Materials Research Express

PAPER

Study on rheological behavior and microstructural evolution of Al-6Mg-0.4Mn-0.15Sc-0.1Zr alloy by isothermal compression

Liyong Lu\(^1\), Feng Jiang\(^2\), Jiachen Liu\(^3\), Jianjun Zhang\(^4\), Mengmeng Tong\(^5\), Hongfeng Huang\(^6\), Pian Xu\(^1\) and Zhongqin Tang\(^1\)

1. Introduction

Aluminum alloys are basic materials of the aerospace, shipbuilding, automobile, and other high-tech industries. Compared with the other aluminum alloys, Al-Mg alloys possess excellent corrosion resistances, high extensibilities, and good weldabilities. They are widely used as structural materials in automobiles, aircraft, and low-temperature applications [1, 2]. Because Al-Mg alloys have only medium strengths, they do not meet the application requirements of rapidly developing models and equipment. To obtain better mechanical properties based on traditional Al-Mg alloys, a higher allowable strength was developed by increasing the Mg content and adding trace amounts of Mn, Sc, and Zr. This led to new type of Al-Mg-Mn-Sc-Zr alloy with better comprehensive properties [3, 4].

The composite addition of alloying elements can significantly improve the properties of the alloy. However, the increase in the content of alloying elements also makes the hot working plastic deformation of the alloy more difficult. Shear plastic instability occurs easily during the deformation process, and shear deformation bands are

Abstract

The effects of trace amounts of Sc/Zr on rheological properties and microstructural evolution of the Al-6Mg-0.4Mn-0.15Sc-0.1Zr alloy under different deformation conditions were studied by isothermal compression tests at deformation temperatures of 280 °C ÷ 460 °C and strain rates of 0.001 ÷ 10 s\(^{-1}\). A constitutive model based on the hyperbolic sine function was established. The dependence of the flow stress on strain, strain rate, and deformation temperature was described. Using experimental data and dynamic material model, machining diagrams of the alloy with strains of 0.3 and 0.5 were obtained, and the hot workability of the alloy was verified. The results showed that the deformation temperature and strain rate had significant effects on the deformation behavior of the alloy. The flow stress of the alloy increased with the increase in strain rate and decreased with the increase in deformation temperature. Under high-temperature and low-strain-rate conditions, the alloy tended to undergo dynamic recrystallization, and the volume fraction and grain size increased with the increase in deformation temperature and the decrease in strain rate. Based on the analysis of the microstructural evolution, the construction of the machining diagram, and the solution of the rheological constitutive equation, suitable values of the deformation temperature were determined to be 380 °C ÷ 460 °C. Moreover, suitable values of the strain rate were determined to be 0.001 and 0.3 s\(^{-1}\). The deformation activation energy of 161.28 kJ mol\(^{-1}\) was obtained.
formed. This affects the properties of the alloy [5]. The flow stress–strain curve of the alloy during hot deformation characterizes its hot deformation behavior. It also reflects the magnitude of load required and energy consumed during hot deformation.

The hot working diagram is based on the superposition of the energy and instability diagrams of the dynamic material model, which can describe the mechanism for the microstructural evolution of the material at different deformation temperatures and strain rates, providing a reference for the determination of the deformation process [6]. The flow stress obtained from hot compression experiment can be used to calculate the constitutive equation and hot processing map [7–9]. The flow stress of the alloy during hot deformation is the comprehensive expression of microstructural evolution of the alloy. The research on hot deformation behavior and hot processing map is a priority to design a suitable processing technology and achieve a good performance. At present, the research on Al-Mg alloys containing Sc has mainly focused on the mechanical properties, corrosion resistance, and superplastic properties [2, 10, 11]. The hot workability of the Al-Mg-Mn-Sc alloy with a high Sc content (Sc > 0.2 wt%) was studied by Wen [6] and Huang [12]. The results showed that the addition of Sc had a significant effect on the hot deformation behavior of Al-Mg alloys. Due to the issue of high cost, Al-Mg-Sc alloys with low Sc content (Sc < 0.2 wt%) have gradually become a hot spot of study. However, the rheological behavior and microstructural evolution of Al-Mg-Mn-Sc-Zr alloys with low Sc content during isothermal compression were rarely studied, which placed a vital role for these alloys to obtain a good overall performance.

