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A Modified Johnson–Cook Model and Microstructure Evolution of As-Extruded AA 2055 Alloy during Isothermal Compression

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Abstract: Isothermal compression behaviors of as-extruded AA 2055 alloy (T6 state) were studied at temperature of 320, 380, 440 and 500 °C with strain rate of 0.001, 0.01, 0.1 and 1 s^{−1} by a Gleeble-3800 testing machine. A modified Johnson–Cook model fitted by polynomial and power-exponential functions were established to describe the flow stress of the alloy. The constitutive models fitted by higher-order polynomials were more accurate than the ones fitted by second-order polynomial and power-exponential functions. The constitutive model fitted by a fourth-order polynomial was chosen for the optimal constitutive model in order to balance the prediction accuracy and model complexity. The modified Johnson–Cook constitutive model could predict the flow stress well, especially in high-temperature zone (around 500 °C) and low-temperature zone (around 320 °C). The dynamic precipitation and dissolution of the T₁ phase during hot compression were discussed. The unusual dynamic precipitation of the T₂ phase was investigated during hot compression by XRD and TEM. The massive dense fine precipitates effectively pinned dislocations or subgrain boundaries to accelerate DRV but suppressed DRX, leading to a low frequency of HAGBs in compressed samples.

Keywords: AA 2055 alloy; isothermal compression; Johnson–Cook model; dynamic precipitation; dynamic dissolution

1. Introduction

Lithium is the lightest metallic element on earth. Former research studies have revealed that an addition of 1 wt.% of lithium into an aluminum matrix can decrease the density of the aluminum alloy by 3% and increase the elastic modulus of the aluminum alloy by 5~6% [1–5]. Compared to the traditional 2xxx and 7xxx aluminum alloys, an Al–Li alloy has a low density, high elastic modulus, high specific strength, high specific elastic modulus and low fatigue crack growth rate [4–7]. An Al–Li alloy also has good mechanical properties at elevated temperature or at low temperature [6,8]. Compared with carbon-fiber-reinforced plastic (CFRP), an Al–Li alloy has irreplaceable advantages in impact resistance, plasticity and repairability. The AA 2055 alloy belongs to the third generation Al–Li alloys and was developed by Aluminum Company of America (Alcoa) in 2012 [8]. The alloy exhibits a higher fracture strength and fracture toughness than other third-generation Al–Li alloys through tweaking the alloying elements without increasing cost [9].

The Al-Li alloy components are always processed by plastic deformation [9]. In some case, the wrought Al alloys require further deformation, such as large-sized as-extruded bars or forged billet, while the reports about isothermal compression of wrought Al alloys are few, therefore an AA 2055 as-extruded bar with a diameter of 100 mm, a length of about 500 mm was chosen as the initial material in this work. The flow stress and constitutive model are important factors to describe the hot plastic deformation of an Al-Li alloy. During the hot deformation, a work hardening and softening of dynamic recovery (DRV) and dynamic recrystallization (DRX) both take place, under the joint impact of several thermal deformation parameters, such as the composition of the alloy, deformation temperature, strain rate and initial state [10–13]. A constitutive model establishes the relationship among flow stress, flow strain, deformation temperature and strain rate. Constructing an accurate and efficient constitutive model is the key to determine process route and optimize process parameters [14].

According to theoretical derivation methods, constitutive models can be divided into phenomenological models, physical models and neural network models [14,15]. A phenomenological constitutive model mainly reflects the macroscopic factors of hot deformation [15], without including the microscopic mechanism of hot deformation, such as the Arrhenius type model [16–18]. A physical model is built on a certain physical background, considering the influence of relevant physical parameters, such as the Zerilli–Armstrong (ZA) model [19–21]. However, in order to improve prediction accuracy, the physical constitutive model always introduces numerous complex variables which are hardly obtained. A neural network model is used to predict the flow stress of hot deformation by manual training because of its efficient nonlinear mapping ability [22,23]. Although a neural network model has a higher prediction accuracy resulting from its nonlinear mapping ability, its convergence speed is slow, and the problem of local minimization is common [14]. Therefore, a phenomenological model is a relatively appropriate model for an FEM analysis and industrial production.

The Johnson–Cook constitutive model is a classical phenomenological model which can predict the constitutive relationship of metal materials including the effect of strain, strain rate and temperature on the flow stress [17,24]. It is widely used in the numerical simulation of metal forming because the model is simple in form and the parameters are easy to obtain. However, since the Johnson–Cook constitutive model assumes that strain, strain rate and temperature have independent effects on the hot deformation behavior, the accuracy of prediction of the original model is inadequate. Our former report established the Arrhenius model of the alloy, which only described the relationship between stress and temperature or strain rate without reflecting the effect of strain on the flow stress [25]. Therefore, a modified Johnson–Cook constitutive model was established considering the coupling effect of temperature, strain rate and strain to improve the accuracy of model prediction and build the direct relationship between flow stress and strain.

The T_1 phase (Al_2CuLi), β' (Al_3Zr), the T_2 phase (Al_6CuLi_3), δ' (Al_3Li) and θ' (Al_2Cu) are the precipitates in the Al-Li-Cu-Zr alloys forming according to the process of hot deformation and heat treatment [2,9,26–28]. The dissolution or precipitation of the second-phase particles in Al-Li alloys may take place during every stage of the hot deformation. The influence of variation of the second-phase particles on the microstructure evolution during hot deformation is non-negligible. The second-phase particles may retard or accelerate dynamic recovery (DRV) and dynamic recrystallization (DRX) depending on their size, spatial distribution and deformation conditions in Al-Li alloys [10,13,29]. According to former research studies [10,11,13,30,31], the dispersed fine second-phase particles can pin dislocations or subgrain boundaries to accelerate DRV but suppress DRX in Al-Li alloys. While the large stable particles ($>1\text{ }\mu\text{m}$ in diameter) can promote DRX by producing regions of high lattice misorientation to offer preferential sites for recrystallization, i.e., particle-stimulated nucleation (PSN) [29,32], the second-phase particles also affect the deformation structure and texture depending on thermomechanical processing and their diameters and deformability. In general, Al-Li alloys are strengthened by precipitates of ordered atomic

structure. Moreover, the fine deformable precipitates are sheared by moving dislocations leading to a strong slip localization on a few slip planes which can puncture adjacent grains to form shear bands [29].

