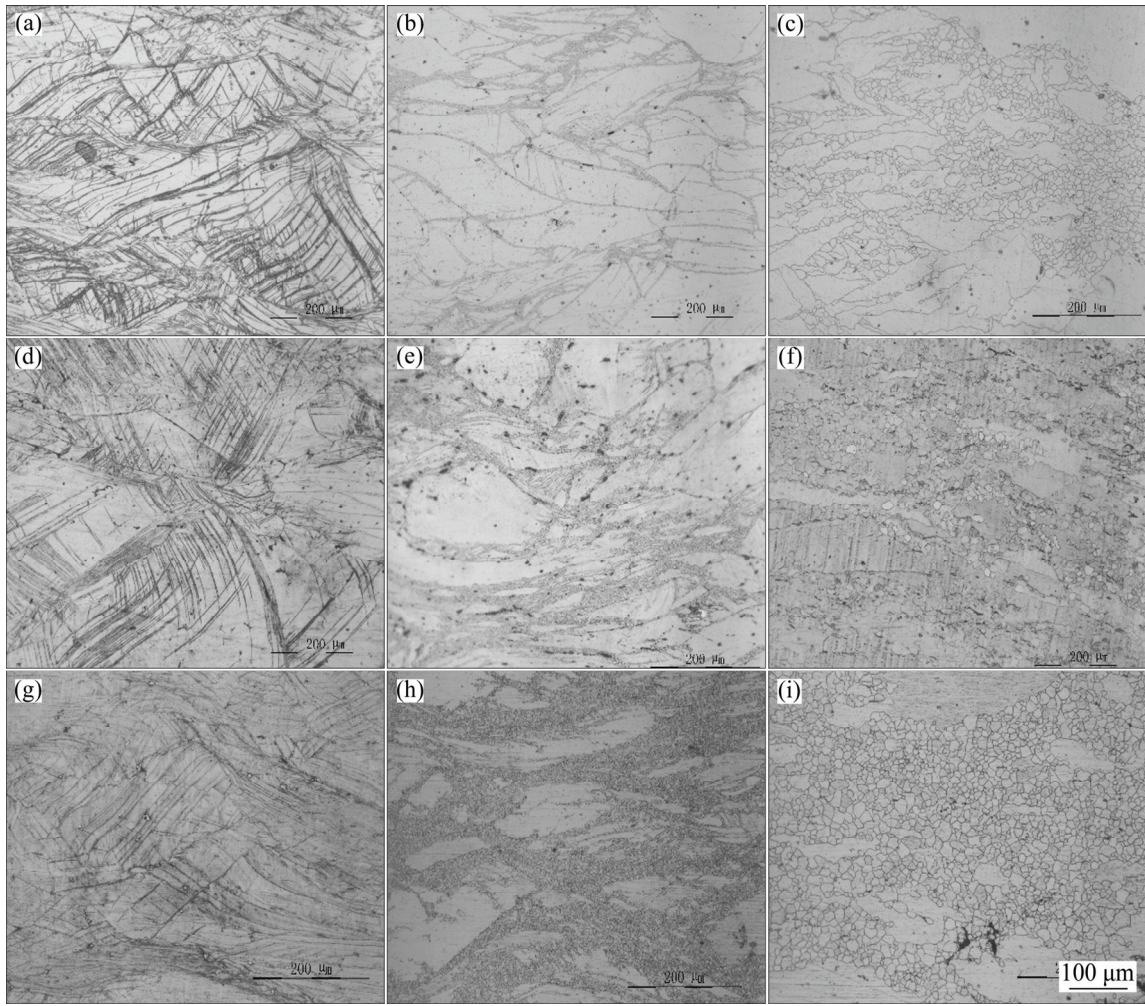


Fig. 6 OM images of as-compressed alloys with strain rate of 1 s^{-1} : (a) Alloy A, 200 °C; (b) Alloy A, 350 °C; (c) Alloy A, 450 °C; (d) Alloy B, 200 °C; (e) Alloy B, 350 °C; (f) Alloy B, 450 °C; (g) Alloy C, 200 °C; (h) Alloy C, 350 °C; (i) Alloy C, 450 °C