For this reason, high-temperature compression deformation tests of a new type of alloy, Al-6Mg-0.4Mn-0.2 wt% Sc, was conducted. The high-temperature thermal deformation behavior of the alloy was studied. Moreover, a constitutive equation model and hot working diagrams were established. The microstructural characteristics of the alloy under different deformation conditions were discussed to provide a basis for the formulation of the hot working process of Al-Mg-Sc alloys with low Sc contents.

2. Materials and experimental methods

The experimental samples were taken from semi-continuous casting oblate ingot with the dimension of 300 mm × 1500 mm × 5000 mm cooled by circulating water. The nominal chemical composition is shown in table 1.

As shown in figure 1, samples were taken from typical positions of the ingot. The microstructures of samples of as-cast alloys are observed.

Table 1. Nominal composition of the alloy (wt%).

|   | Al  | Mg  | Mn  | Sc  | Zr  |
|---|-----|-----|-----|-----|-----|
| Balance | 6.0 | 0.40 | 0.15 | 0.10 |          |

As shown in figure 1, as the ingot is relatively big compared with other castings from most of the other research articles casted in the lab, the cooling speeds in different areas on cross-section perpendicular to the long axis of the ingot are different during cooling of the ingot casting. Therefore, the grain size and the amount of primary phase are all different between edge and center of the ingot. In the center of the ingot, the cooling speed is slow. The grain size of as-cast structure is the largest among other areas in the cross-section. The amount of primary phase in the center is large. Besides, defects are easy to form in the center of ingot. Hence, it is also easy to obtain defects during hot deformation. As a result, samples should be obtained in the center of the ingot to simulate a real industry condition.

Cylindrical compression specimens with diameters of 10 mm and heights of 15 mm were taken from the 1/2 center of the ingot, and the samples were uniformly treated at 350 °C for 8 h before the deformation.

Isothermal compression experiments of the specimens under tantalum lubrication condition on their two ends were carried out on Gleeble-3500 thermal simulator. It could precisely control the temperature, strain rate and strain. Resistance-type heating module was used.

The preliminary researches showed that the best hot deformation temperature range of Al-Mg-Mn alloy is 350 °C–460 °C. However, in this study, during the last period of hot deformation, the Al-Mg-Mn-Sc-Zr alloy experienced low temperature deformation to simulate the real industry production. Therefore, the low temperature experiment was set at 280 °C and 320 °C.

The range of effective strain rate used in the experiment was calculated by the deformation speed in industry production. The strain rate of hot deformation by calculation was in the range from 0.001 s⁻¹ to 10 s⁻¹, which was the experimental range in this study.

As a result, the deformation temperatures were 280, 320, 350, 420, and 460 °C, and the strain rates were 0.001, 0.01, 0.1, 1, and 10 s⁻¹.
Before testing, each sample was heated to a deformation temperature at a rate of 10 °C s⁻¹ and kept for 3 min to achieve a uniform temperature in the sample. After the sample was compressed to 50%, the sample was immediately quenched to maintain the deformed structure. According to maximum deformation ratio in industrial production, the compression ratio of 50% was chosen. Besides, the compression ratio of 50% could be achieved by Gleeble 3500 type dynamic thermal simulator with lubrication condition between compression sample and compression dies. The grain shape and size and grain boundary morphology of the alloy were observed using a Leica DFC metallographic microscope in polarized mode. After mechanical polishing, all the metallographic samples were electropolished with perchlorate + anhydrous ethanol (volume fraction 1:9) and anodized with phosphoric acid + sulfuric acid + distilled water (volume fraction 43:38:19). The microstructure, crystal structure, secondary phase morphology and interface characteristics of the alloy were observed using an FEI F20 transmission electron microscope (TEM) with an accelerating voltage of 200 kV. The TEM samples were first polished to thicknesses of 100 μm by water and metallographic sandpaper grinding, after which they were cut into small round pieces with diameters of 3 mm. The electrolyte was a mixture of methanol and nitric acid (4:1), and the thinning temperature was controlled at −30 to −20 °C.

3. Results and discussion

3.1. Determination of additive amount of Sc
Figures 2(a) and (b) exhibited the microstructures of the alloy samples. The samples were all taken from the ingots with the dimension of 300 mm × 1500 mm × 5000 mm under the same casting conditions. The only difference of these alloys was additive amount of Sc.