The modified Johnson–Cook constitutive model fitted by a second-order polynomial could predict the flow curves well in former reports, in the case where the stress of alloy changed monotonically with the increase of strain on the whole after the stress reached the peak value [33–35]. However, the experimental results in this paper suggest that the overall stress–strain relationship of the Al–Li alloy cannot be considered monotonic after the peak stress at different temperatures and strain rates, which is quite different from previous studies. In order to improve the prediction accuracy, a modified Johnson–Cook model fitted by a power-exponential function and a higher-order polynomial is also established in this work. In addition, the dynamic precipitation and dissolution and the effect of the second phase on DRV and DRX in as-extruded AA 2055 alloy during hot compression are also discussed in this work.

2. Materials and Methods

The as-extruded AA 2055 alloy (T6) with a composition of 3.5% Cu, 1.1% Li, 0.5% Zn, 0.3% Mn, 0.4% Mg, 0.1% Zr, 0.3% Ag, 0.1% Ti, 0.15% Si, 0.2% Fe, <0.15% other elements except the above elements and Al (all in wt.%) was used. The composition of the alloy was analyzed by inductively coupled plasma spectrometry (ICP, ThermoScientific iCAP6300, Waltham, MA, USA). The hot compression experiment was carried out by a Gleeble-3800 thermo-force simulation system (DSI, New York, NY, USA). The heating rate was 5 °C/s with a holding time of 3 min. The temperatures of hot compression were 320 °C, 380 °C, 440 °C and 500 °C, and the strain rates were 0.001 s^{−1}, 0.01 s^{−1}, 0.1 s^{−1} and 1 s^{−1}. The experiment with a higher strain rate of 5 s^{−1} at 380 °C was also carried out, but a deformation instability took place during hot compression; the result can be obtained from our former report [25]. The samples for hot compression were cut by wire electrical discharge machining from the as-extruded AA 2055 alloy with a diameter size of 10 mm × length 15 mm, as shown in Figure 1. The compressed samples were quenched by water with a height reduction of 60%. The detailed experimental procedures and the flow curves can be obtained from our previous report [25], as shown in Figures 1 and 2.

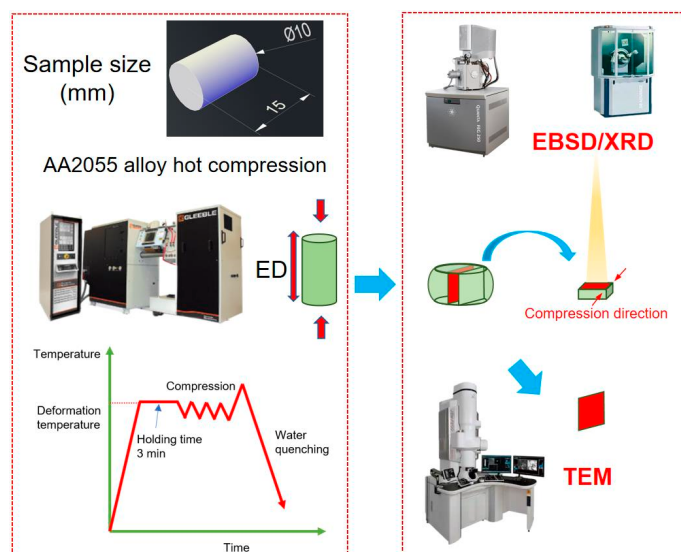


Figure 1. The diagrammatic sketch of the experimental procedure. Adapted from ref. [25].

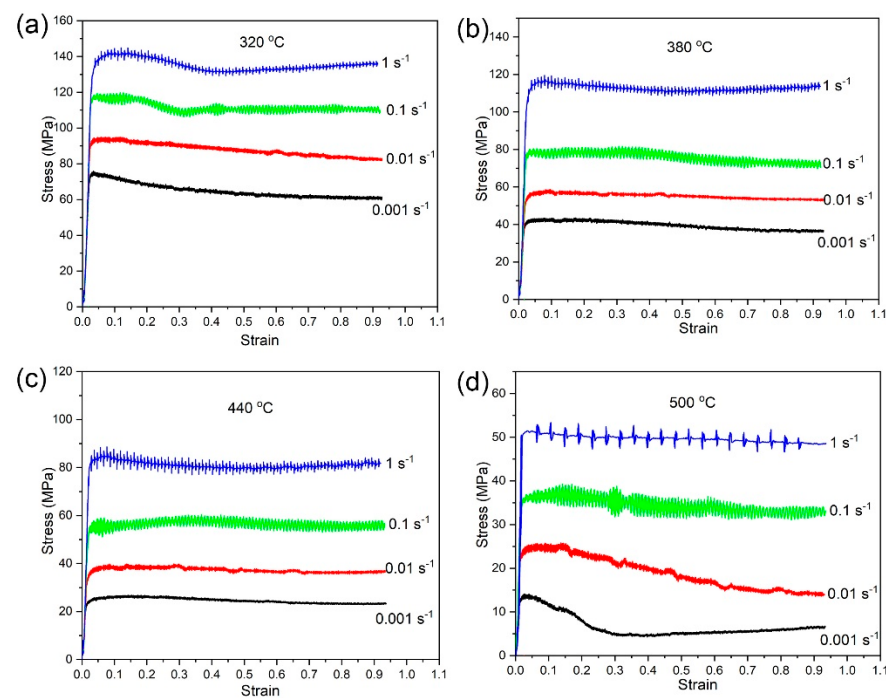


Figure 2. The flow curves of 2055 Al-Li alloy at temperature of (a) 320 °C, (b) 380 °C, (c) 440 °C and (d) 500 °C, with strain rates ranging from 0.001 s^{−1} to 1 s^{−1}. Reprinted from ref. [25].

In order to observe the precipitation behavior during hot compression, the X ray diffraction (XRD) analysis of the original sample and compressed samples were carried out (D8 Advance Bruker AXS, Karlsruhe, Germany) with a scan speed of 4°/min from 20° to 90°. The samples for the electron backscattered diffraction (EBSD) were sectioned parallel to the compression direction (CD) and microstructure observations were taken in the center of the sample, as shown in Figure 1. The samples were mechanically polished carefully, then electrolytically polished in a solution consisting of 10 mL of HClO₄ and 90 mL of C₂H₅OH with a voltage of 25 V at room temperature. The EBSD analysis was conducted on a field emission scanning electron microscope (Qunta 250 FEI, Hillsboro, USA) equipped with an Oxford EBSD detector operating at 30 kV. The foils for transmission electron microscopy (TEM) were also taken from the compressed samples parallel to the compression direction as shown in Figure 1. The samples were polished to 30 µm in thickness followed by ion milling (Leica RES101, Wetzlar, Germany). The TEM analysis was carried out on a transmission electron microscopy (JEOL-F200, Tokyo, Japan) with an accelerating voltage of 200 kV.

