Hot Deformation Behavior and Microstructure Evolution of High-Strength Al-Zn-Mg-Cu Alloy

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Abstract: An isothermal compression experiment was conducted to study the rheological behavior of Al-4.57Zn-1.50Mg-1.92Cu high-strength aluminum alloy at strain rates ranging from 0.1 to 20 s\(^{-1}\) and temperatures in the range of 573 to 773 K. Then, the effects of strain, strain rate, and deformation temperature on material deformation were investigated through orthogonal experiment analysis. According to the research results, strain rate and temperature had significant effects on the level of flow stress. Besides, the constitutive equation was established and demonstrated as applicable to predict the performance accurately. Meanwhile, the processing map under a true strain of 1.1 was built, to assess the deformation safety in different domains. Furthermore, the evolutionary trend of microstructure was observed by means of Scanning Electron Microscope, Electron Back-Scattered Diffraction and Transmission Electron Microscope. It was discovered that dynamic recovery and small-scale dynamic recrystallization played a major role in the softening mechanism of alloy during hot deformation. Moreover, dynamic recrystallization was found to have a significant impact on the hot deformation behavior of the alloys.

Keywords: Al-Zn-Mg-Cu alloy; hot deformation behavior; processing map; dynamic recovery; dynamic recrystallization

1. Introduction

Lightweight of materials is an essential method to improve the battlefield technology of equipment in aerospace, national defense, and military industry. Aluminum (Al) alloy has a high specific strength, good machinability, and excellent mechanical properties [1]. Particularly, 7xxx Al alloys exhibit higher strength and toughness, and thus have come to be the best lightweight material in aerospace and transportation fields [2,3].

The thermal deformation can effectively improve the microstructure and mechanical properties of metal. To obtain the machined components with excellent mechanical properties and ideal microstructure, their thermal deformation behavior and the optical processing parameters should be investigated [4,5]. Therefore, many researchers have studied the deformation behavior of 7xxx Al alloy at high temperatures.

Mirzadeh et al. [6] established the phenomenological and physical constitutive equations of 7075 Al alloy. The results revealed that the latter is more suitable to describe the thermal deformation behavior of 7075 Al alloy. Wang et al. [7] established processing maps to predict the workability of 7050 Al alloy, and proposed the optimum processing parameters with the temperatures ranging from 653 to 693 K and the strain rates ranging from 0.001 to 0.18 s\(^{-1}\). Ke et al. [8] found that both deformation temperatures and strain rates have significant influences on the microstructural evolution of AA7020 Al alloy during hot deformation, and there are uniform deformed grains and recovered structures in the hot-working safe zone in the study of AA7020 Al alloy. Parksy et al. [9] compared the thermal deformation behavior characteristics of as-cast and homogenized AA7075 Al alloy using the processing map. The results lay a foundation for the hot working of as-cast Al...
alloy. Yan et al. [10] established the thermal processing diagram of Al-6.2Zn-0.70Mg-0.3Mn-0.17Zr alloy on the basis of a dynamic material model. Meanwhile, Prasad et al. [11] studied instability criterion, and thereby obtained the optimal deformation range with the deformation temperature of 703~773 K and strain rate of 0.03~0.32 s\(^{-1}\). Constitutive equation and processing map are useful methods to study the flow behaviors and workability of alloys. However, in this paper, we not only study the hot deformation behavior of the alloy through the constitutive equation and processing map, but also discuss the influence of process parameters on the deformation of the material using orthogonal analysis.

In addition, due to the high stacking-fault energy of Al alloy, dynamic recrystallization (DRX) does not easily occur. Dynamic recovery (DRV) is the main softening mechanism during hot deformation. However, DRX often occurs with different processing parameters. Therefore, the microstructure evolution of Al alloys has been widely studied. Sun et al. [12] analyzed the deformation behavior and characteristics of the recrystallized in as-extruded Al-Zn-Cu-Mg alloy. The results demonstrated that continuous dynamic recrystallization (CDRX) preferentially occurs at the original grain boundaries. Tang et al. [13] based their studies on the observed microstructure characteristics and static softening behavior. A model was developed to predict the effects of alloying elements on microstructure evolution, recrystallization and static softening kinetics of Al-Zn-Mg-Cu alloys. Luo et al. [14] investigated the microstructure evolution of Al-Zn-Cu-Mg alloy by hot compression experiments and discovered that the dynamic softening mechanism comes from DRV and partial CDRX. Liu et al. [15] conducted hot compression experiments on AA7085 aluminum alloy, and the results showed that, with the increase of temperature, the dynamic competition of microstructure mechanisms mainly focuses on DRV and DRX. The DRV occurs more easily at higher strain rates, and DRX occurs at lower strain rates. However, there are few references of the microstructure evolution of as-rolled Al-Zn-Mg-Cu alloy.