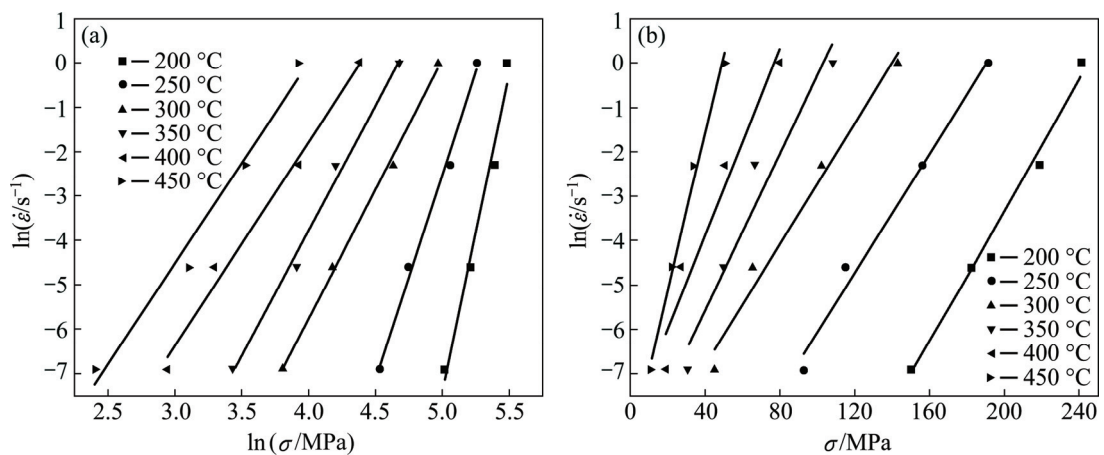


Fig. 7 Relationship between flow stress and strain rate for alloy C during hot compression: (a) $\ln \dot{\epsilon} - \ln \sigma$; (b) $\ln \dot{\epsilon} - \sigma$

results in the activation of the DRX. The addition of the Zn and/or Er can further reduce the SFE of the present magnesium alloy [25]. As a result, the degree of the DRX increased as the contents of Zn and Er increased. Thus, the workability of the present alloys increased with the contents of Zn and Er.

Table 2 shows that the correlation coefficients (R_1^2) of $\ln Z - \ln[\sinh(\alpha\sigma)]$ for alloys A, B and C were 0.9494, 0.9617 and 0.9936, respectively, demonstrating that the flow stress and deformation conditions of alloys A, B and C satisfied the equations of the three alloys during hot working process were hyperbolic sine relationship.

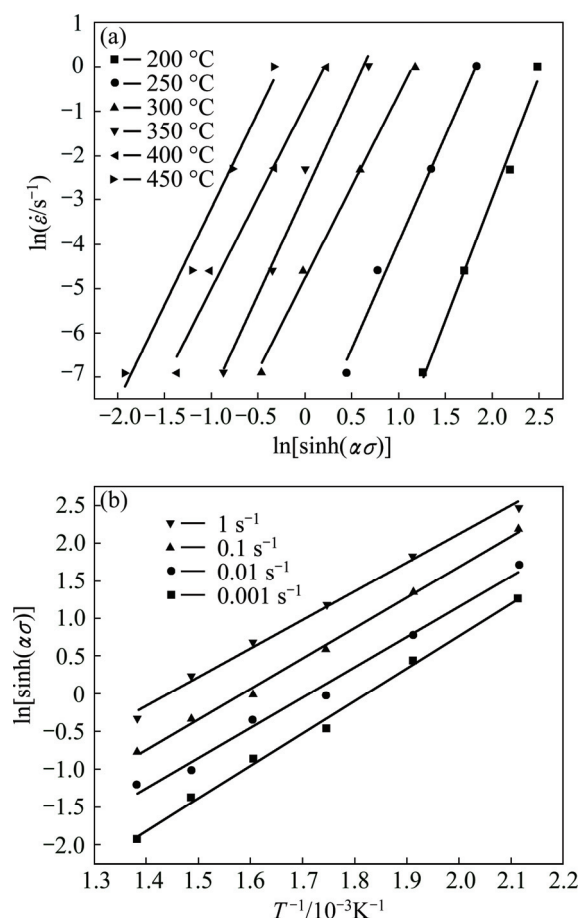


Fig. 8 Relationship of $\ln \dot{\epsilon} - \ln[\sinh(\alpha\sigma)]$ (a) and $\ln[\sinh(\alpha\sigma)] - 1000/T$ (b) for alloy C during hot compression