3. Results and Discussion

3.1. The Modified Johnson–Cook Model

The Johnson–Cook model is suitable for describing hot deformation with a wide range of strains and temperatures [36]. The flow stress is the function of temperature, strain rate and strain, as shown in Equation (1),

$$\sigma = (\gamma_1 + \gamma_2 \varepsilon^n)(1 + K_1 \ln \dot{\varepsilon}^*)(1 - T^{*m}) \quad (1)$$

where σ , ε , $\dot{\varepsilon}$, and $\dot{\varepsilon}^*$ are the flow stress, the effective plastic strain, the strain rate and the ratio of strain rate to reference strain rate. γ_1 , γ_2 and n denote the yield strength, the strain hardening coefficient and the strain hardening exponent at a reference strain rate and reference temperature. K_1 is the strain-rate hardening coefficient and m is the thermal softening exponent. The homologous temperature T^* can be calculated by Equation (2),

where T_r , T_m and T represent the reference temperature, melting point of the material and the current temperature, respectively [36]:

$$T^* = \frac{T - T_r}{T_m - T_r} \quad (2)$$

The relationship between stress and strain can be fitted by Equations (3) and (4) according to the compression curves with the effect of the variation of temperature and strain rate on stress ignored.

$$\sigma = (a_1 + a_2 \varepsilon^1 + a_3 \varepsilon^2) \quad (3)$$

$$\sigma = (a_1 + a_2 \varepsilon^{a_3}) \quad (4)$$

Considering the effect of the strain rate on the flow stress, the relationship between σ and $\ln(\dot{\varepsilon}/\varepsilon_r)$ with the compression data at temperature of 593.15 K (320 °C) and strain rate of 0.001 s^{-1} can be expressed as Equation (5), as shown in Figure 3.

$$\sigma = b_1 + b_2 \ln(\dot{\varepsilon}/\varepsilon_r) \quad \varepsilon_r = 0.001 \text{ s}^{-1} \quad (5)$$

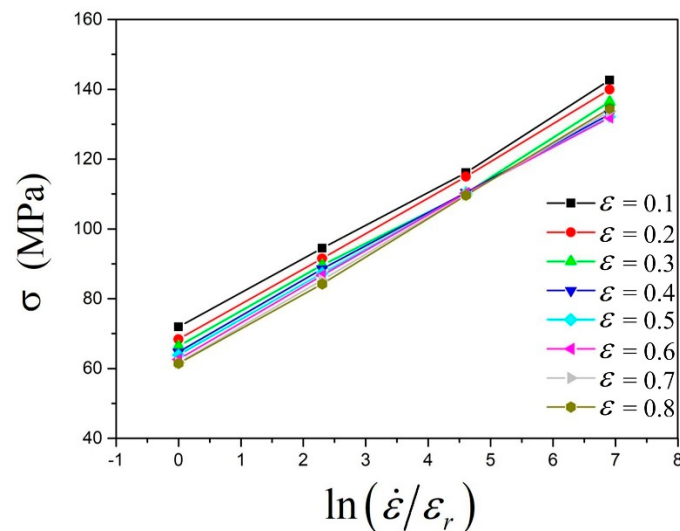


Figure 3. The relationship between σ and $\ln(\dot{\varepsilon}/\varepsilon_r)$ at $T = 593.15 \text{ K}$ (320°C).

When considering the effect of temperature variation on stress, it can be found from Figure 4a that the logarithm of the stress is linear with the difference between the current temperature and the reference temperature ($T - T_r$), when the strain rate is 0.01 s^{-1} , and the relationship between the stress and the temperature can be expressed by Equation (6).

$$\sigma = \exp(M(T - T_r)) \quad (6)$$

where $T_r = 593.15 \text{ K}$ (320°C), M is a parameter reflecting the coupling effect of temperature and strain rate, which can be expressed as Equation (7).

$$M = c_1 + c_2 \ln(\dot{\varepsilon}/\varepsilon_r) + c_3 (\ln(\dot{\varepsilon}/\varepsilon_r))^2 \quad (7)$$

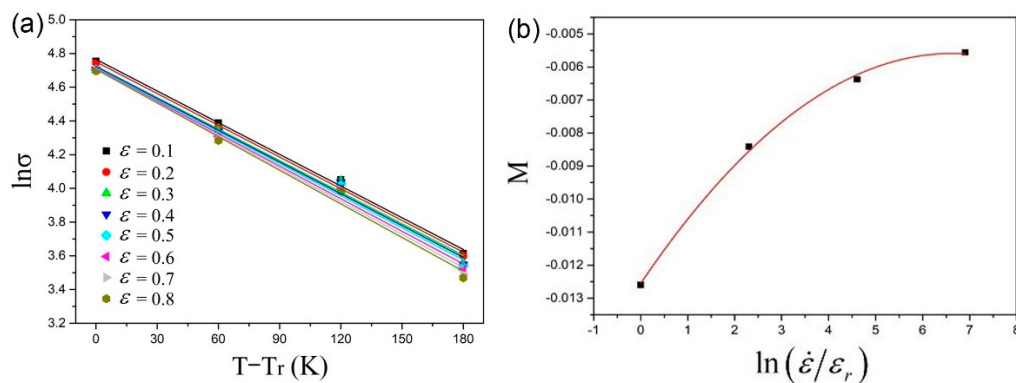


Figure 4. (a) The relationship between $(T - T_r)$ and $\ln \sigma$; (b) the relationship between $\ln(\dot{\epsilon}/\epsilon_r)$ and M .

The values of c_1 , c_2 and c_3 can be determined by the values of M with strain rates of 0.001 s^{-1} , 0.01 s^{-1} , 0.1 s^{-1} and 1 s^{-1} , respectively.