Therefore, the objective of this study was to characterize the hot deformation behavior of the as-rolled Al-4.57Zn-1.50Mg-1.92Cu alloy based on the isothermal compression experimental results. The Arrhenius equation with strain compensation and the processing map were established through the true stress–strain curves. In addition, the influence of different processing parameters on the deformation of the material was analyzed by orthogonal experiment. Finally, the influence of the main processing parameters on the microstructure was discussed through microscopic characterization.

2. Experimental Methods

As-rolled Al-4.57Zn-1.5Mg-1.92Cu alloy was used for experiments. Its original microstructures are exhibited in Figure 1, compression direction (CD) for compression direction. Base Material (BM) indicates a typical rolling microstructure with an average grain size of 14.0 μm (Figure 1a) and a high proportion of low-angle grain boundaries (LAGBs) (2° < 15°) of 78% owing to the introduction of a high-density dislocation substructure during rolling [16] (Figure 1b).

![Figure 1. Microstructure morphology of Al-Zn-Mg-Cu alloy: (a) inverse pole figure (IPF); (b) high- and low-angle grain boundaries.](image-url)
The cylindrical sample processed a diameter of 10 mm and a height of 15 mm. Isothermal compression experiments were conducted on a Gleeble-3500 thermal simulation machine. During the compression, graphite sheets were added to both ends of the samples, and lubricant was evenly coated to reduce the friction effect. Then, the samples were heated to the test temperature at a heating rate of 10 °C/s and kept for 3 min to ensure the heat balance before deformation. The samples were compressed at temperatures of 573, 623, 673, 723 and 773 K, with strain rates of 0.1, 1, 10 and 20 s\(^{-1}\), respectively, and deformation amounts of 70%. The samples were cooled with water immediately after deformation to retain the microstructure of the samples after hot compression. The samples for microstructure analysis were sectioned in the center parallel to the compression axis. The Scanning Electron Microscope (SEM), Electron Back-Scattered Diffraction (EBSD) and Transmission Electron Microscope (TEM) techniques were used to observe the microstructure at the center of the samples.

3. Results and Discussion
3.1. True Stress–Strain Curves

Figure 2 depicts the true stress–strain curves of Al-Zn-Mg-Cu alloy under hot compression. The flow stress increases rapidly with the increase of strain at the initial stage, and then the growth rate of the flow stress decreases slightly with the deformation progress. The flow stress reaches a stable state when the strain is about 0.3. At the initial stage of deformation, the strain rate increases rapidly from zero to the rate required for the testing. Meanwhile, the dislocation density increases sharply. The interaction of the dislocations hinders the movement of the dislocation, presenting as work hardening [17]. Simultaneously, the deformation energy storage is very low due to the small strain, making it insufficient to drive DRV. Therefore, work hardening occupies a dominant position, and it is reflected in the true stress–strain curves that the flow stress rises rapidly [18]. With the increase of the strain, the defect density of dislocation and vacancy in the alloy keep rising, and the deformation energy storage in the alloy rapidly increases, providing enough driving force for dislocation climbing. Dislocation rearrangement and annihilation occur, accompanied by screw dislocation climbing, which is dynamic softening, and the increasing rate of true stress is moderate [19]. Then, the flow stress reaches a stable stage with the dynamic softening and work hardening reaches dynamic equilibrium. However, the curves have a slight upward trend when the strain rate is 1 s\(^{-1}\), which suggests that the work hardening effect again exceeds the dynamic softening effect; the flow stress will rise with the increase of strain [20].

Meanwhile, it can be seen from Figure 3 that the flow stress decreases with the increase of the deformation temperature and the decrease of the strain rate. This is because the increase of the deformation temperature provides higher grain boundary mobility for the nucleation and growth of DRX grains [21], while the decrease of strain rate can induce the longer dislocation movement time, and the accumulation of deformation energy [22]. Additionally, the deformation activation energy of atoms increases with the increase of deformation temperature, leading to further displacement of dislocations and vacancies [23]. Therefore, the DRV and DRX caused by dislocation slip and climb are improved.
Figure 2. True stress–strain curves of Al-Zn-Mg-Cu alloy: (a) 573 K, (b) 623 K, (c) 673 K, (d) 723 K, (e) 773 K.

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Figure 3. Relationship among flow stress (a) deformation strain rate and (b) temperature.
3.2. Orthogonal Analysis

An orthogonal experiment was used to analyze the influence of deformation parameters (strain, deformation temperature, and strain rate) on the flow stress of Al-Zn-Mg-Cu alloy. The quasi-horizontal method is used to establish the orthogonal table because the strain rate is only four. The influencing factors are provided in Table 1.

