Hot deformation behavior and microstructure characteristic of 2055 Al-Li alloy during uniaxial compression

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Abstract
The hot deformation behavior of as-extruded 2055 Al-Li alloy at T6 state has been investigated by uniaxial hot compression test in a Gleeble-3800 thermal simulation machine at the temperature of 320 °C, 380 °C, 440 °C, 500 °C with strain rate ranging from 0.001 s−1 to 1 s−1. Deformation instability occurred under the deformation condition of 380 °C/5 s−1 and at the deformation temperature of 500 °C via the shape of flow curve and the side view observation of the compressed samples. The serration in flow curve of the low strain rate suggests occurring of DSA. The constitutive equation based on Arrhenius equation were established with the deformation activation energy of 174.6 kJ mol−1. The microstructures of the compressed samples were analyzed by OM and TEM, the DRXed grains could be observed in the compressed samples at 440 °C with strain rate of 1 s−1 and 0.001 s−1. The dominated precipitates in the alloy were plate-shaped T1 phase and spherical β1 phase. The precipitates-dislocations interaction was analyzed by HRTEM. The dislocations could be pinned by both the T1 phase and β1 phase, and the precipitates and dislocations around them divided the grain into small zones, and the misorientation across the dislocation was about 0.8°.

1. Introduction
Aluminum lithium alloys have been widely used in aerospace industry due to their low density, high elastic modulus, high specific strength and stiffness, and good mechanical performance at room temperature [1–6]. 2055 Al-Li alloy is the latest product of the third generation aluminum-lithium alloy developed by Aluminum Company of America (Alcoa) in 2012 [7, 8]. 2055 Al-Li alloy has higher fracture strength and fracture toughness than other third generation Al-Li alloys without increasing cost by minor adjustments of type and content of alloying elements [8].

The aviation components are always processed by plastic deformation, therefore, the researches about hot deformation behavior of Al-Li alloys have been frequently reported for the past few years. Due to the alloying of Cu, Zr and the other elements, the Al-Li alloys always contain a sophisticated combination of precipitations depending on different heat treatment and content of the alloying elements [9], which gives the Al-Li alloys noteworthy precipitation strengthening potential. The T1 phase is the important precipitate in Al-Cu-Li system, which is hexagonal, of symmetry p6/mmm, forms as discs on the {111} plane of Al matrix [10]. The relationship between precipitates and deformation behavior in Al-Li alloys also attracted considerable attention, which can be divided into two situations, i.e. deformation prior to precipitation and precipitation prior to deformation.

In general, artificial aging after deformation can improve strength of the Al-Li alloy due to precipitating strengthening relying on the alloy composition and heat treatment process. It was proposed that pre-strain before artificial aging could accelerate precipitation [11, 12]. An Al-Cu-Li alloy processed by equal channel angular pressing (ECAP) was reported by M A Munoz-Morris et al, in which the fine precipitates enhanced strength–ductility combinations by delaying strain localization and fracture [13]. Recent investigation from
Wang et al indicated that pre-stretch prior to aging treatment promoted fine T1 phase precipitating throughout the matrix to narrow the PFZ (precipitation free zone) and inhibit the growth of GBPs (grain boundary precipitation), which resulted in simultaneous improvement of strength and elongation in an extruded Al-Cu-Li alloy [14].

The effect of precipitates on deformation behavior also has been investigated in recent years in Al-Li alloy. Some reports revealed that T1 phase in Al–Cu–Li alloys could homogenized plastic flow than other shearable precipitates [10, 15, 16], which was attributed to the fact that the T1 phase was hardly shearable. Nevertheless, the other investigation suggested that this precipitate could be sheared by dislocation [17]. In the report from M. Lewandowska et al [15], the massive dislocation loops formed around T1 phase during a fatigue test of Al-Li alloys. And Aladar et al also proposed localized plasticity in the matrix accompanying the shearing of T1 phase [17]. Therefore the interaction between T1 phase and dislocation need to be illustrated further.