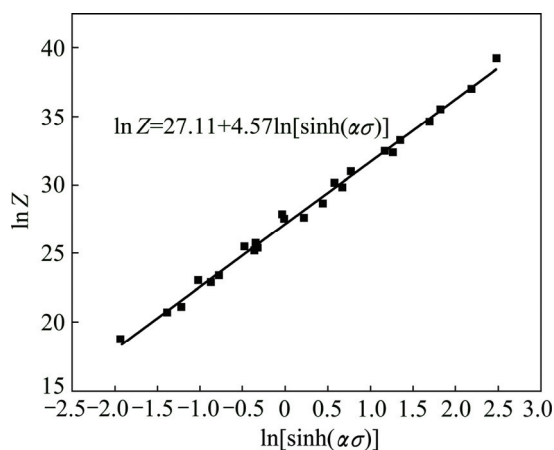


Fig. 9 Relationship of $\ln Z - \ln[\sinh(\alpha\sigma)]$ for alloy C during hot compression

Table 2 Material constants in constitutive equations of alloys A, B and C at peak stress

Alloy	n_1	β' MPa ⁻¹	α' MPa ⁻¹	Q' (kJ·mol ⁻¹)	$\ln A$	n	R_1^2
A	9.135	0.111	0.0122	159.98	29.11	5.638	0.9494
B	8.357	0.099	0.0117	151.52	27.53	5.204	0.9617
C	7.329	0.097	0.0132	154.36	27.11	4.566	0.9936

Therefore, the constitutive equations determined by Eq. (3) and (5) are as follows:

For alloy A

$$\sigma_p = 81.84 \ln \left\{ (Z/4.369 \times 10^{12})^{1/5.64} + [(Z/4.369 \times 10^{12})^{2/5.64} + 1]^{1/2} \right\},$$

$$Z = \dot{\epsilon} \exp[159980/(RT)];$$

For alloy B

$$\sigma_p = 84.80 \ln \left\{ (Z/9.076 \times 10^{12})^{1/5.2} + [(Z/9.076 \times 10^{12})^{2/5.2} + 1]^{1/2} \right\},$$

$$Z = \dot{\epsilon} \exp[151520/(RT)];$$

For alloy C

$$\sigma_p = 75.83 \ln \left\{ (Z/5.923 \times 10^{11})^{1/4.57} + [(Z/5.923 \times 10^{11})^{2/4.57} + 1]^{1/2} \right\},$$

$$Z = \dot{\epsilon} \exp[154360/(RT)].$$

In order to test the predictive accuracy of the new developed constitutive equations for the alloys, a comparison between the experimental and predicted peak stresses under different deformation conditions is shown in Fig. 10. It suggested that the relationship between the experimental and predicted values was linear, and the values of the correlation index R_1^2 were 0.951 (alloy A), 0.960 (alloy B) and 0.994 (alloy C), respectively. The higher R_1^2 values revealed a high accuracy of the predicted data from the developed constitutive equation. In other words, the newly developed constitutive equations could forecast the flow stress of the present investigated alloys during hot deformation in a wide strain rate range of 0.001–1 s⁻¹ accurately.

3.3 Processing map

The processing map is usually made on the basis of the principles of dynamical material model (DMM) [29], which is superposition of power dissipation map and instability map with the related parameters of strain rate and temperature. The efficiency of power dissipation (η) can be calculated by Eq. (6), which describes the hot deformation behavior of metals that undergo dynamic recovery (DRV) and DRX on Gleeble hot compression simulation [30].

$$\eta = \frac{2m}{m+1} \quad (6)$$

where m stands for the strain rate sensitivity index, which can be calculated by the following equation at the deformation temperature and strain rate.

$$m = (\partial \ln \sigma) / (\partial \ln \dot{\epsilon}) \quad (7)$$

The high efficiency of power dissipation (η) cannot stand for the ability of workability of materials because it may be present in the instability region of the fixed material. It is important to acquire the instability region

by using the instability criterion [29]. The instability criterion was shown as follows:

$$\xi_p(\dot{\epsilon}) = \frac{\partial \ln \frac{m}{m+1}}{\partial \ln \dot{\epsilon}} + m < 0 \quad (8)$$

When the value of $\xi_p(\dot{\epsilon})$ is negative, the deformation of the material is possible to flow instability because of the adiabatic shear bands or flow localization. The higher absolute value of $\xi_p(\dot{\epsilon})$ just leads to higher possibility of flow instability. The values of $\xi_p(\dot{\epsilon})$ are obtained by Eq. (8).