**Table 1.** Al-Zn-Mg-Cu alloy factor level table.

| No. | A (Strain) | B (Deformation Temperature/K) | C (Strain Rate/s⁻¹) |
|-----|------------|-------------------------------|---------------------|
| 1   | 0.1        | 573                           | 0.1                 |
| 2   | 0.3        | 623                           | 1                   |
| 3   | 0.6        | 673                           | 10                  |
| 4   | 0.9        | 723                           | 20                  |
| 5   | 1.1        | 773                           | 20                  |

The orthogonal table L{'25'}{\(5^6\)} is introduced, as presented in Table 2. The flow stress values under the corresponding deformation conditions were taken as the objective function to evaluate the influence significance of these factors. Variance analysis was conducted using the range \(K_i\) of each factor in Table 2.

The sum of the squares of the total deviations \(S_T\) is:

\[
S_T = \sum_{i=1}^{n} \sigma_i^2 - \frac{1}{n} \left( \sum_{i=1}^{n} \sigma_i \right)^2
\]  

(1)

where \(n\) denotes the number of tests and \(\sigma\) is the flow stress (MPa). The \(S_T\) can be obtained from Equation (1), \(S_T = 40,000.03\). The sum of deviation squared of each factor is \(S_j\):

\[
S_j = \frac{1}{r} \left( \sum_{i=1}^{r} K_i^2 \right) - \frac{1}{n} \left( \sum_{i=1}^{n} \sigma_i \right)^2
\]  

(2)

where \(r\) represents the number of repetitions at each level. The sum of deviation squares of factors A (strain), B (deformation temperature) and C (strain rate) can be calculated as \(S_A = 1087.322, S_B = 29,000.66\) and \(S_C = 8290.3366\), respectively. The \(S_j\) of the error is \(S_{\text{erro}}\):

\[
S_{\text{erro}} = S_T - S_A - S_B - S_C
\]  

(3)

Therefore, the value of \(S_{\text{erro}}\) can be calculated as 1621.706. The total degree of freedom \(f_T\) is \(n - 1 = 24\), and the degree of freedom of each factor is:

\[
f_j = r - 1
\]  

(4)

The values of degree of freedom for factors A (strain), B (deformation temperature) and C (strain rate) can be obtained: \(f_A = 4, f_B = 4, f_C = 3\). Therefore, the degree of freedom of error is:

\[
f_{\text{erro}} = f_T - f_A - f_B - f_C = 24 - 4 - 4 - 3 = 13
\]  

(5)

The mean deviation sum of squares \(MS\) is:

\[
MS = \frac{S_j}{f_j}
\]  

(6)
According to Equation (6), the values of $MS$ for each factor and the error can be obtained. Then, the $F$ value of each factor can be calculated as:

$$F = \frac{MS_j}{MS_{err}}$$ (7)

The calculated values are listed in Table 3. As indicated in Table 3, the value of $F_A$ is less than $F_{0.1}(4,13)$, which indicates that the change of strain has little effect on the flow stress. However, the values of $F_B$ and $F_C$ are higher than those of $F_{0.01}(4,13)$ and $F_{0.01}(3,13)$, respectively, indicating that deformation temperature and strain rate have significant effects on the flow stress. Meanwhile, the deformation temperature has the largest effect on the flow stress.

Table 2. Experimental scheme of orthogonal analysis of Al-Zn-Mg-Cu alloy.

| No. | Factors | Stress (MPa) |
|-----|---------|--------------|
| 1   | A 1     |              |
| 2   | A 1     |              |
| 3   | A 1     |              |
| 4   | A 1     |              |
| 5   | A 1     |              |
| 6   | A 1     |              |
| 7   | A 1     |              |
| 8   | A 1     |              |
| 9   | A 1     |              |
| 10  | A 1     |              |
| 11  | A 1     |              |
| 12  | A 1     |              |
| 13  | A 1     |              |
| 14  | A 1     |              |
| 15  | A 1     |              |
| 16  | A 1     |              |
| 17  | A 1     |              |
| 18  | A 1     |              |
| 19  | A 1     |              |
| 20  | A 1     |              |
| 21  | A 1     |              |
| 22  | A 1     |              |
| 23  | A 1     |              |
| 24  | A 1     |              |
| 25  | A 1     |              |

K1 593.5     814.9     373.3
K2 576.5     625.1     506.9
K3 529.9     530.4     604.9
K4 510.4     452.4     1240.3
K5 519.1     306.8

$T = 2729.4$
Table 3. Variance analysis of flow stress of Al-Zn-Mg-Cu alloy.

| Sources of Variance       | Sum of Deviation Squared S | Degrees of Freedom f | Sum of Squares of Mean Deviation MS | F Value | Significant |
|---------------------------|---------------------------|----------------------|------------------------------------|---------|-------------|
| Strain                    | 1087.322                  | 4                    | 217.8304                           | 2.179   | -           |
| Deformation temperature   | 29,000.66                 | 4                    | 7250.165                           | 58.119  | **          |
| Strain rate               | 8290.3366                 | 3                    | 2763.4455                          | 22.152  | **          |
| Error                     | 1621.706                  | 13                   | 124.7466                           | -       | -           |
| Total                     | 40,000.03                 | 24                   | -                                  | -       | -           |

$F_{0.01}(4,13) = 5.205, F_{0.01}(3,13) = 5.739, F_{0.1}(4,13) = 2.43.$ (“-” means insignificance; “**” means significance).