In summary, the modified Johnson-Cook constitutive model can be expressed as the following equation, which considers the coupling effect among temperature, strain rate and strain. The model fitted by a polynomial of order two is shown below:

$$\sigma = (a_1 + a_2\epsilon + a_3\epsilon^2) [1 + a_4 \ln(\dot{\epsilon}/\epsilon_r)] \exp \left[\left[a_5 + a_6 \ln(\dot{\epsilon}/\epsilon_r) + a_7 (\ln(\dot{\epsilon}/\epsilon_r))^2 \right] (T - T_r) \right] \quad (8)$$

The model fitted by the power function is shown below:

$$\sigma = (a_1 + a_2\epsilon^{a_3}) [1 + a_4 \ln(\dot{\epsilon}/\epsilon_r)] \exp \left[\left[a_5 + a_6 \ln(\dot{\epsilon}/\epsilon_r) + a_7 (\ln(\dot{\epsilon}/\epsilon_r))^2 \right] (T - T_r) \right] \quad (9)$$

When $T = 593.15 \text{ K}$ and $\dot{\epsilon} = 0.001 \text{ s}^{-1}$, Equations (8) and (9) can be simplified as Equations (3) and (4), respectively. According to the corresponding data from compressive stress-strain curves, the values of a_1 , a_2 and a_3 in Equations (8) and (9) were 74.19562, -30.20233 , 17.79643 and 122.49779, -62.66139 and 0.09106, respectively.

When $T = 593.15 \text{ K}$, Equations (8) and (9) can be simplified as follows:

$$\sigma = (a_1 + a_2\epsilon + a_3\epsilon^2) [1 + a_4 \ln(\dot{\epsilon}/\epsilon_r)] \quad (10)$$

$$\sigma = (a_1 + a_2\epsilon^{a_3}) [1 + a_4 \ln(\dot{\epsilon}/\epsilon_r)] \quad (11)$$

Based on the data of strain rates of 0.001 s^{-1} , 0.01 s^{-1} , 0.1 s^{-1} and 1 s^{-1} , the parameter a_4 in Equations (10) and (11) were 0.156564 and 0.156746, respectively. Moreover, the parameters a_5 , a_6 and a_7 in Equations (8) and (9) were -0.01255 , 0.0021 and -0.000158905 , respectively. The detailed parameters of the modified Johnson-Cook constitutive model are listed in Table 1.

Table 1. The modified Johnson-Cook constitutive model parameters.

Model	A1	A2	A3	A4	A5	A6	A7	A8	A9	A10
Polynomial of second order	74.1956	-30.2023	17.7964	0.156564	-0.01255	0.0021	-0.000158905	-	-	-
Polynomial of third order	73.8173	-23.176	-5.02855	19.0750	0.156445	-0.01255	0.0021	-0.000158905	-	-
Polynomial of fourth order	72.2395	17.4164	-235.842	466.424	-276.519	0.156543	-0.01255	0.0021	-0.000158905	-
Polynomial of fifth order	69.6888	103.5759	-993.862	3003.928	-3835.08	1764.705	0.1563425	-0.01255	0.0021	-0.000158846
Power-exponential function	122.498	-62.6614	0.09106	0.156746	-0.01255	0.0021	-0.000158905	-	-	-

The prediction accuracy of the constitutive model can be judged by the correlation coefficient R and average relative error $AARE$, as shown below.

$$R = \frac{\sum_{i=1}^N (\sigma_{ei} - \bar{\sigma}_e)(\sigma_{pi} - \bar{\sigma}_p)}{\sqrt{\sum_{i=1}^N (\sigma_{ei} - \bar{\sigma}_e)^2} \sqrt{\sum_{i=1}^N (\sigma_{pi} - \bar{\sigma}_p)^2}} \quad (12)$$

$$AARE(\%) = \frac{1}{N} \sum_{i=1}^N \left| \frac{\sigma_{ei} - \sigma_{pi}}{\sigma_{ei}} \right| \times 100\% \quad (13)$$

The closer the value of R is to 1, the smaller the value of $AARE$, the more accurate the prediction of the model. Comparing the predicted data of the constitutive model with the actual compression stress–strain curve, the values of R for the model fitted by a second-order polynomial and a power-exponential function were 0.9639 and 0.9835, respectively, and the corresponding values of $AARE$ were 0.106722 and 0.10648, respectively. That is, the prediction of the constitutive model fitted by a power-exponential function was more accurate than that fitted by a second-order polynomial. Figures 5 and 6 show the comparison between the prediction of the constitutive model fitted by a second-order polynomial and the actual compression stress–strain curves and the comparison between the prediction of the constitutive model fitted by a power-exponential function and the actual compression stress–strain curves.

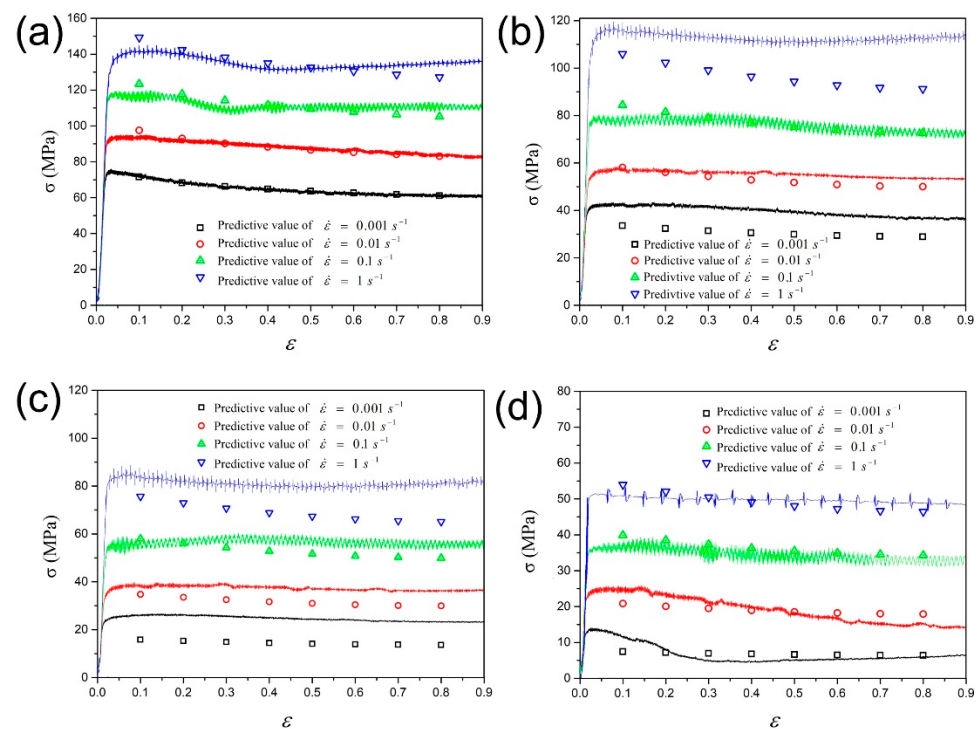


Figure 5. Comparison between the predictive value of the constitutive model fitted by a second-order polynomial and the experimental stress–strain curves at temperature of (a) 320 °C, (b) 380 °C, (c) 440 °C and (d) 500 °C, respectively.