Figure 2(a) shows the microstructure of the alloy with no Sc content. The as-cast coarse structure is observed, which shows a typical as-cast dendritic structure characteristic. Dendrite spacing is about 200–400 μm. Figure 2(b) shows the alloy with 0.15% Sc content, which is the main research alloy in this study. The average grain size is about 45 μm. It is observed that dendritic structure is eliminated and equiaxed grains are distributed.

Figures 2(c) and (d) show the microstructure of the alloy with 0.25% Sc content. The equiaxed grains are also observed and the average grain size is about 90 μm, as shown in figure 2(d). However, primary phases in square shape with dimension of 20 μm are observed in the microstructure, which is confirmed to be Al₃(Sc, Zr) phase by EDS analysis, as shown in figure 2(c). When Al₃(Sc, Zr) phase in micron scale is formed, it is impossible to break or be eliminated by subsequent hot deformation process. Eventually, it plays a negative effect on overall performance of the alloy. Therefore, during the casting of the alloy, appropriate additive amount of Sc is beneficial to grain refinement of as-cast aluminum alloy.

3.2. Initial microstructure
The microstructure of the alloy after the uniform heat treatment of the ingot is shown in figure 3 The optical microscopy (OM) results show that the grains of the alloy is uniformly distributed, and the average grain size is
Figure 2. Microstructures of alloys with (or without) different additive amount of Sc. (a) Al-6Mg-0.4Mn-0.10Zr; (b) Al-6Mg-0.4Mn-0.15Sc-0.10Zr; (c) and (d) Al-6Mg-0.4Mn-0.25Sc-0.10Zr.

Figure 3. Initial microstructure of as-homogenized Al-6Mg-0.4Mn-0.15Sc-0.1Zr alloy prior to deformation: (a) OM; (b) SEM and EDX of black phases marked in (b); (c) bright field and (d) corresponding dark field TEM images of Al₃(Sc, Zr) particles.
about 45 μm, shown in figure 3(a). The significant decrease in the grain size is due to the heterogeneous nucleation of Sc and Zr elements added in the melting and casting processes. This produces a refinement effect on the microstructure [13].

The TEM analysis show that there are two kinds of precipitated particles distributed in the grains. The first kind of precipitated phase consists of long and rod-shaped particles with an average length of 70 ~ 90 nm. The particles are relatively small and sparsely distributed. The energy spectrum analysis confirms that this kind of precipitated phase was an Al-Mn phase with a high manganese content, shown in figure 3(b). Based on the morphology of precipitated phase and the Al/Mn content, the particles in this precipitated phase are Al₄Mn/Al₅Mn particles [14]. The second kind of precipitated phase is composed of nanometer-sized spherical particles, which are numerous and dispersed. Superlattice diffraction spots are observed by selected area diffraction analysis along the [111] crystal shear band. Thus, the particles of this precipitated phase are Al₅(Sc, Zr) particles with a coherent relationship with the aluminum matrix, shown in figures 2(c), (d). The nanometer-sized Al₅(Sc, Zr) particles mainly precipitated during the uniform heat treatment of the alloy, possess good thermal stabilities, and play a significant role in improving the comprehensive properties of the Al-Mg alloy.

3.3. True stress-strain curves

The uniform alloy samples are compressed at constant temperature under different temperature and strain rate conditions, the changes of the stress and strain are monitored in real time during the experiment, and the true stress-strain curves of the alloy under different deformation conditions are obtained based on the experimental data. As shown in figure 4, the flow stress-strain curve of the alloy under various deformation conditions is mainly divided into two stages. In the initial stage of hot compression, the flow stress increases rapidly with the increase in strain. After reaching a certain compression strain (0.02 ~ 0.1), the flow stress reaches a maximum, after which the flow stress decreases to a certain extent. However, the flow stress remains stable with the increase in the strain. The stress-strain curve tends toward a horizontal stable state.

The variation of the flow stress with strain is the result of the competition between work hardening and softening in the hot deformation process, which indirectly reflects the different deformation mechanisms of the alloy at different deformation stages. In the initial stage of hot compression, the external force causes significant dislocation proliferation in the alloy, and dislocation entanglement and accumulation occur in the movement process, resulting in a significant work hardening effect. The Al-Mn phase and dispersed Al₅(Sc, Zr) particles in the alloy matrix effectively hinder the movement of the dislocations and make the work hardening effect more evident, and the work hardening effect is dominant. The flow stress increases significantly with the increase in deformation. With the progress of hot compression deformation, the deformation energy storage and external heat input increase the driving force of dynamic crystallization, a large number of dislocations rearrange and are offset by sliding and climbing, and the work hardening of the matrix is gradually weakened. The effect of dynamic softening increases gradually. When the flow stress reaches a peak value, with the development of dynamic recrystallization, the effects of work hardening and dynamic softening are in a state of relative equilibrium, and the flow stress basically remains at the same level with the increase in strain. At this time, the alloy enters a stable plastic deformation stage [15, 16].

