Serrated flow behavior in 2090 Al–Li alloy influenced by texture and microstructure

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Abstract

The center- and surface-layer specimens of 2090 Al–Li alloy plate was solution-treated, peak-aged, or reversion-treated. The flow stresses of the center layer specimens were higher than those of the surface-layer specimens, regardless of the heat treatments. The textures of the surface-layer specimens were approximated by the \{001\}⟨110⟩ orientation, while those of the center-layer specimens were approximated by the \{011\}⟨211⟩ orientation. Both the surface- and center-layer specimens of the solution-treated alloy gave rise to extensive serrations in their flow curves at a strain rate of 2 × 10⁻⁴ s⁻¹. For the peak-aged alloy with spherical δ′ precipitates, the surface-layer specimen underwent serrated flows, whereas the flow curves of the center-layer specimen were devoid of serration. Serrated flows occurred in both the surface-layer and center-layer specimens of reversion-treated alloy with plate-like T1 precipitates. The findings are discussed in terms of the texture, microstructure, serration characteristics, and kinetics of the onset of serrated flows. The study results indicate that the serrated flow of the Al–Li alloy cannot be explained by a single mechanism, but can be attributed to both the shearing of δ′ or T1 precipitates and dynamic strain aging.

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Keywords: 2090 Al–Li alloy; Serrated flow; Texture; Precipitates; Dynamic strain aging; Strain rate sensitivity

1. Introduction

Explanation of the phenomenon of serrated flow, or the Portevin-Le Chatelier (PLC) effect, in Al–Li alloys is based on two lines of thought: one is dynamic strain aging (DSA) involving dynamic interaction between dissolved lithium solute atoms and mobile dislocations [1–4], and another is shearing of δ′ (Al₃Li) precipitates by mobile dislocations [5–8]. This controversy hints that the mechanism for serrated flow in Al–Li alloys remains ambiguous and further clarification is needed on the matter.

To further clarify the serrated flow mechanism in the case of Al–Li alloys and to better understand this phenomenon, the present work was carried out with 2090 Al–Li alloy and focused on the influence of microstructure and texture on the serrated flow behavior. For this purpose, several different heat-treatments, such as solution treatment, peak-aging for δ′ precipitates, and reversion treatment for T1 precipitates, were employed in this study.

In general, aluminum alloys are known to exhibit considerable anisotropy of mechanical properties. This anisotropy is usually related to the crystallographic texture and in turn active slip systems. Since the PLC effect involves interactions between gliding dislocations and other crystal defects, one should expect that its intensity may also exhibit anisotropy [9]. Such anisotropy has been reported in a commercial Al–Mg alloy by Cheng and Morris [10] and in an Al–Li–Cu–Zr model alloy by Mizera and Kurzydlowski [9]. They accounted for the observed differences in the PLC effect intensity in terms of anisotropy of the microstructure or the morphology of the grains. On the other hand, little investigation has been focused on the textural effect with respect to the PLC behavior.

In previous studies [11,12], it has been suggested that the serrated flow behavior in a peak-aged 2090 Al–Li alloy is strongly affected by the specimen texture when elongated in the rolling direction. From these studies, it was motivated to investigate flow behaviors of solution-treated and reversion-treated 2090 Al–Li alloys, as well as the peak-aged alloy to better understand the texture effect [13]. Some speculations and suggestions were
given in the previous papers, which are desired to be resolved. To do this, solution-treated, peak-aged, and reversion-treated surface- and center-layer specimens of 2090 Al–Li alloy plate were tensile-tested at various temperatures and strain rates. The test results are discussed in terms of texture, microstructure, serration characteristics, and the activation energy for the onset of serrated flow of the specimens.

The study results indicate that the serrated flow of the Al–Li alloy cannot be explained by a single mechanism, but can be attributed to both the shearing of $\delta'$ or $T_1$ precipitates and DSA.