The effect of precipitates on deformation behavior is the interaction between precipitates and dislocations essentially, and also is related to the dynamic recovery (DRV) and dynamic recrystallization (DRX) of the Al-Li alloy during hot deformation. Former researches revealed that the dispersed fine precipitates could enhance the tendency of DRV [18, 19] but suppress DRX of the Al-Li alloys by inhibiting the rearrangement of dislocations and sub-grain boundary migration [19–22]. Whereas the large particles (> 1 μm in diameter) could promote the DRX by producing regions of high lattice misorientation, i.e. particle-stimulated nucleation (PSN) in Al alloys [23, 24].

In fact, artificial aging prior to deformation attracted increasing attention in the metal and alloys, not only in Al alloy, but also in Mg alloys. Latest research revealed that the interaction between precipitates and dislocation or twin could improve the strength of wrought Mg alloy [25–27].

The construction of constitutive model is of great significance to analyze the formability of the metals. Recently, lot of reports about constitutive model or flow curve of the Al-Li alloys [3, 10, 18, 19, 22, 28, 29]. However, there are few reports on constitutive model of 2055 Al-Li alloy so far. In this paper, the flow curves of a commercial as-extruded 2055 Al-Li alloy (T6) at 0.001–1 s\(^{-1}\) / 320 °C–500 °C were obtained experimentally, and the Arrhenious equation based on hyperbolic sinusoidal model was established. The effect of precipitates on plastic deformation behavior and interaction between precipitates and dislocation were also discussed in this work.

2. Materials and Methods

The Al-Li alloy in this work was commercial as-extruded 2055 Al-Li alloy at T6 state. The chemical composition of the alloy was provided by the supplier as shown in table 1. The X ray diffraction (XRD) analysis of the original sample was carried out, the result was shown in figure 2. It could be concluded that the dominant second phase was Al\(_2\)CuLi phase (T1 phase). The samples for compression experiment were machined by a CNC wire-cut electric discharge machine into cylinders with a size of diameter 10 mm × length 15 mm. The external surface of the sample was polished by sandpaper in order to obtain good surface roughness. The thermal compression experiment was carried out on a Gleeble-3800 thermal simulator (Gleeble-3800; St. Paul, USA). The heating rate was 5 K s\(^{-1}\), the holding time was 3 min The experimental temperatures were 320 °C, 380 °C, 440 °C and 500 °C and strain rates ranged from 0.001 s\(^{-1}\) to 1 s\(^{-1}\). The sample was immediately quenched by water after hot compression with height reduction of 60%. The samples for optical microscopy (OM; Olympus GX53,
Tokyo, Japan) were taken from the longitudinal section of the compressed samples. The samples were firstly polished by SiC abrasive paper, then were precisely polished by Al₂O₃ turbid liquid, finally were etched by Keller’s reagent (2.5 mlHNO₃ + 1.5 mlHCL + 1 mlHF + 95 mlH₂O). The foils of the compressed specimen were taken for transmission electron microscopy (TEM; JEOL-F200, Tokyo, Japan) analysis. The foils for TEM were first mechanically polished to 30 μm in thickness, and then thinned by ions beam (Leica RES101, Wetzlar, Germany). TEM observation was conducted at an accelerating voltage of 200 kV. The diagrammatic sketch of the experimental procedure is shown in figure 1.

3. Results and Discussions

3.1. Flow curves

The true stress-true strain curves of 2055 Al-Li alloy are shown in figure 3. It can be concluded that the strain rate has a significant impact on the flow stress of the alloy. When the temperature is constant, the flow stress of the alloy increases remarkably with the strain rate increasing. Likewise, deformation temperature has a significant effect on the flow stress of the alloy. When strain rate is constant, the flow stress decreases gradually with the deformation temperature increasing.