The effects of the deformation temperature and strain rate on the peak flow stress of the experimental alloy are shown in figure 4(f). Overall, the true stress-strain curve shows a similar trend, i.e., the flow stress increases with the increase in strain rate. In addition, it decreases with the increase in deformation temperature.

3.4. Establishment of processing maps

The plastic deformation process of the material is accompanied by a change in energy. The energy $P$ of the external action on the material is composed of a plastic deformation dissipative energy $G$ and microstructure evolution dissipative energy $J$. The total power $P$ can be described as follows:

$$P = \sigma \dot{\varepsilon} = G + J = \int_0^\sigma \sigma \mathrm{d}\varepsilon + \int_0^\sigma \varepsilon \mathrm{d}\sigma$$

(1)

where $\sigma$ is the flow stress (MPa) and $\dot{\varepsilon}$ is the strain rate (s⁻¹). At a certain deformation temperature and strain, the distribution of the energy dissipation for these two modes is assigned by the strain rate sensitivity index $m$, which can be expressed as follows:

$$m = \left( \frac{\partial J}{\partial G} \right)_{e,T} = \frac{\varepsilon \mathrm{d}\sigma}{\sigma \mathrm{d}\varepsilon} = \left( \frac{\partial \ln \sigma}{\partial \ln \varepsilon} \right)_{e,T}$$

(2)

The dissipative energy $J$ is closely related to the evolution of microstructure, and the deformation of the alloy is assumed to be in an ideal linear dissipative state:
When dynamic recovery occurs in the actual deformation process, the microstructural evolution, such as recrystallization and phase transformation, will lead to the change of dissipation $J$. At this time, $m = 1$, but it is in the range of $0 \sim 1$. The ratio of energy dissipation to the ideal linear dissipation caused by the microstructural evolution is the power dissipation efficiency factor:

$$J_{\text{max}} = \int_{0}^{\sigma} \dot{\varepsilon} d\sigma = \frac{m \sigma^{2}}{1 + m} = \frac{\sigma^{2}}{2} \quad (m = 1) \quad (3)$$

Because $\eta$ is sensitive to the microstructural evolution mechanism of the alloy during deformation, the change in the hot deformation mechanism of the alloy can be judged by the change of $\eta$. The machining power dissipation efficiency diagram can be obtained by drawing the $\eta$ isoline on the plane with the strain rate and deformation temperature as the coordinates. Usually, the processing parameters corresponding to higher $\eta$ values are more favorable to the stable plastic deformation of the alloy. To further eliminate the condition of plastic instability deformation, the instability criterion established by Prasad [17] is used to define the parameters of plastic instability deformation:

$$\eta = \frac{J}{J_{\text{max}}} = \frac{2m}{m + 1} \quad (4)$$

**Figure 4.** Typical true stress-strain curves of the Al-6Mg-0.4Mn-0.15Sc-0.1Zr alloy deformed at (a) 0.001 s$^{-1}$, (b) 0.01 s$^{-1}$, (c) 0.1 s$^{-1}$, (d) 1 s$^{-1}$, (e) 10 s$^{-1}$, and (f) the peak flow stress of corresponding specimens.
The machining instability diagram can be obtained by drawing the Zeta ($\varepsilon < 0$) isoline on the plane with strain rate and deformation temperature as coordinates. The plastic deformation of the alloy is not ideal during deformation in the unstable region. By combining the machining instability diagram with the machining power dissipation efficiency diagram, the hot working diagram, which can be used to predict the optimal plastic deformation process of the alloy, can be obtained.