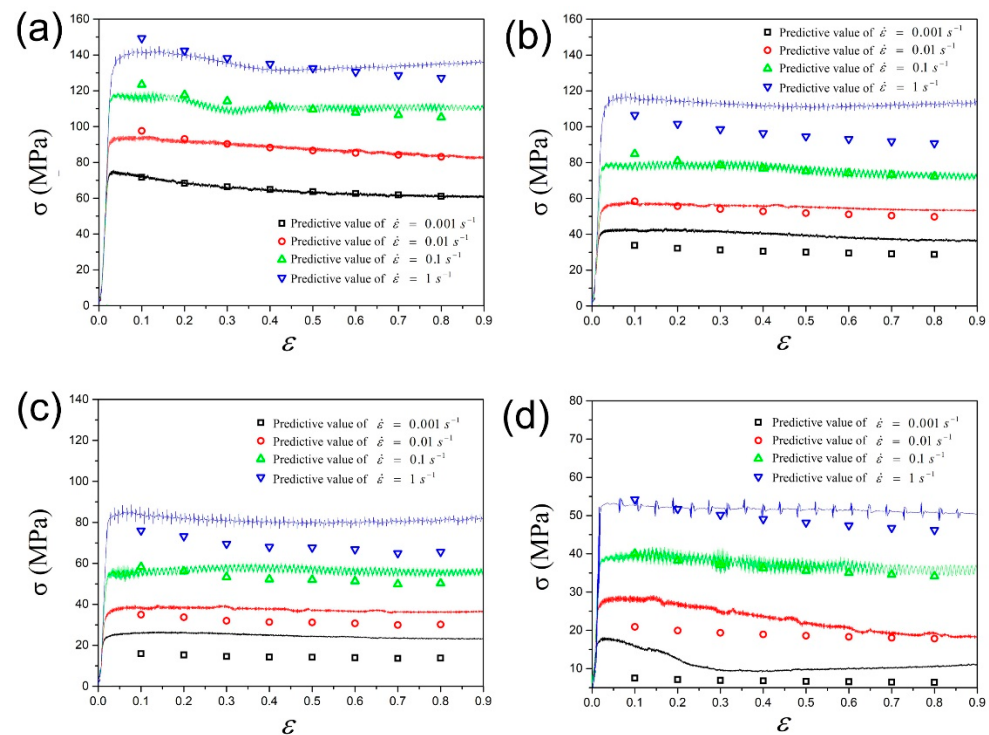


Figure 6. Comparison between the predictive value of the constitutive model fitted by a power-exponential function and the experimental stress-strain curves at temperature of (a) 320 °C, (b) 380 °C, (c) 440 °C and (d) 500 °C, respectively.

In order to improve the prediction accuracy, the constitutive model fitted by a higher-order polynomial was developed. The stress was expressed as a polynomial function of the third, fourth and fifth order of the strain, and the corresponding constitutive models are shown in Equations (14)–(16), respectively.

$$\sigma = \left(a_1 + a_2 \varepsilon^1 + a_3 \varepsilon^2 + a_4 \varepsilon^3 \right) \left[1 + a_5 \ln(\dot{\varepsilon}/\varepsilon_r) \right] \exp \left[\left[a_6 + a_7 \ln(\dot{\varepsilon}/\varepsilon_r) + a_8 (\ln(\dot{\varepsilon}/\varepsilon_r))^2 \right] (T - T_r) \right] \quad (14)$$

$$\sigma = \left(a_1 + a_2 \varepsilon^1 + a_3 \varepsilon^2 + a_4 \varepsilon^3 + a_5 \varepsilon^4 \right) \left[1 + a_6 \ln(\dot{\varepsilon}/\varepsilon_r) \right] \exp \left[\left[a_7 + a_8 \ln(\dot{\varepsilon}/\varepsilon_r) + a_9 (\ln(\dot{\varepsilon}/\varepsilon_r))^2 \right] (T - T_r) \right] \quad (15)$$

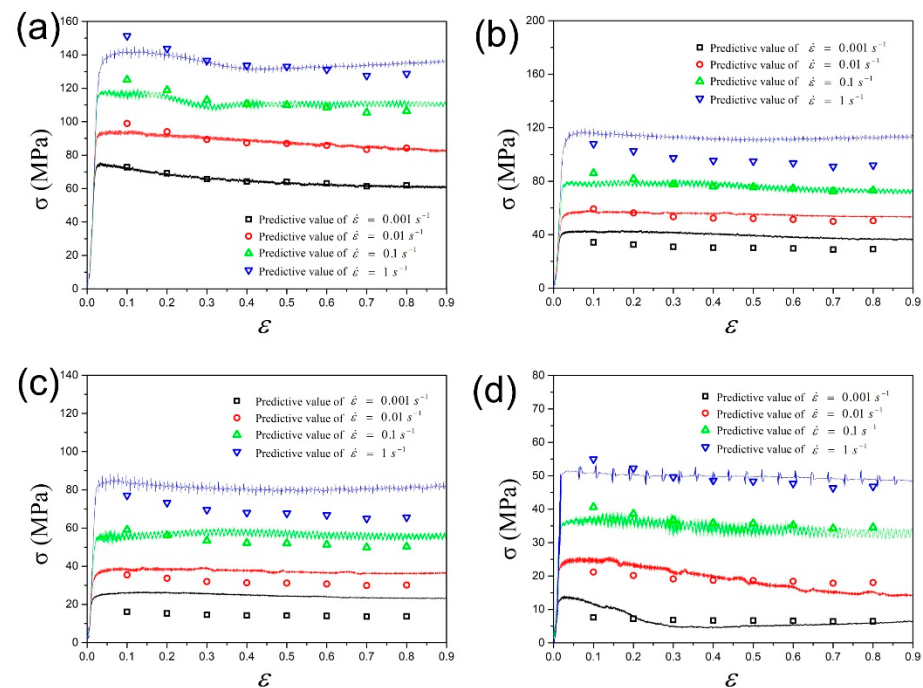
$$\sigma = \left(a_1 + a_2 \varepsilon^1 + a_3 \varepsilon^2 + a_4 \varepsilon^3 + a_5 \varepsilon^4 + a_6 \varepsilon^5 \right) \left[1 + a_7 \ln(\dot{\varepsilon}/\varepsilon_r) \right] \exp \left[\left[a_8 + a_9 \ln(\dot{\varepsilon}/\varepsilon_r) + a_{10} (\ln(\dot{\varepsilon}/\varepsilon_r))^2 \right] (T - T_r) \right] \quad (16)$$