All the flow curves can be divided into the elastic deformation stage and the plastic deformation stage. In the elastic deformation stage, the flow stress increases linearly with strain increasing. While in the plastic deformation stage, the value of flow stress increases to a peak at first (the corresponding stress is called peak stress p), then gradually decreases and tends to level, and finally reaches a steady state (the corresponding value of stress is called steady state stress s). In the plastic deformation stage, the change tendency of flow stress is the result of competition between hardening mechanism and softening mechanism of 2055 Al-Li alloy during the hot compression.

In the initial stage of plastic deformation, the dislocation density in the alloy increases rapidly. When the dislocation density reaches a critical value, the cross slip of dislocations will be activated, and the dislocations will annihilate when the positive dislocations meet the negative ones during the process of cross-slip, resulting in softening of the alloy. However, before the stress reaches the peak value, the work hardening is the dominant mechanism, and the softening caused by cross slip is not enough to overcome the work hardening, thus the flow stress increases rapidly. As the strain increasing, dynamic recovery (DRV) and dynamic recrystallization (DRX) will occur until the accumulated value of dislocation density exceeds the critical value required for DRV and DRX. The occurrence of DRV and DRX results in rearrangement of extensive intertangled dislocations in the grain then forming a new interface, which can sharply reduce the dislocation density in the grain of the alloy and soften the Al-Li alloy significantly [30, 31]. At this stage, the softening resulting from DRV and DRX can be
competitive with the work hardening, therefore, the work hardening does not dominate the hot deformation any more.

The flow stress decreases slightly then keeps relatively stable due to dynamic balance between softening and hardening. It can be observed that the shape of true stress-strain curve differs from each other depending on the condition of deformation, as shown in figure 3. For instance, the curves of 320 °C/1 s\(^{-1}\) and 500 °C/0.01 s\(^{-1}\) have obvious single broad peak suggesting remarkable dynamic softening during hot deformation, while the curves of 380 °C/0.01 s\(^{-1}\) and 440 °C/0.001 s\(^{-1}\) rise to a plateau without obvious falling. The classical flow curves containing single peak are regarded as the indication of DRX [30, 31]. And the characteristic of monotonically rising to a plateau stress always suggests that the alloy underdoes only DRV [30, 31]. In general, the dominant dynamic softening mechanism of aluminium alloy is DRV at low temperature, and DRX tends to be found occurrence at high temperature with low strain rate. The phenomenon observed in this work is inconsistent with the general rules, which agrees with recent report about flow curves of Al-Li alloy [3]. Therefore, the occurrence of DRX in Al-Li alloy may not be only evaluated by flow curves.

It can be concluded from figure 3 that the peak stress and the corresponding strain increase with strain rate increasing, which is mainly attributed to the propagation or annihilation of the dislocation in the 2055 Al-Li alloy. On one hand, with strain rate increasing, the dislocation propagation rate also increases sharply, the strong work hardening of the alloy makes the peak stress higher. On the other hand, the increasing of strain rate shortens the evolution of DRV and DRX in the grain of the alloy. There is no adequate time for dislocation rearrangement forming low angle grain boundary or low angle grain boundary (LAGB) transforming to high angle grain boundary (HAGB), therefore, it needs larger strain that the softening rate resulting from DRV and DRX can catch up with work hardening rate, i.e. the strain corresponding to the peak stress increases [30].