3.3. Constitutive Equation

According to the flow stress curve, the flow stress in the hot working process is related to the deformation temperature and strain rate, expressed by the Arrhenius equation [24]:

$$\dot{\varepsilon} = A_1\sigma^{n_1}\exp\left(-\frac{Q}{RT}\right) \quad (a\sigma < 0.8) \quad (8)$$

$$\dot{\varepsilon} = A_2\exp(\beta\sigma)\exp\left(-\frac{Q}{RT}\right) \quad (a\sigma > 1.2) \quad (9)$$

$$\dot{\varepsilon} = A_3\left[\sinh(\alpha\sigma)^n\right]\exp\left(-\frac{Q}{RT}\right) \quad \text{for all } \sigma \quad (10)$$

where $\dot{\varepsilon}$ denotes the strain rate, s$^{-1}$; $\sigma$ represents flow stress, MPa; $Q$ designates the activation energy of thermal deformation, J/mol; $R$ indicates the gas constant, 8.314 J/(mol·K); $T$ refers to the absolute temperature, K; $A$, $n_1$, $\beta$, $\alpha$ and $n$ are the material constants, and $\alpha = \beta / n_1$. The natural logarithms were applied for both sides for Equations (8)–(10) and can be expressed as follows:

$$\ln\dot{\varepsilon} = n_1\ln\sigma + \ln A_1 - \frac{Q}{RT} \quad (11)$$

$$\ln\dot{\varepsilon} = \beta\sigma + \ln A_2 - \frac{Q}{RT} \quad (12)$$

$$\ln\dot{\varepsilon} = n\ln[\sinh(\alpha\sigma)^n] + \ln A_3 - \frac{Q}{RT} \quad (13)$$

According to the above equations, the $\ln\dot{\varepsilon} - \ln\sigma$ and $\ln\dot{\varepsilon} - \sigma$ curves in different temperatures are fitted, as illustrated in Figure 4. After calculation, $n_1 = 10.419$, $\beta = 0.0976$, and $\alpha = \beta / n_1 = 0.0094$.

The relationship between temperature and strain rate can be established by using the Zener–Hollomon ($Z$) parameter:

$$Z = A\sinh(\alpha\sigma)^n = \dot{\varepsilon}\exp\left(\frac{Q}{RT}\right) \quad (14)$$

By transforming Equation (14) according to the definition of hyperbolic sine function, a constitutive model of flow stress $\sigma$ of different materials can be established with the $Z$ parameter and Arrhenius equation:

$$\sigma = \frac{1}{\alpha}\ln\left\{\left(\frac{Z}{A}\right)^{\frac{1}{n}} + \left[\left(\frac{Z}{A}\right)^{\frac{1}{n}} + 1\right]^{\frac{1}{2}}\right\} \quad (15)$$
Equation (10) is transformed into:

\[ \ln[\sinh(\alpha \sigma)] = \frac{Q}{nRT} + \ln \dot{\varepsilon} - \ln A_3 \]  

(16)

Therefore, the values of \( n \) and \( Q \) can be obtained from \( \ln \dot{\varepsilon} - \ln[\sinh(\alpha \sigma)] \) and \( 1000/T - \ln[\sinh(\alpha \sigma)] \) curves, as exhibited in Figure 4 (\( n = 7.458, Q = 188.03 \text{ kJ mol}^{-1} \)).

Equation (17) can be obtained from Equation (14) as:

\[ \ln Z = \ln A + n \ln[\sinh(\alpha \sigma)] \]  

(17)

The \( Z \) value can be calculated using Equation (17), and the values of \( A \) and \( n \) can be acquired by the \( \ln Z - \ln[\sinh(\alpha \sigma)] \) curve, as shown in Figure 5. Therefore, \( n = 8.0364 \) and \( A = 1.38 \times 10^{12} \).

Figure 4. Plots used for the calculation of hot deformation constants when \( \varepsilon = 0.1 \): (a) \( \ln \dot{\varepsilon} - \ln \sigma \); (b) \( \ln \dot{\varepsilon} - \sigma \); (c) \( 1000/T - \ln[\sinh(\alpha \sigma)] \); (d) \( \ln \dot{\varepsilon} - \ln[\sinh(\alpha \sigma)] \).

Figure 5. Relationship between \( \ln Z \) and \( \ln[\sinh(\alpha \sigma)] \).
According to the above calculations, each parameter value at different strains can be calculated [25], and the results are presented in Table 4.