Based on the basic theory for constructing a machining diagram and the stress-strain data of the alloy under different conditions, shown in figure 4, the effect of the strain on the hot deformation behavior of the alloy is considered. The hot working diagram of the experimental alloy is constructed when the strain was 0.3 and 0.5. The number on the color isoline represents the magnitude of the power dissipation factor $\eta$, and the gray region is the area of rheological instability, shown in figure 5. As shown in figure 5(a), when the strain is 0.3, there are regions of high power dissipation efficiency at the strain rate of $0.003 \text{ s}^{-1}$, temperature of $320 \degree C$ and strain rate of $0.03 \text{ s}^{-1}$, temperature of $440 \degree C$, and the peak power dissipation efficiency is about 40%. Based on the grey area in the diagram, the peak value of the power dissipation efficiency is approximately 40% at the strain of 0.3, temperature of $320 \degree C$, and the strain rate of $0.03 \text{ s}^{-1}$. At this time, the hot working instability region of the alloy is mainly distributed in the low-strain-rate and high-deformation-temperature ranges, while the plastic instability tends to occur for low values of the power dissipation factor $\eta$. Furthermore, in the process of isothermal compression, high values of the power dissipation factor are mainly distributed in the low-strain-rate and high-deformation-temperature ranges.

### 3.5. Microstructural evolution

The change of the deformation mechanism of the alloy is indirectly reflected by the change of the power dissipation factor $\eta$ and the plastic instability region shown in the machining diagram during plastic deformation. To explore the plastic deformation behavior of the tested alloy, the samples with a high power dissipation factor $\eta$ and plastic instability are selected for comparative analysis of the microstructure, as shown in figures 6 and 7. From the results of the OM microstructure analysis shown in figure 6, the microstructural deformation uniformity of the alloy is sensitive to the strain rate. The intragranular shear band is formed in some grains at low-temperature and high-strain rate conditions ($\geq 1.0 \text{ s}^{-1}$). When the strain is increased to 0.5, the power dissipation efficiency factor in the region close to temperature of $440 \degree C$ further increases to 43%, and the machining instability zone is also significantly expanded under the conditions of a high strain rate and low deformation temperature. In general, the high-value region of the power dissipation factor $\eta$ of the experimental alloy during isothermal compression deformation is mainly distributed in the low-strain-rate and high-deformation-temperature ranges, while the plastic instability tends to occur for low values of the power dissipation factor $\eta$. Furthermore, in the process of isothermal compression, high values of the power dissipation factor are mainly distributed in the low-strain-rate and high-deformation-temperature ranges.

![Processing maps of Al-6Mg-0.4Mn-0.15Sc-0.1Zr alloy under strains of (a) 0.3 and (b) 0.5.](image)

\[
\xi(\dot{\varepsilon}) = \frac{\partial \ln \left( \frac{m}{m+1} \right)}{\partial \ln \dot{\varepsilon}} + m < 0
\]
distributed in pancake shape, and the degree of non-uniform deformation of the alloy is aggravated, shown in figure 6(c). When the strain rate is reduced to 0.001 s\(^{-1}\), the pancake deformation structure is disappeared and the grain deformation is uniform, and a large number of fine new grains are observed at the deformed grain boundary. Typical characteristics of a dynamic recrystallizing structure is shown in figure 6(d).

There are a large number of Al-Mn phase and Al\(_3\)(Sc, Zr) particles in the alloy matrix shown in figure 3. In the process of deformation, these precipitates exert an important influence on the formation of the shear band and dynamic recrystallization by interacting with dislocations. Based on the processing diagram and OM microstructure analysis, for a strain rate of 0.1 s\(^{-1}\), the alloy transitions from a plastic instability zone to a stable deformation zone, and the power dissipation factor \(\eta\) increases gradually, shown in figure 5(b). The microstructural evolution is representative, and the deformation substructure characteristics of the alloy samples with different deformation temperatures at this strain rate are observed by TEM, as shown in figure 7.