The values of R and AARE of the constitutive models fitted by different higher-order polynomials are shown in Table 2. It can be concluded that the constitutive models fitted by higher-order polynomials were more accurate than the models fitted by a second-order polynomial and a power-exponential function. Although the prediction accuracy of the constitutive model can be improved by increasing the order of the fitting polynomial, the improvement of prediction accuracy is limited when the order of the polynomial exceeds four. Moreover, the model becomes more complex as the number the parameters of the model increases. Therefore, the constitutive model fitted by a fourth order polynomial was chosen in this work.

Table 2. The values of R and AARE of constitutive models.

Models	Polynomial of 2nd Order	Polynomial of 3rd Order	Polynomial of 4th Order	Polynomial of 5th Order	Power-Exponential Function
AARE (%)	10.6722	9.6964	8.7812	8.7238	10.6481
R	0.9639	0.9842	0.9843	0.9847	0.9845

Figure 7 shows the comparison between the predictive value of the constitutive model fitted by a fourth-order polynomial and the experimental compression stress–strain curve. Figure 8 shows the correlation between the predicted flow stress value of the model fitted by a fourth-order polynomial and the experimental value. According to the comparison between the predicted and experimental values, the modified Johnson–Cook constitutive model could predict the flow stress during the isothermal compression and behaved better in the high-temperature zone (around 500 °C) and in the low-temperature zone (around 320 °C) than that in the medium-temperature zone (380~440 °C).

**Figure 7.** Comparison between the predictive value of the constitutive model fitted by a fourth-order polynomial and the experimental stress–strain curves at temperature of (a) 320 °C, (b) 380 °C, (c) 440 °C and (d) 500 °C, respectively.

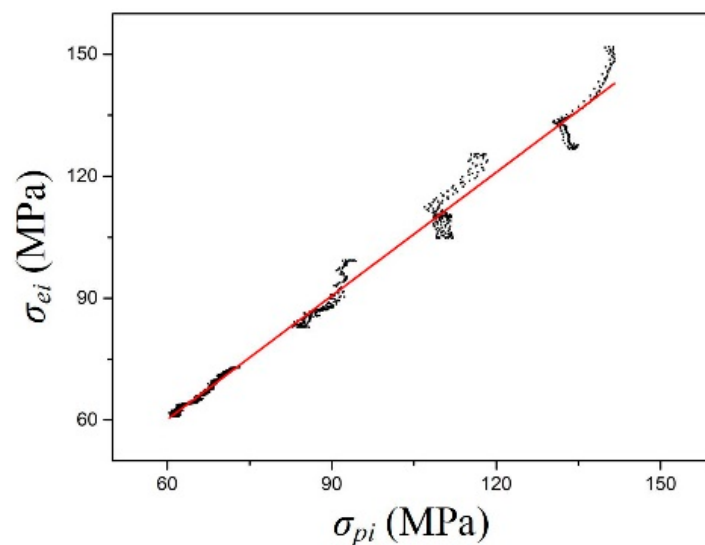


Figure 8. The correlation between the predicted flow stress value of the model fitted by a fourth-order polynomial and the experimental value.

3.2. Microstructures of the Compressed Samples

Figure 9 shows the XRD patterns of the original sample and the compressed samples with different deformation conditions. The selected area electron diffraction (SAED) patterns of the T_1 phase and β' phase were obtained in the compressed sample at $440\text{ }^\circ\text{C}/0.001\text{ s}^{-1}$ from a previous report [25]. The original sample contained the T_1 phase (Al_2CuLi) and β' (Al_3Zr) phase, while all the compressed samples contained the T_2 phase (Al_6CuLi_3) besides the T_1 phase and the β' phase. It is known that the T_1 phase is the predominant precipitate in Al-Cu-Li alloys, which has been widely investigated and forms as thin hexagonal plates with a $\{111\}_{\text{Al}}$ matrix habit plane [1–4]. The metastable Al_3Zr phase always presents spherical particles in third-generation Al-Li alloys and is an important precipitate in the Al-Cu-Zr system, which has a $\text{Li}_2 \rightarrow \text{D0}_{23}$ structural transformation with a Cu addition [26]. The T_2 phase is a stable intermetallic (icosahedral, point group $m\bar{3}5$) forming via solid-state phase transformation on HAGBs, which may be detrimental to the toughness of the Al-Li alloy [37]. Figure 10a shows the schematic precipitate microstructure of the precipitates in compressed samples [9].

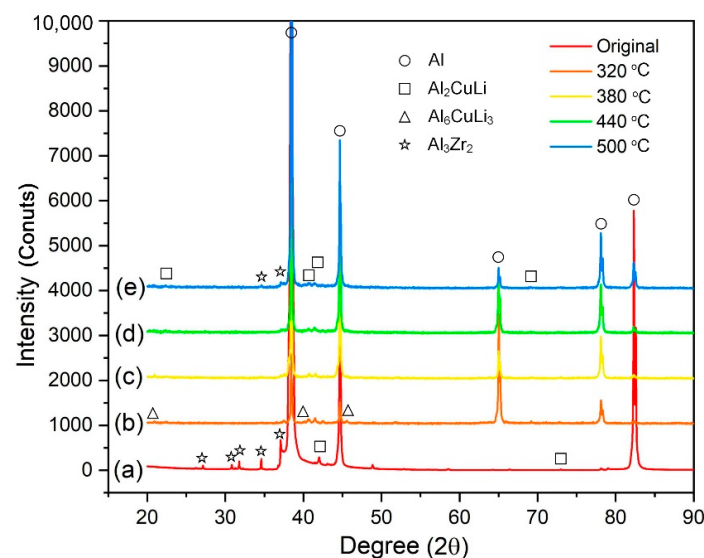


Figure 9. The XRD patterns the deformed alloys: (a) original sample, compressed sample: (b) $320\text{ }^\circ\text{C}/0.001\text{ s}^{-1}$; (c) $380\text{ }^\circ\text{C}/0.001\text{ s}^{-1}$; (d) $440\text{ }^\circ\text{C}/0.001\text{ s}^{-1}$; (e) $500\text{ }^\circ\text{C}/0.001\text{ s}^{-1}$.