Figure 4 shows the flow curve of 2055 Al-Li alloy at 380 °C with strain rate of 5 s\(^{-1}\). The true strain of the sample is only 0.27 suggesting occurrence of deformation instability. The research from Ou et al showed that the optimum hot deformation condition for 2060 Al-Li alloy was 380 °C–500 °C and 0.01–3 s\(^{-1}\) [22]. Therefore, it
may be concluded that the strain rate of 3–5 s\(^{-1}\) (or higher) is not appropriate for hot deformation of Al-Li alloy. The flow curve of the sample at 500 °C with strain rate of 0.001 s\(^{-1}\) in figure 3(d) is also abnormal. Figure 5 shows the side view of the compressed samples of the 2055 Al-Li alloy, it can be seen that deformation instability occurs in all the compressed samples at 500 °C, which indicates that the deformation temperature should not be over 500 °C for the 2055 Al-Li alloy, which agrees with the research from Ou et al [22]. The serration in the flow curves should be noteworthy in figure 3. The obvious waves in the flow curves at higher strain rate (1 s\(^{-1}\) or 0.1 s\(^{-1}\)) should be caused by experimental equipment. While the serration in the flow curves at lower strain rate (0.01 s\(^{-1}\) or 0.001 s\(^{-1}\)) also can be observed. Figure 6 shows the enlarged flow curves corresponding to figure 3(a). The serration surrounded by yellow hollow ellipses should be caused by dynamic strain aging (DSA), which is a common phenomenon in Al alloy [32, 33]. Ignoring the influence of experimental equipment, the serration caused by DSA is more obvious in the flow curves at lower strain rate. The result agrees well with recent report from Valdes-Tabernero et al that DSA is suppressed at higher strain rate owing to lack of time for solute atoms to ‘arrest’ gliding dislocations [32].

3.2. Constitutive relationship

During hot plastic deformation, the strain rate of the alloy is controlled by a thermal activation process, i.e. Arrhenius equation. In order to describe the relationship between flow stress, deformation temperature and strain rate, Sellars and McTegart [34] have proposed the Arrhenius equation in hyperbolic sinusoidal form empirically, i.e.
\[ \dot{\varepsilon} = A \sinh (\alpha \sigma)^n \exp \left( -\frac{Q}{RT} \right), \]  
\[ \text{where } Q \text{ is the activation energy for hot deformation (kJ mol}^{-1} \text{), } R \text{ is the gas constant (8.314 J mol}^{-1}\text{K}^{-1} \text{), } T \text{ is deformation temperature (K), } n \text{ and } A \text{ are the material constants, } \alpha \text{ is the stress multiplier.} \]

When the stress level is high [35], the equation can be expressed by equation (2).

\[ \dot{\varepsilon} = A_1 \sigma^{n_1} \exp \left( -\frac{Q}{RT} \right) \]  

When the stress level is low [35], the equation can be expressed by equation (3).

\[ \dot{\varepsilon} = A_2 \exp (\beta \sigma) \exp \left( -\frac{Q}{RT} \right) \]

where \( n_1, A_1, A_2 \) and \( \beta \) are the material constants independent of deformation temperature and strain rate.

According to equation (3), if the values of \( \alpha, n, Q \) and \( A \) can be solved out, the constitutive equation can be obtained.

To simplify the equations, the natural logarithms was applied of both side of equations (1)–(3), and equations (1)–(3) can be expressed as below, respectively.

\[ \ln \dot{\varepsilon} = \ln A + n \ln \sinh (\alpha \sigma) - \frac{Q}{RT} \]  
\[ \ln \dot{\varepsilon} = \ln A_1 + n_1 \ln \sigma - \frac{Q}{RT} \]  
\[ \ln \dot{\varepsilon} = \ln A_2 + \beta \sigma - \frac{Q}{RT} \]

In the present work, the peak stress is used in the \( \sigma \) term. The plot of \( \ln \dot{\varepsilon} \) Versus \( \ln \sigma \) has been presented in figure 7(a). According to equation (5), a liner relationship exists between \( \ln \dot{\varepsilon} \) and \( \ln \sigma \) with a slope of \( n_1 \) Likewise, the plot of \( \ln \dot{\varepsilon} \) Versus \( \ln \sigma \) has been presented in figure 7(b). According to equation (6), a liner relationship also exists between \( \ln \dot{\varepsilon} \) and \( \sigma \) with a slope of \( \beta \). The values of the peak stress and corresponding strain are taken to calculate the value of \( n_1 \) and \( \beta \) by using a linear regression and taking the average value of slopes. And the values of \( n_1 \) and \( \beta \) are 7.066 and 0.1177, respectively. Then the value of the stress multiplier \( \alpha \) can be calculated as \( \alpha = \beta / n_1 = 0.0167 \).