Considering the efficiency of power dissipation (η)

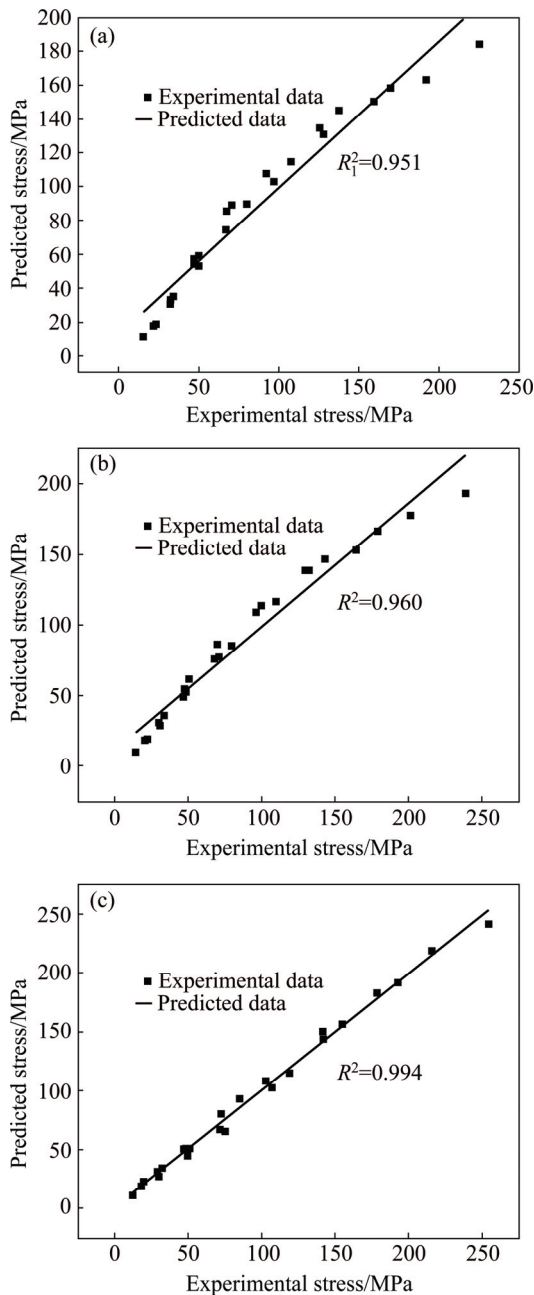


Fig. 10 Comparison between experimental and predicted stresses during hot compression: (a) Alloy A; (b) Alloy B; (c) Alloy C

and instability criterion (ξ_p), the processing maps of alloys A, B and C were constructed during hot compression in the temperature range of 300–450 °C with strain rates of 0.001–1 s⁻¹ after hot compression to the strains of 0.1, 0.3 and 0.6, as shown in Fig. 11. It was found that the instability domain decreased with increasing strain for alloy A. However, the instability domain increased with the increase in the strain of alloys B and C. Comparing with alloys A and C, it was found that the instability domain of alloy B was minimal at a constant strain. It was indicated that the alloy B was most easy to be processed because of the addition of Zn and/or Er, which agreed well with the result of activation energy.

It is well known that the activation energy of plastic flow is attributed to the volume self-diffusion of dislocation climbing at high temperature more than 300 °C. As a result, the strain localization will activate the formation of the bulges of grain boundaries, and the DRX usually occurs via nucleation in the slip bands [30,31]. The magnesium alloy is the material with low stacking fault energy (SFE) because of the low SFE on the basal plane, which results in occurrence of the DRX during hot deformation processing. The additions of the Zn and Y have a significant effect on decreasing the SFE, and it will further activate the occurrence of the DRX [25]. Specially, the addition of RE (RE standing for rare earth) plays a key role in reducing the value of the SFE of the magnesium matrix, which promotes the occurrence of the DRX significantly [32,33]. Therefore, alloys B and C are easy to achieve the balance between the working hardening and the working softening during deformation, especially for alloy B.