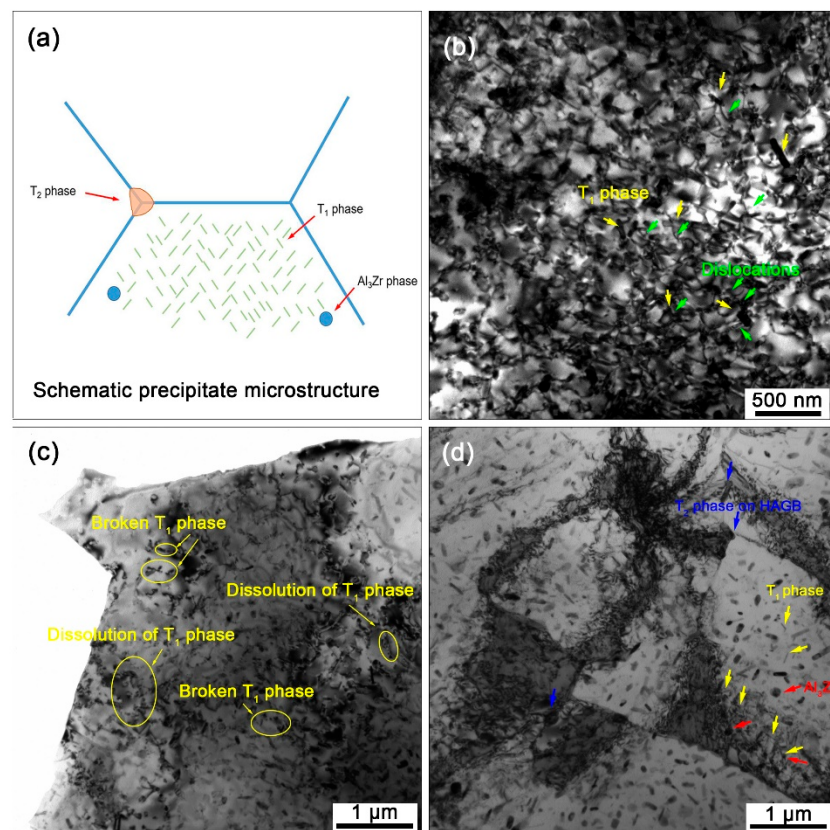


Figure 10. (a) Schematic precipitate microstructure of the compressed samples; (b) strong interaction between dislocation and T_1 phase in sample at 320 °C/0.001 s^{−1}; (c) fracture and dissolution of T_1 phase in sample at 380 °C/0.001 s^{−1}; (d) the BF TEM image of representative precipitates in compressed sample at 320 °C/0.001 s^{−1}.

When the deformation temperature is low (320 °C), plenty of dislocations are introduced into the grains and the strong interaction between the T_1 phase and dislocations are also observed, as shown in Figure 10b. The green arrows indicate dislocations, and the yellow ones indicate the T_1 phase. The dislocation motion is impeded by precipitates distributing like a network. On the other hand, part of the T_1 precipitates are distorted or broken by the dislocation movement, as shown in Figure 10c. The report from Sha et al. [38] reveals that plastic deformation could change the orientation between the precipitates and the matrix slightly, as well as the atomic structure of the interface, resulting in a growth of the interface mismatch energy. In addition, the plastic deformation can increase the distortion energy of the alloy. The surface energy of the precipitates also increases with precipitates being broken. The combined effects of the above factors result in the rise of free energy and redissolution of the T_1 phase, which induces the transformation of the T_1 phase from rodlike to spherical, as shown in Figure 10c. The redissolution of the T_1 phase make the Al matrix become a supersaturated solid solution again followed by a secondary precipitation of the precipitates.

Moreover, the dislocations induced by deformation provide heterogeneous nucleation positions for the T_1 phase nucleation. The stress field caused by the interaction between the dislocation and T_1 phase also provides the required energy for the T_1 phase nucleation. The short-circuit diffusion channel provided by the dislocations reduces the atomic diffusion activation energy, which is beneficial to increase the precipitates' nucleation rate [39]. A former research [40] showed that further deformation of the precipitates during the process of redissolution might lead to a secondary precipitation of the precipitates. It can be observed that the fine T_1 phase reprecipitated around the dislocation and subgrain boundary as indicated by the yellow arrows in Figure 10d. In addition, former investigations [41]

suggested that Al_3Zr particles and the dislocation loops around the Al_3Zr phase could provide heterogeneous nucleation positions for the T_1 phase nucleation, which can be observed in Figure 10d.

The T_2 phase was found to be stable to about 420 °C and might reprecipitate at the high-angle grain boundaries (HAGBs), preferentially during the aging of the 8090 alloy [37]. However, in this work, the T_2 phase only existed in the compressed sample but not in the original sample according to the results of XRD. The temperature of deformation ranged from 320 °C to 440 °C, which was suitable for the reprecipitation of the T_2 phase. High-angle grain boundaries generated by dynamic recrystallization provided positions for the T_2 phase nucleation. The distortion energy caused by the deformation provided the driving force for the T_2 phase nucleation. Therefore, it was a reasonable result that the dynamic precipitation of the T_2 phase took place during hot compression. As shown in Figure 10d, the coarse T_2 particles can be observed at the HAGBs indicated by blue arrows.