2. Experimental procedures

The material used in this study was a near peak-aged 2090-T81 Al–Li alloy (2.05% Li, 2.86% Cu, 0.12% Zr, balance Al) plate of 12.7 mm in thickness, produced by Alcoa, UK. Tensile specimens of 1 mm $\times$ 6 mm $\times$ 25 mm (ASTM B557M-94) in gauge dimension with the tensile axis along the rolling direction and the width direction along the transverse direction and specimens for texture measurement with a dimension of 10 mm $\times$ 10 mm $\times$ 1 mm were cut from the surface (S) and center (C) of the plate by spark erosion. Specimens for transmission electron microscopy (TEM) were prepared such that their foil normal is along the rolling direction of the plate, and tension-tested specimens for TEM were prepared from their middle regions with the foil normal being along the plate normal. Disc specimens of approximately 36 mg weight and 5 mm diameter for differential scanning calorimetry (DSC) were prepared from heat-treated slices.

The specimens were subjected to three different heat-treatments: solution treatment at 550 °C for 30 min, peak-aging at 190 °C for 18 h following solution treatment, and reversion treatment at 275 °C for 2 min following peak-aging. All the heat treatments were performed in a salt bath. The heat-treated specimens were quenched to cold water at room temperature and kept at $-18^\circ$C to prevent phase transformation before tests.

The tensile specimens were ground and bright-polished prior to the tensile tests. Tensile tests were carried out on an Instron-5582 testing machine at initial strain rates of $2 \times 10^{-5}$ to $2 \times 10^{-2}$ s$^{-1}$ in a thermal chamber regulated at temperatures ranging from 15 to 140 °C. The resolution of the Instron machine was 0.5 MPa under the present testing conditions used, which meant that no serration could be detected when the stress drop in serrated flow was lower than 0.5 MPa. The stress drop magnitude for serration is defined as the engineering stress difference between the maximum and minimum points of an individual serration. The tensile data were processed using a computer program to generate the serration number, the size of stress-drop,

![Fig. 1. (a) Optical microstructure and dark-field TEM micrographs of (b) solution-treated, (c) peak-aged, and (d) reversion-treated 2090 Al–Li alloys. L, LT, and ST stand for longitudinal (rolling), long transverse (transverse), and short transverse (normal) directions of plate.](image-url)
and the stress-drop and elapsed times. Two or three specimens were tensile tested under the same conditions. The different specimens yielded almost the same results.

The specimens for TEM were twin-jet electropolished at 40 V in a 25% HNO$_3$–75% methyl alcohol solution cooled to $-30^\circ$ to $-40^\circ$C. All TEM foils were examined in a Philips CM20 transmission microscope operated at 200 kV.

The DSC specimens were examined using a Perkin-Elmer DSC-7 system scanning from room temperature to 545 $^\circ$C at a heating rate of 10 $^\circ$C/min. The orientation of specimens for the thermal analysis is immaterial.

The texture specimens were measured with an X-ray texture goniometer in the back reflection mode with Ni filtered Cu K$\alpha$ radiation. The (1 1 1), (2 0 0), and (2 2 0) partial pole figures were measured and used to calculate the orientation distribution functions (ODFs) by the WIMV method [14].

3. Results and discussion

3.1. Microstructures

Fig. 1 shows an optical microstructure and dark-field TEM micrographs of the solution-treated, peak-aged, and reversion-treated specimens. Elongated grains can be seen in the optical microstructure, which implies that mechanical fibering generated during rolling has not been completely eliminated even after the solution treatment. A number of very fine particles in the solution-treated specimen (Fig. 1b) are thought to be $\delta'$ (Al$_3$Li) precipitates because $\delta'$ precipitation takes place during quenching from the solution treatment temperature or rapidly at room temperature [15]. The white particles in the TEM micrograph of the peak-aged specimen are $\delta'$ precipitates having L1$_2$ crystal structure (Fig. 1c). It is also known that the $\delta'$ phase is fully coherent and has a very small lattice mismatch with the fcc $\alpha$ matrix [16]. The white streaks in the reversion-treated specimen are plate-like T$_1$ (Al$_2$CuLi) precipitates (Fig. 1d). This reversion treatment indicates that the $\delta'$ precipitates formed during peak-aging have been dissolved out completely, and T$_1$ precipitates became the dominant precipitate phase. The T$_1$ phase forms as thin plates on the {1 1 1} matrix planes with the following orientation relationship [15].