The processing map was constructed to obtain the optimum working windows. It was reported that the typical efficiency values associated with the DRX were 35% to 45% [34]. The DRX was beneficial to the material deformation. Table 3 shows the peak dissipation efficiencies of alloys A, B and C at the strains of 0.1, 0.3 and 0.6, respectively. It was suggested that the values of the peak dissipation efficiency of the three alloys were similar. It was indicated that the effects of the composition and the strain on the peak dissipation efficiency were not obvious in the present investigation. According to the processing map, the optimal parameters for the present alloys could be concluded, as seen in Table 3. Taking the strain of 0.3 as an example, the optimum hot working window for alloy A was in the temperature range of 380–430 °C with a strain rate less than 0.1 s⁻¹; the optimum hot working window for alloy B was in the temperature range of 380–450 °C with strain rate of 0.01–0.1 s⁻¹; the optimum hot working window for alloy C was in the temperature range of 390–440 °C with strain rate of 0.01–0.1 s⁻¹.

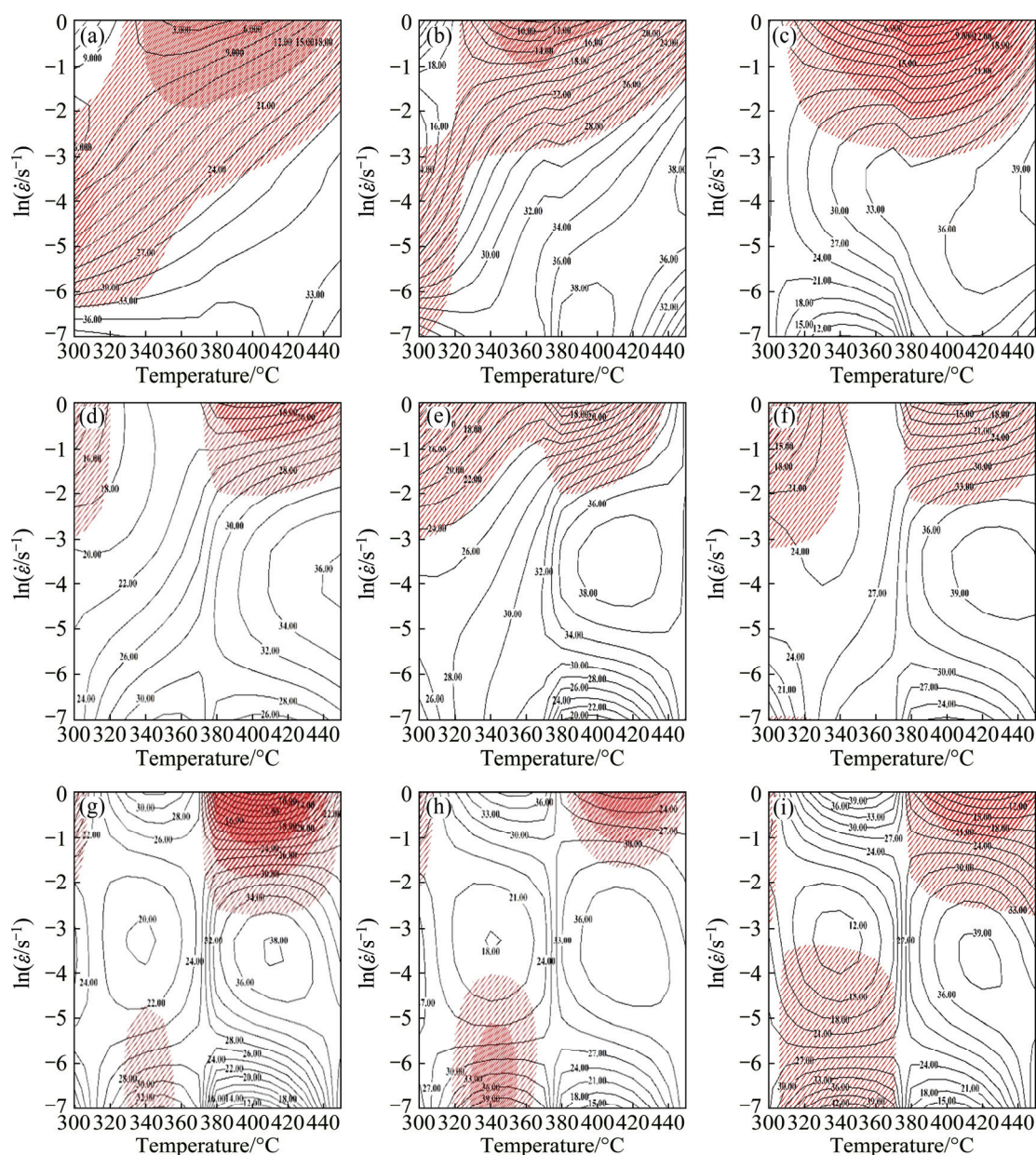


Fig. 11 Hot processing maps of Mg–Zn–Er alloys at different strains during hot deformation: (a) $\varepsilon=0.1$, alloy A; (b) $\varepsilon=0.3$, alloy A; (c) $\varepsilon=0.6$, alloy A; (d) $\varepsilon=0.1$, alloy B; (e) $\varepsilon=0.3$, alloy B; (f) $\varepsilon=0.6$, alloy B; (g) $\varepsilon=0.1$, alloy C; (h) $\varepsilon=0.3$, alloy C; (i) $\varepsilon=0.6$, alloy C

Table 3 peak dissipation efficiency and optimum hot working parameters for alloys A, B and C in strain range of 0.1–0.6

Alloy	$\eta_p/\%$	$t/^\circ\text{C}$	$\dot{\varepsilon}/\text{s}^{-1}$
A	36–39	380–450	0.001–0.01
B	36–39	380–450	0.001–0.1
C	36–39	380–450	0.01–0.1

4 Conclusions

1) The hot deformation behaviors of Mg–3Zn–0.5Er (alloy A), Mg–4.5Zn–0.75Er (alloy B) and Mg–6Zn–1Er (alloy C) alloys were investigated by hot compression test on a Gleeble–1500 thermal simulator.