The hot plastic deformation of the Al-Li alloy is a process of thermal activation. The activation energy \( Q \) is defined as the minimum energy required to activate the plastic deformation at defined strain rate and deformation temperature. It is an important physical parameter to evaluate the difficulty of the plastic deformation of the alloy. The expression of the activation energy \( Q \) can be obtained by taking the partial differential of equations (4)–(6), as shown below,

\[ Q = Rn \frac{\partial[\ln \sinh (\alpha \sigma)]}{\partial(1/T)} \bigg|_{\dot{\varepsilon}} = Rn_1 \frac{\partial(\ln \dot{\varepsilon})}{\partial[\ln \sinh (\alpha \sigma)]} \bigg|_{T} \frac{\partial[\ln \sinh (\alpha \sigma)]}{\partial(1/T)} \bigg|_{\dot{\varepsilon}} \]  

Figure 6. The enlarged flow curves corresponding to yellow rectangle zone of figure 2(a).
The values of \( \frac{\partial \ln \dot{\varepsilon}}{\partial \ln \sinh(\alpha\sigma)} \bigg| T \) and \( \frac{\partial \ln \sinh(\alpha\sigma)}{\partial (1/T)} \bigg| T \) can be obtained by taking slopes of plots of \( \ln \dot{\varepsilon} - \ln \sinh(\alpha\sigma) \) and \( \ln \sinh(\alpha\sigma) - 1/T \) respectively, and the plots are shown in figure 8(a) and (b). The value of \( \frac{\partial \ln \dot{\varepsilon}}{\partial \ln \sinh(\alpha\sigma)} \bigg| T \) (or \( n \)) can be obtained by a liner by using a linear regression and taking the average value of slopes, which is similar to above method, and the value of \( n \) is 4.8056. Similarly, the value of \( \frac{\partial \ln \sinh(\alpha\sigma)}{\partial (1/T)} \bigg| \dot{\varepsilon} \) can be obtained. Therefore, the value of activation energy \( Q \) is about 174.6 kJ mol\(^{-1}\). The activation energy of 2055 Al-Li alloy is higher than that for self-diffusion of pure Al (145 ~ 170 kJ mol\(^{-1}\)) [36]. Generally, the activation energy of Al-Li is higher than 200 kJ mol\(^{-1}\), such as 205 kJ mol\(^{-1}\) for 2060 Al-Li alloy [22], 236 kJ mol\(^{-1}\) for 2090 Al-Li alloy [37], 282 kJ mol\(^{-1}\) for the 8090 Al-Li alloy and 287 kJ mol\(^{-1}\) for Al-Li 8091 alloy [22, 38]. Some values of activation energy of the Al-Li alloys have been listed in table 2. The value of activation energy is influenced by lots of factors, such as composition of the alloy, the state of the alloy, the style of hot deformation and range of strain rate and temperature. Besides, precipitates or dispersoids can improve the activation energy via interaction between and dislocation to impede dynamic recovery [22]. The less activation energy of 2055 Al-Li alloy than those of other Al-Li alloys might be due to better deformability of the as-extruded Al-Li alloy at elevated temperature.

Former research [40] revealed that the relationship among hot deformation behavior, temperature and strain rate can be expressed by Zener-Hollomon parameter, i.e. \( Z \) parameter, which is an important parameter to reflect the effect of temperature and strain rate on hot deformation behavior of the alloy, as shown in equation (8).

\[
Z = \dot{\varepsilon} \exp \left( \frac{Q}{RT} \right)
\]  

\[(8)\]
Therefore, the value of \( Z \) can be evaluated by taking the value of activation energy \( Q \) and the specific strain rate \( \dot{\varepsilon} \) and deformation temperature \( T \) into equation (8). Substituting the strain rate \( \dot{\varepsilon} \) in the equation (8) by the equation (1), the \( Z \) parameter can be expressed as below [40].