### Table 4. Equation parameters corresponding to different strains of Al-Zn-Mg-Cu alloy.

| ε  | n     | α      | Q      | lnA   |
|----|-------|--------|--------|-------|
| 0.1| 10.42 | 0.00936| 188.03 | 27.95 |
| 0.2| 11.11 | 0.00814| 202.86 | 28.35 |
| 0.3| 11.72 | 0.00857| 208.71 | 29.63 |
| 0.4| 12.51 | 0.00884| 217.32 | 30.24 |
| 0.5| 12.72 | 0.00921| 223.04 | 30.86 |
| 0.6| 12.92 | 0.00897| 219.21 | 31.04 |
| 0.7| 12.51 | 0.00940| 201.89 | 29.33 |
| 0.8| 12.01 | 0.01209| 198.81 | 26.26 |
| 0.9| 9.23  | 0.01301| 196.72 | 24.13 |
| 1.0| 8.11  | 0.01402| 194.61 | 22.01 |
| 1.1| 7.21  | 0.01402| 194.61 | 22.01 |

According to Table 4, the values of $n$, $α$, $Q$, $\ln A$ and $ε$ can be obtained by using quartic polynomial fitting. Although the influence of strain on flow stress is not significant from the orthogonal test, it can be seen from Table 4 that strain has some influence on $n$, $α$, $Q$, and $\ln A$. Consequently, strain compensation to the equation can improve its accuracy.

Fitting results of various parameter curves are illustrated in Figure 6. The hyperbolic sinusoidal constitutive equation of flow stress in high-temperature deformation of Al-Zn-Mg-Cu alloy can be obtained as follows:

\[
\dot{\varepsilon} = A_3 [\sinh(\alpha \sigma)]^n \exp \left( -\frac{Q}{RT} \right),
\]

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

\[
n = 9.0561 + 16.8917\varepsilon - 44.4153\varepsilon^2 + 77.8969\varepsilon^3 - 54.4872\varepsilon^4
\]

\[
\alpha = 0.0129 - 0.0545\varepsilon + 0.2127\varepsilon^2 - 0.3232\varepsilon^3 + 0.1699\varepsilon^4
\]

\[
Q = 184.5317 - 11.4613\varepsilon + 695.2168\varepsilon^2 - 1440.4401\varepsilon^3 + 779.3124\varepsilon^4
\]

\[
\ln A = 27.5461 + 0.6795\varepsilon + 29.6471\varepsilon^2 - 36.3727\varepsilon^3 + 1.1072\varepsilon^4
\]

(18)

The flow stress values under different conditions were calculated by substituting the material parameters of Al-Zn-Mg-Cu alloy at the strain rate of 0.1 s$^{-1}$ and the temperature of 573, 625, 673, 723 and 773 K into the equation to test the accuracy of the developed constitutive equation. Figure 7 exhibits the comparison between the flow stress values and the predicted values at the strain rate of 0.1 s$^{-1}$. It can be revealed that the corrected curves are consistent with the predicted curves. With the purpose of obtaining the error between the experimental data and the predicted values from constitutive equation more clearly, the error analysis (Equations (19) and (20)) is introduced [26]:

\[
R_{er} = \left| \frac{\sigma_E - \sigma_C}{\sigma_E} \right| \times 100\% \tag{19}
\]

\[
R_{av} = \frac{1}{N} \sum_{i=1}^{N} \left| \frac{\sigma_E - \sigma_C}{\sigma_E} \right| \times 100\% \tag{20}
\]

where $\sigma_C$ denotes the calculated values of flow stress; $\sigma_E$ represents the flow stress correction values.

$R_{er}$ and $R_{av}$ indicate a relative error and average relative error, respectively. According to Equations (19) and (20), the error between the flow stress and the constitutive equation can be obtained. It can be observed that the relative error between the experimental values and the modified values of the constitutive equation is small, the average relative error is only 4.325 % and $R = 0.992$. 
3.4. Processing Map and Instability Area

The construction of the processing map is mainly based on the Dynamic Material Model (DMM) [24]. The processing map is superimposed by the power dissipation diagram and the plastic instability diagram. The safe area and the instability area are apparent, contributing to determining the metal processing area, obtaining the corresponding processing parameters, and providing a reference for production. The total dissipated power ($P$) is given by:

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

(21)
where $G$ represents the dissipation energy stemming from plastic deformation and $J$ represents the dissipated power caused by microstructure. The efficiency of power dissipation $\eta$ can be defined as:

$$
\eta = \frac{J}{J_{\text{max}}} = \frac{2m}{m + 1}
$$

(22)

$\eta$ is an important element to indicate power dissipation. In general, the best hot processing parameters appear at maximum $\eta$. In addition, the criterion of rheological instability is as follows:

$$
\xi(\dot{\varepsilon}) = \frac{\partial \ln(m/(m + 1))}{\partial \ln \dot{\varepsilon}} + m < 0
$$

(23)

where $\xi(\dot{\varepsilon})$ represents the instability parameter. According to the above formula, the thermal processing map at true strain 1.1 is established, as shown in Figure 8. The contour numbers represent the $\eta$ values, and the gray area suggests the instability area. There is a rheological instability region at the deformation temperature of 573–673 K and a strain rate of 0.1–2.85 s$^{-1}$ (Domain I). Generally, when $\eta$ is greater than 30% in the stable region, the deformation mechanism can be regarded as DRX, and the optimized deformation process parameters are generally selected in this region [27]. Besides, a stable region with $\eta > 30\%$ is located in the deformation temperature range of 700–773 K and the strain rate range of 0.1–1.65 s$^{-1}$ (Domain II). Therefore, the optimum deformation process parameters of this alloy should be selected at a high deformation temperature and large strain.