As shown in figure 7(a), the Al-Mn phase and Al\(_3\)(Sc, Zr) particles distributed in the matrix at low temperatures increase the resistance to dislocation movement and significantly increase the dislocation density in the deformed matrix. Many dislocation entanglements form a straight dislocation boundary. The grains are divided by the dislocation boundary and form a deformed microstrip. The dislocations in the microstrip can not cross the dislocation boundary, and entanglement occurs in the microstrip. Due to the high-density dislocation entanglement in the local region, the deformation energy storage and deformation heat energy in this region increase, and the formation of intragranular shear band is induced, shown in figure 6(a), which lead to shear plastic instability deformation. At a relatively high deformation temperature (320 \(^{\circ}\)C), the high deformation temperature provides energy for the evolution of the microstructure. During the deformation process, the dislocation has a higher driving force, but it is not easily tangled locally, and the deformation microstrip disappears. The dislocation density in the crystal and grain boundary decrease significantly, but a certain number of dislocation walls and dislocation entanglements remain in the deformed matrix, indicating that the microstructural evolution of the alloy is in a dynamic recovery stage, shown in figure 7(b). When the deformation temperature is higher than 420 \(^{\circ}\)C, the deformed matrix possesses a morphology characteristic of dynamic recrystallization, a large number of dislocations are rearranged at the grain boundary, a small amount of dislocation entanglement occurs locally, and the dislocation density in the grains is very small. The small angle subgrain boundary formed by dislocation rearrangement can be clearly observed at the edge of the deformed grain boundary, and the submicron Al–Mn phase and Al\(_3\)(Sc, Zr) lattice particles distributed within and at the subgrain boundary are clearly discernible, shown in figure 7(c). Due to the high strain rate and the precipitated...
phase in the matrix hindering the movement of dislocations and the migration of new grain boundaries, there is a blocking effect on the dynamic recrystallization process. Even if the deformation temperature increases to 460 °C, the recrystallized grains are not coarsened significantly, shown in figure 7(d).

In summary, with an increase in the deformation temperature, the work hardening process caused by the deformed microstrip and dislocation cell transfers to a dynamic recrystallization softening process characterized by a low dislocation density and a small-size subgrain boundary. The power dissipation factor $\eta$ significantly increases. However, the recrystallization process consumes considerable stored deformation energy, preventing the potential risk of material failure caused by stress concentration and improving the stability of the thermoplastic deformation of the alloy. In general, when the plastic working strain is more than 0.5, a suitable hot working temperature is 380 °C with a strain rate of 0.001 ~ 0.3 s$^{-1}$.

3.6. Strain rate prediction equations
To further understand the inherent nature of the thermoplastic deformation behavior of the tested alloy, it is necessary to analyze the constitutive characteristics of the alloy during the thermal deformation process. Based on the stress-strain data, shown in figure 4, constructed hot working diagram, shown in figure 5, and microstructure characterization, shown in figures 6 and 7, the functional relationship between the strain rate, deformation temperature, and flow stress is analyzed. Five material constants $\alpha$, $\beta$, $Q$, $n$, and $A$ [18, 19] are used to develop the constitutive equation of the alloys. The relationship between the flow stress, deformation rate, and deformation temperature during hot deformation is described by the Arrhenius constitutive equation of the deformation activation energy [20–22]:

Figure 7. TEM micrographs of the specimens deformed under different conditions: (a) 280 °C and 0.1 s$^{-1}$, (b) 320 °C and 0.1 s$^{-1}$, (c) 420 °C and 0.1 s$^{-1}$, and (d) 460 °C and 0.1 s$^{-1}$.
There are three forms of the stress function $f(\sigma)$, which are expressed as follows [23, 24]:

\[ f(\sigma) = A_1 \sigma^{n_1} \quad (\alpha \sigma < 0.8) \]  
\[ f(\sigma) = A_2 \exp(\beta \sigma) \quad (\alpha \sigma < 1.2) \]  
\[ f(\sigma) = A[\sinh(\alpha \sigma)]^n \quad (For \ all \ \sigma) \]  

where $R$ is the ideal gas constant $(8.314 \text{ J mol}^{-1} \text{K}^{-1})$, $T^{-1}$ is the absolute temperature, $Q$ is the thermal deformation activation energy, $\sigma$ is the peak or steady state of flow stress, $A_1,A_2,n_1,n,\beta, \text{and} \alpha$ are material constants, and $\alpha = \beta/n_1$.

By introducing $f(\sigma)$ into equation (6) and taking the logarithm, expressions for the low and high stress levels are obtained respectively as follows:

\[ \ln \varepsilon = \ln A_1 + n_1 \ln \sigma - \frac{Q}{RT} \]  
\[ \ln \varepsilon = \ln A_2 + \beta \sigma - \frac{Q}{RT} \]  

In the above formulas, there is a linear relationship between $\ln \varepsilon$ and $\ln \sigma$, $\ln \varepsilon$ and $\sigma$, and the slopes of the stress-strain curves are $n_1$ and $\beta$ for equations (10) and (11), respectively. Based on the stress-strain curve of the alloy during hot compression, shown in figure 3, the flow peak stress data at different deformation conditions are collected and plotted by linear regression analysis, and the results are shown in figure 8. The slopes of each line in the graph are obtained, and the average values are calculated, yielding $n_1 = 6.523, \beta = 0.067 \text{ MPa}^{-1}$, and $\alpha = \beta/n_1 = 0.0103 \text{ MPa}^{-1}$.