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

(9)

The natural logarithms is applied of both side of equation (9), and equation (9) can be transformed into equation (10). The scatter plot of \( \ln Z \) Versus \( \ln \sinh(\alpha\sigma) \) has been plotted in figure 6, and a liner regression has been carried out. The values of \( \ln Z \) and \( \ln \sinh(\alpha\sigma) \) exhibit high linear correlation with a linearly dependent coefficient of 0.98911, as shown in figure 9. According to equation (10), the value of \( \ln A \) is the intercept of the fitted curve in figure 6 whose value is 26.6050. Thus, the value of \( A \) is \( 3.5843 \times 10^{11} \) s\(^{-1}\).

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

(10)

Based on the definition of hyperbolic sine function, i.e. \( \sinh x = 1/2(\exp x - \exp(-x)) \) and equation (10), the constitutive equation can be expressed as shown in equation (11). Substituting the measured values of \( A, \alpha, n \) and \( Q \) into the equation (8) and equation (11), the constitutive equation during hot compression of the 2055 Al-Li alloy is obtained as shown in equation (12) and equation (13).

\[
\sigma = \frac{1}{\alpha} \ln \left\{ \frac{Z}{A} \right\}^{1/n} + \left[ \left( \frac{Z}{A} \right)^{2/n} + 1 \right]^{1/2}
\]  

(11)

\[
Z = \dot{\varepsilon} \exp \left( \frac{174.6}{8.314T} \right)
\]  

(12)

\[
\sigma = \frac{1}{0.0167} \ln \left\{ \frac{Z}{3.5843 \times 10^{11}} \right\}^{1/4.8056} + \left[ \left( \frac{Z}{3.5843 \times 10^{11}} \right)^{2/4.8056} + 1 \right]^{1/2}
\]  

(13)

According to classical dynamic deformation criterion from Prasad [41], the effectiveness of power dissipation \( \eta \) and the strain rate sensitivity \( m \) can characterize the workability of a metal material, and the

| Al-Li alloys | The value of Q (kJ mol\(^{-1}\)) | State                      |
|-------------|-------------------------------|----------------------------|
| 1460        | 303.14 [19]                   | As-cast alloy for compression test |
| 2060        | 205 [22]                      | As-cast alloy for compression test |
| 2060        | 251.06 [39]                   | As-rolled alloy at T8 state for tensile test |
| 2090        | 236 [37]                      | As-cast alloy for compression test |
| 2090        | 219 [37]                      | As-cast alloy for torsion test |
| 8090        | 282 [38]                      | Solution-treated rolled plate for torsion test |
| 8091        | 287 [38]                      | Solution-treated rolled plate for torsion test |
| 2055        | 174.6                         | As-extruded alloy at T6 state for compression test |

Figure 9. Relationships between flow stress and strain rate of the experimental alloys: \( \ln Z - \ln \sinh(\alpha\sigma) \).
is the power which has been dissipated by plastic work, is dissipater power which is related to the pile-up, as shown in Figure 11. To compare

3.3. Microstructures

Where expressions are presented below.

\[
\eta = \frac{J}{f_{\max}} = \frac{P - G}{P/2} = 2\left(1 - \frac{G}{P}\right) = 2\left[1 - \frac{1}{\sigma_b} \int_0^\varepsilon \sigma \, d\varepsilon\right]
\]  

Where \( G \) is the power which has been dissipated by plastic work, \( J \) is dissipater power which is related to the metallurgical mechanisms taking place dynamically to dissipate power (2016, Mohammad). The strain rate sensitivity map and power dissipation map obtained at logarithmic strain of 0.6 are presented in Figure 10. As is seen, the delineations of the maps are highly similar, which agrees well with the reports before [42, 43]. While the negative strain rate sensitivity is not observed in the map which is different from the result of 7075 Al alloy [Mohgaddam, 2016 # 10981]. The negative value of strain rate sensitivity always displays in the zone of lower strain rate and temperature, and is related to DSA [Mohgaddam, 2016 # 10981].