Figure 8. Processing map of Al-Zn-Mg-Cu alloy under strain 1.1.

Figure 9 exhibits the microstructural characterization of an Al-Zn-Mg-Cu alloy deformed in the instability area. Micropores appear around the second phase ($\eta$ phase MgZn$_2$) in Figure 9b. The reason for the micropores can be attributed to the large deformation of the soft matrix. Moreover, the hard and brittle second phase is difficult to coordinate deformation at the high strain rate. Additionally, Figure 9c depicts a lot of holes. As the deformation occurs, the stress around the particles is concentrated, leading to the breakage and gradual debonding of the large particles forming the holes. To sum up, micropores and the fragmentation of large particles are the main causes of instability.
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Figure 9. Microstructure in the instability area of Al-Zn-Mg-Cu alloy: (a) IPF, (b,c) SEM, (d) KAM, (e) TEM, (f) Distribution of recrystallized grains, sub-crystal structure and deformed structure in the EBSD maps.

Figure 9d,e suggests that dislocation density is high in the instability area. There exist a lot of dislocation lines, dislocation pileup, and dislocation accumulation, forming dislocation walls and a dislocation tangles area [28,29]. Dislocation lines and dislocation pileup indicate that the dislocation is not consumed by DRV and maintains the initial state of deformation. However, the emergence of the dislocation wall and the dislocation tangles reflect the occurrence of DRV. Dislocation entanglement is contributed by dislocation climbing and slippage, and dislocation walls are generated through the mutual cancellation of unlike dislocations at the dislocation pileup [30,31], leading to a formation of a large number of sub-crystal structures, as illustrated in Figure 9f. According to the above microstructure analysis, the process parameters involved in the instability zone should be avoided in the hot working process.

3.5. Microstructure Evolution
3.5.1. Effects of Temperature on Microstructure Evolution

Figure 10 exhibits the EBSD results of hot compression of Al-Zn-Mg-Cu alloy at the strain rate of 0.1 s⁻¹. As observed from the figure, the original grains are elongated perpendicular to the compression direction, and a large number of LAGBs are found in the deformed grains under different deformation conditions [32]. Some deformed grains also exhibit orientation gradients, owing to the formation of substructures and continuous DRX [33].

As indicated in Figure 10a,d, the large grains in the original solution are elongated after deformation at 573 K and 673 K, while the morphologic structure is still retained. With an increase in temperature, the grain width increases, and the structure shows a significant DRV feature. When the temperature rises to 673 K, the grain boundary presents a significant serrated shape, and small equiaxed DRX grains appear, suggesting that DRX begins to occur [34]. As the temperature is increased to 773 K (Figure 10g), the DRX grains become more and more apparent, and the grains’ size and number also increase. This can be explained by the fact that, with the increase of deformation temperature, the atomic activity is enhanced, the recrystallization activation energy is reduced, and DRX is more likely to occur. Therefore, the softening mechanism of Al-Zn-Mg-Cu alloy changes from DRV to DRX during thermal deformation [35].
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Figure 10 exhibits the EBSD results of hot compression of Al-Zn-Mg-Cu alloy at the strain rate of 0.1 s\(^{-1}\). As observed from the figure, the original grains are elongated perpendicular to the compression direction, and a large number of LAGBs are found in the deformed grains under different deformation conditions [32]. Some deformed grains also exhibit orientation gradients, owing to the formation of substructures and continuous DRX [33].

Figure 10. IPF maps, KAM maps and statistics of misorientation angle at different temperatures (\(\dot{\varepsilon} = 0.1 \text{ s}^{-1}\)) (a–c) \(T = 573 \text{ K}\); (d–f) \(T = 673 \text{ K}\); (g–i) \(T = 773 \text{ K}\).

Figure 10b,e,h indicates Kernel Average Misorientation (KAM); Figure 10c,f,i presents statistics of misorientation angle of Al-Zn-Mg-Cu alloy. It can be discovered that, under the constant strain rate, the dislocation density decreases significantly with the rise of temperature. Meanwhile, the content of high-angle grain boundary (HAGBs) increases. At the strain rate 0.1 s\(^{-1}\), the content of HAGBs increases from 24.97% to 52.08% with the increase of deformation temperature from 573 K to 773 K. This phenomenon can be explained as follows: with the increase of deformation temperature, the atomic activity increases, the activation energy for dislocation motion decreases. As a result, the cancellation and annihilation are dislocated, the occurrence of DRV and DRX is promoted, and the content of HAGBs increases.