2) The amounts of twins were present as the compensated factors to advance the total deformation as well as the basal slip below the temperature of 300 °C. The DRX occurred sharply as the temperature increased to 300 °C, especially at 450 °C, which led to the complete balance between the working hardening and working softening.

3) The constitutive equations of alloys A, B and C were developed. The high correlation coefficient (R_1^2) suggested that the flow stress could be calculated by the new developed constitutive equations accurately.

4) The deformation activation energy (Q) of alloy A was higher than those of alloys B and C because of the addition of Zn and/or Er, which had a great effect on

decreasing the stacking fault energy (SFE), as a result of a sharp DRX in alloys B and C.

5) The instability domain area of alloy B was the smallest among the studied alloys at the same strain on the basis of the present processing maps. The optimum working window of alloy B was determined at strain of 0.6, i.e., the strain rate from 0.001 to 0.1 s⁻¹ and the temperature from 390 to 450 °C.

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Mg-Zn-Er 合金基于 Gleeble-1500 热压缩模拟的变形行为及其加工图

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摘 要: 利用 Gleeble-1500 热模拟机研究固溶态 Mg- x Zn- y Er 合金($x/y=6$, $x=3.0, 4.5, 6.0$; $y=0.50, 0.75, 1.00$) 在变形温度为 200~450 °C、应变速率为 $0.001\sim 1\text{ s}^{-1}$ 下的热压缩变形行为。研究表明, 在热压缩变形过程中加工硬化和加工软化同时发生, 并相互竞争, 其中合金加工软化主要由动态再结晶引起。构建了 Mg-Zn-Er 合金的本构方程, 该本构方程能比较精确地预测合金的峰值应力。添加 Zn、Er 合金化元素致使 Mg-3Zn-0.5Er(合金 A)具有较高变形激活能。在温度和应变速率的二维平面内建立了合金的热加工图, 并提供了合金的最优加工条件(应变量为 0.3, Mg-3Zn-0.5Er 合金(合金 A): 380~430 °C, $<0.1\text{ s}^{-1}$; Mg-4.5Zn-0.75Er 合金(合金 B): 380~450 °C, $0.01\sim 0.1\text{ s}^{-1}$; Mg-6Zn-1Er 合金(合金 C): 390~440 °C, $0.01\sim 0.1\text{ s}^{-1}$)。与 Mg-3Zn-0.5Er(合金 A)和 Mg-6Zn-1Er(合金 C)相比, Mg-4.5Zn-0.75Er(合金 B)表现较优的热加工窗口。

关键词: Mg-Zn-Er 合金; 压缩行为; 显微组织演变; 本构方程; 加工图

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