3.3. Microstructures

Figures 11(a)–(d) show the OM micrographs of the compressed samples under deformation condition of 320 °C/1 s⁻¹, 320 °C/0.001 s⁻¹, 440 °C/1 s⁻¹, and 440 °C/0.001 s⁻¹ respectively. The coarse long compressed grains can be observed in all the OM micrographs. The microscopic characteristics of DRV or DRX can be hardly observed in figure 11(a) (320 °C/1 s⁻¹), while obvious DRXed characteristics exist in the microstructure of compressed sample of 320 °C/0.001 s⁻¹, as shown in figure 11(b). There are lots of sub-grain boundaries and some fine DRXed grain in figure 11(c) (440 °C/1 s⁻¹), and the original grain boundaries transform from straight ones to sawtooth ones. In figure 11(d), the DRXed grains have grown up obviously, and the original straight can be hardly observed. It can be concluded that both strain rate and temperature have significant influence on the microstructures of the 2055 Al-Li alloy during hot compression, which is consistent with the flow curves above.

To compare figure 11(a) against figure 11(b) (or figure 11(c) against figure 11(d)), the lower strain rate provides sufficient time for DRV and DRX during hot compression and results in remarkable softening, as shown in figures 3(a) and (c). Similarly, to compare figure 11(a) against figure 11(c) (or figure 11(b) against figure 11(d), the higher deformation temperature leads to more remarkable DRV and DRX, which can significantly soften the 2055 Al-Li alloy, as shown in figures 3(a) and (c).

Figures 12(a)–(d) show the bright field TEM micrographs of the compressed samples with different 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 respectively. The number density of the dislocation in the grain increases sharply during the hot deformation with the strain rate of 1 s⁻¹ at 320 °C, and the dislocations interact with each other forming dislocation nets and dislocation pile-up, as shown in figure 12(a). With the strain rate decreasing the number density of the dislocation in the grain decreases and some sub-grain boundaries form with dislocation slipping and climbing, as shown in figure 12(b). In the figure 12(c), the sharply marginated micro-sized DRXed grain close to the dislocation wall near the original grain boundary can be observed, which indicates that the high deformation temperature provide adequate energy for DRV and DRX comparing against figure 12(a). Figure 12(d) shows the high angle grain boundaries form after hot compression with strain rate of 0.001 s⁻¹ at 440 °C, which is consistent with the result from figure 11(d).
Figure 13 shows the bright field TEM images of precipitates in the 2055 Al-Li alloy with T6 state and the SAED (selected area electron diffraction) patterns with electron beam parallel to [100] and [110] axis of Al matrix. In figure 13(b), there are weak diffraction spots at the positions of 1/3{220}Al and 2/3{220}Al besides the strong diffraction spots of Al matrix. According to former reports [9, 44–46], the phase are the common precipitate T1 phase (Al2CuLi) in the Al-Li-Cu alloys, and the precipitate is hexagonal, of symmetry p6/mmm and forms as platelets on the {111} planes of the Al matrix, as indicated by the blue arrows in figure 13(a). In figure 13(c), the weak diffraction spots at the position of 1/2{111} can be observed as indicated by yellow arrows. According to former research [47], the phase should be β' phase (Al3Zr) as indicated by yellow arrows in...
Figure 13. (a) The bright field TEM micrograph of the precipitates of the 2055 Al-Li alloy taken from the compressed sample with strain rate of 0.001 s$^{-1}$ at 440 °C; (b) the SAED pattern with electron beam paralleling to [100]$_{Al}$; (c) the SAED pattern with electron beam paralleling to [110]$_{Al}$.