The effects of the deformation temperature and deformation rate on the deformation behavior of the materials can be expressed by the Zener-Holloman equation:

\[ Z = \varepsilon \exp\left(\frac{Q}{RT}\right) = A[\sinh(\alpha \sigma)]^n \]  

The logarithm and partial derivative of both sides are taken, yielding the following results:

\[ Q = R \left\{ \frac{\partial \ln \varepsilon}{\partial \ln [\sinh(\alpha \sigma)]} \right\}_T \left\{ \frac{\partial \ln [\sinh(\alpha \sigma)]}{\partial (1/T)} \right\}_T = nRS \]  

where $n$ is the average slope of $\ln \varepsilon - \ln[\sinh(\alpha \sigma)]$ at different temperatures, and $S$ is the average slope of $\ln[\sinh(\alpha \sigma)] - 1000/T$ at different temperatures. The data obtained from the calculation are plotted, and the values of $n$ and $S$ are determined by a linear fitting. A deformation activation energy of $Q = 161.279 \text{ kJ mol}^{-1}$ is determined for the alloy by linear regression analysis, as shown in figure 9. This activation energy is lower than that of the Al-Zn-Mg-Er-Zr alloy reported by Wu [25] (189.67 kJ mol$^{-1}$), but much higher than the bulk self-diffusion activation energy (142 kJ mol$^{-1}$) of pure aluminum, indicating that dislocation cross-slip is the main deformation mechanism. Thus, the softening mechanism is dynamic recovery [26, 27].

The $\ln-\ln$ plot of equation (12) shows that there is a good linear relationship between $\ln Z$ and $\ln[\sinh(\alpha \sigma)]$. $\ln A$ is the intercept of $\ln Z - \ln[\sinh(\alpha \sigma)]$. Equation (12) is used to determine the $Z$ value at different temperatures and strain rates, as shown in figure 10. The least squares method is used for the linear fitting to determine $A = 1.101 \times 10^{11} \text{ s}^{-1}$.
Based on the above analysis, the material constants of the alloy are obtained, as summarized in table 2. The material parameters are substituted into the equation (12), and a flow stress constitutive equation as a function of the Z parameter is established as follows:

$$\sigma = 97.087 \ln \left\{ \frac{Z}{1.101 \times 10^{11}} \right\}^{4.58066} + \left[ \left( \frac{Z}{1.101 \times 10^{11}} \right)^{2/4.58066} + 1 \right]^{1/2} \right\} \quad (14)$$

where

$$Z = \dot{\varepsilon} \exp \left( \frac{161279}{RT} \right)$$

4. Conclusion

(1) With the increase in strain, the flow stress increased sharply in the initial stage and subsequently reached a peak value. After reaching the peak flow stress, the flow stress gradually decreased to a stable state. With the increase in strain rate and the decrease in deformation temperature, the flow stress of the alloy increased.
The flow stress behavior was related to the different microstructures and the dynamic softening mechanism.

2) The constitutive equation for the alloy was established by using a hyperbolic sine function, and the dependence of the deformation temperature and strain rate on the flow stress was obtained. The deformation activation energy of the alloy was calculated to be 161.28 kJ mol$^{-1}$.

3) The temperature and strain rate significantly influenced the microstructural evolution of the experimental alloy during the hot deformation. From the perspective of the microstructural evolution, the dynamic rheological softening was mainly caused by dynamic recovery and dynamic recrystallization. The presence of Al$_3$(Sc, Zr) particles, which were coherent with the matrix, significantly affected the microstructural evolution and effectively inhibited the dynamic recrystallization behavior by effectively pinning the dislocation movement and subgrain boundary slip.

4) The volume fraction of dynamic recrystallization increased with the increase in deformation temperature. Based on the analysis of the microstructural evolution, the construction of the machining diagram, and the solution of the rheological constitutive equation, suitable values of the deformation temperature of the experimental alloy were determined to be $380^\circ$C–$460^\circ$C and suitable values of the strain rate were determined to be 0.001 and 0.3 s$^{-1}$.

**ORCID iDs**

Feng Jiang  
https://orcid.org/0000-0003-0239-259X

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