Figure 11 depicts the distribution of recrystallized grains, sub-crystal structure and deformed structure in the EBSD maps of Al-Zn-Mg-Cu alloy at different temperatures with the strain rate of 0.1 s\(^{-1}\). It can be discovered by comparison that the deformed structure is dominated and the content of recrystallized structure is very low at the deformation temperature 573 K. When the temperature rose from 573 K to 773 K, the relative content of recrystallization increases from 2.2% to 56.2%, the relative content of the sub-crystal structure fluctuates slightly, and the relative content of the deformed structure decreases from 64.1% to 15.7%. With the increase in temperature, the content of recrystallized structure increases slightly, the content of sub-crystal structure increases, and the content of deformed structure decreases.
As indicated in Figure 10a,d, the large grains in the original state were significantly reduced at 573 K. Meanwhile, a small amount of deformation and sub-crystal structure is also present. Therefore, the dynamic softening mechanism of Al-Zn-Mg-Cu alloy is mainly DRV, while the dynamic softening mechanism of Al-Zn-Mg-Cu alloy is primarily DRX at a low strain rate (0.1 s\(^{-1}\)) and high deformation temperature (DRX fraction = 56.2%). Figure 12 shows the TEM figures of Al-Zn-Mg-Cu alloy deformed at 0.1 s\(^{-1}\). It can be seen from the figure that there is an increase of secondary grains during hot compression. The second phase is precipitated as \(\eta'\) (MgZn\(_2\)) by diffraction spot analysis. The size of the second phase decreases significantly with the increasing temperature. At the temperature 673 K and strain rate 0.1 s\(^{-1}\), the second phase is distributed in the matrix, and the number is small. This does not promote recrystallization. Therefore, the content of the recrystallization structure is relatively lower under this condition. When the temperature is 773 K, the number and size of the second-phase particles significantly increase (Figure 12b), and the recrystallization is enhanced.

Besides, the recrystallized structure is dominant when the temperature rises to 773 K. Meanwhile, a small amount of deformation and sub-crystal structure is also present. Therefore, the dynamic softening mechanism of Al-Zn-Mg-Cu alloy is mainly DRV, while the dynamic softening mechanism of Al-Zn-Mg-Cu alloy is primarily DRX at a low strain rate (0.1 s\(^{-1}\)) and high deformation temperature (DRX fraction = 56.2%).

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Figure 11. Distribution of recrystallized grains, sub-crystal structure and deformed structure in the EBSD maps of samples under different conditions: (a) 0.1 s\(^{-1}\), 573 K; (b) 0.1 s\(^{-1}\), 673 K; (c) 0.1 s\(^{-1}\), 773 K.

3.5.2. Effects of Strain Rate on Microstructure Evolution

Figure 13 shows the EBSD maps of Al-Zn-Mg-Cu alloy at the deformation temperature of 673 K. It can be illustrated from the figures that, under a certain temperature, the aspect ratio of elongated grains decreases, and the equiaxed recrystallized grains increase with the decrease of the strain rate. This is mainly because, at a certain temperature, the deformation time of unit strain becomes longer and the activation time of dislocation is shortened with the decrease of strain rate. This impedes DRV, DRX and other softening processes from occurring or being fully conducted, manifested as the length–diameter ratio increases with the increase of strain rate. The number of recrystallized grains increases with the increase of the strain rate [3].

Figure 12. TEM figures of Al-Zn-Mg-Cu alloy: (a) 0.1 s\(^{-1}\), 673 K; (b) 0.1 s\(^{-1}\), 773 K.
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**Figure 13.** IPF maps, KAM maps and statistics of misorientation angle at different strain rates (T = 673 K) (a–c) \(\dot{\varepsilon} = 0.1 \text{ s}^{-1}\); (d–f) \(\dot{\varepsilon} = 1 \text{ s}^{-1}\); (g–i) \(\dot{\varepsilon} = 10 \text{ s}^{-1}\).

Figure 13b,e,h provides the KAM maps and orientation difference angle distribution. It can be found that the dislocation density decreases significantly and the HAGBs gradually increase with the increase of the strain rate. As the strain rate gradually increases from 0.1 to 10 s\(^{-1}\), the content of HAGBs increases from 30.72% to 35.93% and then 40.71%, successively, and the content of HAGBs increases.

With the increasing strain rate, the dislocation movement speed is faster, and it is easier to produce dislocation pileups and to form LAGBs. However, LAGBs decrease with the increase of the strain rate in this research. DRV and DRX softening behavior can take place fully because of the increase in the number of dislocations per unit time. For example, DRV can cause dislocation rearrangement, and DRX will consume dislocation. The two dynamic softening behaviors will decrease the LAGBs content and increase the HAGBs content. Dislocation and DRV or DRX offset each other, and the content of the HAGBs increases.