Figure 11 shows the EBSD IPF maps of the compressed samples with deformation conditions of 320 °C/1 s^{−1}, 320 °C/0.001 s^{−1}, 440 °C/1 s^{−1} and 440 °C/0.001 s^{−1}. Figure 12 shows the misorientation angle distribution histograms of the samples. It can be observed that a large fraction of low-angle grain boundaries (<15°, LAGBs) existed in all the samples. With the deformation temperature increasing and strain rate decreasing, the frequency of HAGBs increased obviously. Furthermore, the sample of 440 °C/0.001 s^{−1} had a maximum frequency of HAGBs of 21.6% and a maximum average of misorientation of 9.91°. It could effectively promote DRX to increase the temperature or decrease the strain rate by providing sufficient energy or time to facilitate dislocation rearrangement and forming DRXed grains.

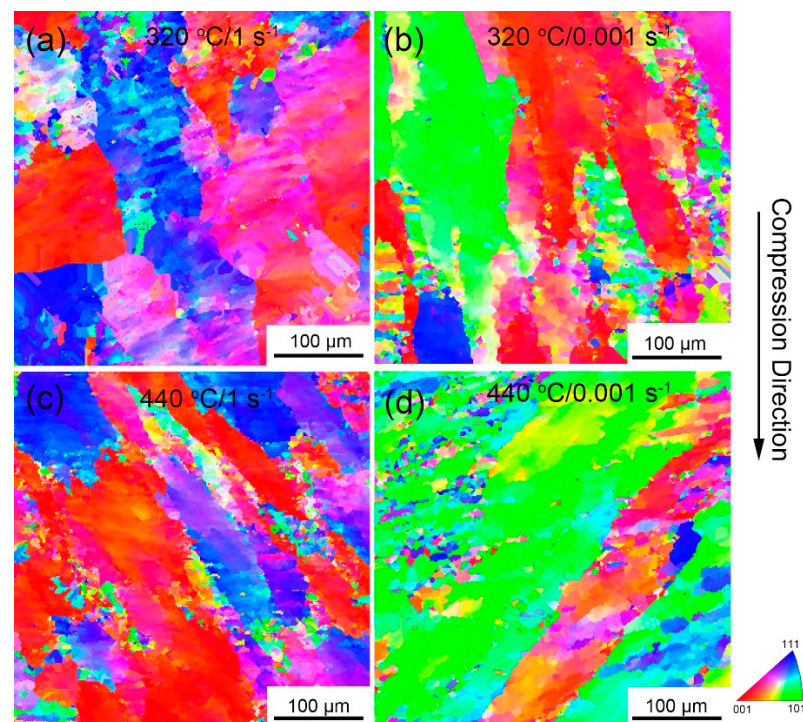


Figure 11. EBSD IPF images of compressed samples: (a) 320 °C/1 s^{−1}; (b) 320 °C/0.001 s^{−1}; (c) 440 °C/1 s^{−1}; (d) 440 °C/0.001 s^{−1}.

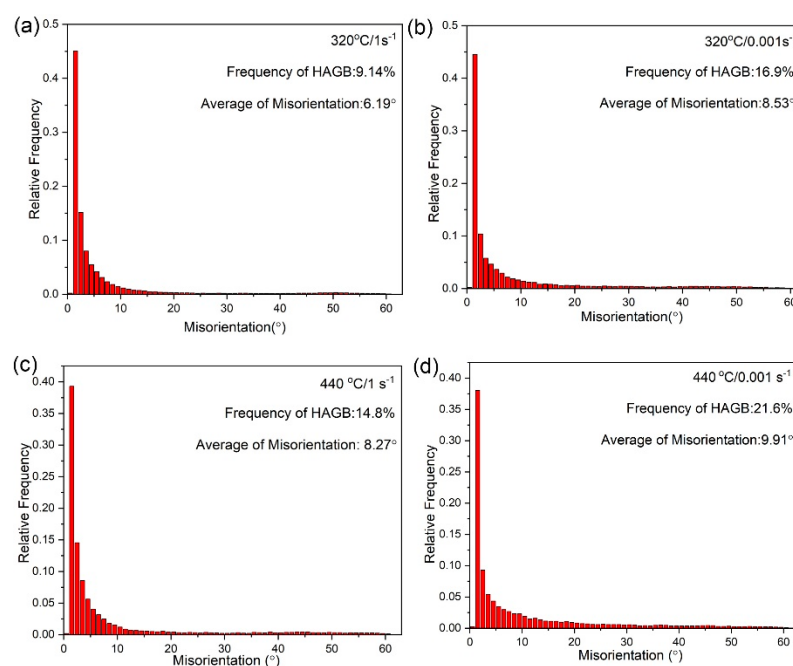


Figure 12. Misorientation angle distribution histograms of the samples: (a) 320 °C/1 s^{−1}; (b) 320 °C/0.001 s^{−1}; (c) 440 °C/1 s^{−1}; (d) 440 °C/0.001 s^{−1}.

As mentioned above, the second-phase particles may retard or accelerate dynamic recovery (DRV) and dynamic recrystallization (DRX) depending on their size and spatial distribution. The large particles (>1 μm in diameter) can promote DRX via PSN. A recent report from Wang et al. [41] revealed that a fraction of recrystallization of 67% in an Al-Cu-Li alloy was triggered by a sparse coarsened T_1 phase via PSN during hot compression at 380 °C/0.01 s^{−1}. There were massive dense and fine T_1 phase precipitates in the Al-Cu-Li alloys in this work, which effectively pinned dislocations (as shown in Figure 10b) or subgrain boundaries to accelerate DRV but suppress DRX. Therefore, a very small fraction of HAGBs existed in all the compressed samples.

4. Conclusions

In this work, the isothermal compression behaviors of as-extruded AA 2055 alloy were studied. A modified Johnson–Cook model was established to describe the flow stress of the alloy. The microstructures of compressed samples were investigated via XRD, EBSD and TEM. Several conclusions could be obtained, as shown below.

- (1) The constitutive model fitted by a fourth-order polynomial was chosen for the optimal modified Johnson–Cook constitutive model in order to balance the prediction accuracy and model complexity.
- (2) The modified Johnson–Cook constitutive model could predict the flow stress of the as-extruded AA 2055 alloy during isothermal compression well, especially in the high-temperature zone (around 500 °C) and the low-temperature zone (around 320 °C).
- (3) The massive dense fine precipitates effectively pinned dislocations or subgrain boundaries to accelerate DRV but suppress DRX. An unusual dynamic precipitation of the T_2 phase was observed in the compressed samples.

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