Figure 14 shows the interaction between the precipitates and dislocations. Figure 14(a) shows a BF TEM micrograph in the grain of the compressed sample with deformation condition of 440 °C/0.001 s$^{-1}$ with electron beam paralleling to [110]$_{Al}$ direction. Lots of dislocations are pinned by the precipitates forming dislocation nets and dislocation walls as indicated by red arrows. It seems to that the precipitates and dislocations around them divide the grain into small zones. Recent report [9] proposed the plate-like T$_1$ phase taking responsible for bulk of precipitation strengthening, while it can be observed that the spherical β' phase also can pin dislocation in figure 14(a). Figures 14(b) and (c) show the enlarged BF TEM image and DF TEM image of the precipitates and dislocations. Figures 14(d) and (e) are the HRTEM images corresponding to the yellow and blue rectangular regions in figure 14(b) respectively. The interface between the T$_1$ precipitate and Al matrix is about 5 nm wide, as shown in figure 14(d). A nano-scale interface resulting from the dislocation in the Al matrix is about 2.2 nm wide as shown in figure 14(e), the corresponding reduced FFT pattern at the top-right corner has weak diffraction spots to suggest existing of lattice imperfection. Figure 14(f) shows the inverse FFT pattern corresponding to red rectangular region in figure 14(e), and the red line and blue line represent the lattice plane {111}$_{Al}$ of Al matrix distributed in the two sides of the nano-scale interface respectively. The angle between the red line and the blue line is 0.8° as shown in figure 14(f), i.e. the misorientation between upper left zone and lower right zone of the Al matrix is 0.8°, which suggest that the regions separated by dislocations and precipitates in the grain might be regarded as sub-grains to some extent. The latest research from Feng et al revealed that the precipitates play more important role than the solute atoms in deformation-induced grain refinement in a Al–Cu–Mg alloy, which is attributed to the strong precipitate-dislocation interaction which can enhance dislocation multiplication, but reduce dislocation mobility and promote precipitate dissolution and solute segregation to grain boundaries [48]. In fact, earlier research from Deschamps et al also revealed that the influence of precipitation on plastic deformation of Al–Cu–Li alloys is negligible, especially the influence of T$_1$ phase, which was also attributed to the strong interaction between dislocation and precipitates [16, 17]. Above all the the precipitate-dislocation interaction plays a considerable role in deformation behavior and deformation-induced grain refinement, or further the mechanical property of the Al alloy, therefore the method of aging before deformation may provide another probability to modify mechanical properties of Al alloy.
4. Conclusions

The hot deformation behavior of as-extruded 2055 Al-Li alloy at T6 state was investigated by uniaxial hot compression test at the temperature of 320°C, 380°C, 440°C, 500°C with strain rate ranging from 0.001 s$^{-1}$ to 1 s$^{-1}$. The microstructures of the compressed samples were analysed by OM and TEM. Several conclusions from this work can be obtained, as shown below.

(1) The flow stress decreased with deformation temperature increasing and strain rate decreasing. Deformation instability occurred under the deformation condition of 380°C/5s$^{-1}$ and at the deformation temperature of 500°C. The constitutive equation based on Arrhenius equation were established with the deformation activation energy of 174.6 kJ mol$^{-1}$.

(2) The DRXed grains were observed in the compressed samples at 440°C with strain rate of 1 s$^{-1}$ and 0.001 s$^{-1}$. DRV and DRX were co-responsible for the dynamic restoration during at high temperature.

(3) The dominated precipitates in the alloy were plate-shaped T$_1$ phase and spherical $\beta_1$ phase. The dislocations could be pinned by both the T$_1$ phase and $\beta_1$ phase, and the precipitates and dislocations around them divided the grain into small zones, and the misorientation across the dislocation was about 0.8°.

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