Figure 14 suggests the distribution of recrystallized grains, sub-crystal structure and deformed structure in the EBSD maps of Al-Zn-Mg-Cu alloy at a deformation temperature of 673 K. With the increase of the strain rate, the relative content of recrystallized structure increases from 12.44% to 20.21%, the relative content of sub-crystal structure increases from 45.17% to 71.02%, and the relative content of deformation structure decreases from 42.39% to 8.77%. Under constant deformation temperature, with the increase of strain rate, the relative content of recrystallized structure and sub-crystal structure increases while the deformed structure decreases. This is due to the fact that the number of produced dislocations per unit time and the probability of DRV and DRX increase with the increasing strain rate at the certain deformation temperature. On the contrary, DRV and DRX may not be completed due to the increase of the dislocation movement speed. Therefore, the increase rate of recrystallization structure content is small, and the main softening mechanism is DRV.
Figure 13b,e,h provides the KAM maps and orientation difference angle for recrystallization. Therefore, the content of recrystallized structure increases with the increasing strain rate.

Figure 14 shows the TEM pictures of Al-Zn-Mg-Cu alloy deformed at 673 K. It can be indicated that the second phase is significantly coarsening with the increase of the strain rate, causing dislocation concentration in the deformation process and uneven deformation, providing favorable conditions for recrystallization. Therefore, the content of recrystallized structure increases with the increasing strain rate.

After isothermal, experimental compression, the softening mechanism is mainly DRV accompanied by a small part of DRX. Similarly, the deformation mechanism of 5xxx Al alloy is DRV, while the volume fraction of DRX increases significantly with increasing temperature and decreasing strain rate [21]. In addition, on the deformation mechanism of Al alloy, it was also found that the main deformation mechanism of SiC/AA6061 composite is DRX (Table 5) [36].

![Figure 14](image1.png)

![Figure 15](image2.png)

Figure 14. Distribution of recrystallized grains, sub-crystal structure and deformed structure in the EBSD maps of samples under different conditions: (a) 0.1 s⁻¹, 673 K, (b) 1 s⁻¹, 673 K, (c) 10 s⁻¹, 673 K.

Figure 15 shows the TEM pictures of Al-Zn-Mg-Cu alloy deformed at 673 K. It can be indicated that the second phase is significantly coarsening with the increase of the strain rate, causing dislocation concentration in the deformation process and uneven deformation, providing favorable conditions for recrystallization. Therefore, the content of recrystallized structure increases with the increasing strain rate.

![Figure 15](image3.png)

Figure 15. TEM figures of Al-Zn-Mg-Cu alloy: (a) 0.1 s⁻¹, 673 K, (b) 10 s⁻¹, 673 K.

| $Q$ (KJ/mol) | Material                        | Reference |
|-------------|---------------------------------|-----------|
| 166         | 5083 Al alloy                   | [37]      |
| 168.08      | Al-5.8Zn-2.3Mg-1.5Cu-0.21Cr alloy | [38]      |
| 196.27      | Al-Cu-Mg-Ag alloy               | [39]      |
| 150.25      | Al-Zn-Mg-Sc-Zr alloy            | [40]      |

Table 5. Thermal deformation activation energy of various materials.

4. Conclusions

Herein, the stress-strain curves of Al-Zn-Mg-Cu alloy were plotted after hot compression experiments were conducted. Then, an analysis was conducted as to the evolutionary trend of microstructure Al-Zn-Mg-Cu alloy under different deformation conditions. The conclusions are drawn as follows:
The flow stress decreases with the increase of temperature and the decrease of strain rate. Meanwhile, the true stress–strain curve shows that the rheological stress increases rapidly with the increase of strain. When $\varepsilon = 0.3$, the curve enters the steady flow stage.

The influence of hot processing parameters on the flow stress was analyzed using orthogonal experimental analysis. The results revealed that deformation temperature and strain rate exerted significant influence on the flow stress of Al-Zn-Mg-Cu alloy.

The Arrhenius-type equation was established. The $R$ and AARE values were 0.992 and 4.325%, respectively, which indicated that the equation could accurately characterize the flow stress.

According to the process maps, the optimum hot working condition is between 573 and 673 K, and strain rate ranges from 0.1 to 2.85 s$^{-1}$. Furthermore, the micropores and the fragmentation of large particles are the main causes of instability region.

According to the microstructure analysis, DRV is the primary softening mechanism of Al-Zn-Mg-Cu alloy during hot deformation. It is accompanied by a tiny amount of DRX. As deformation temperature and strain rate increase, the relative content of dynamic recrystallized structure shows an upward trend and that of the deformed structure is the opposite.

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