\[
\{0001\}_{T_1}//\{1 1 1\}_\alpha \\
\{1 0 1 0\}_{T_1}//\{1 1 0\}_\alpha
\]

3.2. Textures

Fig. 2 shows the orientation distribution functions of the solution-treated, peak-aged, and reversion-treated surface- and center-layer specimens. Regardless of the heat treatments, the textures of the S and C specimens can be approximated by the \{0 0 1\} \{1 1 1\} and \{0 1 1\} \{2 1 1\} orientations, respectively. From
Table 1
Calculated Schmid factors for tensile axes of [1 1 0] and [1 1 2] of fcc crystal having \{1 1 1\}(1 1 0) slip systems

<table>
<thead>
<tr>
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<td>[1 1 0] axis</td>
<td>0</td>
<td>(\sqrt{6}/6)</td>
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<td>[1 1 2] axis</td>
<td>0</td>
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the fact that the \{0 0 1\}(1 1 0) orientation is the major component in shear deformation texture of aluminum alloys, and that the surface layer of aluminum undergoes shear deformation (e.g., [17]), the surface layers of the Al–Li alloy plate must have been shear deformed during rolling. The shear texture remains little changed even after annealing [18]. On the other hand, the center layers, which undergo plane strain compression, show a texture approximated by the \{0 1 1\}(2 1 1) orientation, which is obtained in many copper alloys with low stacking fault energies [19], Cu–1% P alloy [20], and Cu–Mn alloys with high strain-hardening rates [21]. These results and the mechanical fibering in Fig. 1a indicate that the deformation texture of the starting plate remained essentially unchanged even after solution treatment, peak-aging, and reversion treatment. In other words, the texture of matrix was not influenced by the heat treatments.

3.3. Flow behavior

Fig. 3 shows the tensile stress–strain curves of the solution-treated, peak-aged, and reversion-treated surface- and center-layer specimens. Severe serrated flows occurred in both the S and C specimens of solution-treated alloy (curves 5 and 6). For the peak-aged alloy, the S specimen undergoes a serrated flow whereas the C specimen does not show serration (curves 1 and 2). For the reversion-treated alloy, both the S and C specimens show serration, with the serration is more extensive in the S specimen than in the C specimen (curves 3 and 4). Those results show that serration in the tensile flow of the present Al–Li alloy is affected by not only its microstructure but also its texture.

3.4. Flow stresses

Fig. 3 shows that the C specimen has about 1.3–1.4 times higher flow stress than the S specimen, even though they were subjected to the same heat treatments and so have the same microstructures. It also shows that the C specimen has higher strain-hardening rate than the S specimen. This behavior has been discussed elsewhere [11,12]. The number of active slip systems in the C specimen is likely to be larger than that of the S specimen (Table 1). As the number of active slip systems increases, the interaction between dislocations is likely to increase and the dislocation density increases with increasing strain, that is, the strain hardening rate increases, if other hardening conditions are the same. Therefore, the C specimen can give rise to higher strain-hardening rates than the S specimen at a given strain rate.

Fig. 4. For primary (1st) slip system having \(\sqrt{6}/6\) in Schmid factor being activated, dislocations pile up at spherical \(\delta^*\) particle. When stress on dislocation at head (A) of dislocation pile-up reaches shear yield stress of precipitates, precipitates are sheared, resulting in stress drop (C-type). \(\perp\) does not necessarily mean edge dislocation.
3.5. Texture effect on serrated flow in peak-aged specimen

The peak-aged S specimen shows serration (curve 2 in Fig. 3), whereas flow curve of the C specimen is devoid of serration (curve 1 in Fig. 3). When the (0 0 1)[1 1 0] oriented S specimen is elongated, it can be supposed that the (1 1 1)[1 0 1], (1 1 1)[0 1 1], (1 1 1)[0 1 1], and (1 1 1)[1 1 0] slip systems are equally activated because the Schmid factors on the systems are the same (Table 1), and dislocations pile up on the slip planes at $\delta'$ spherical precipitates. When the stress on the dislocation at the head of the dislocation pile-up reaches the shear yield stress of the precipitates, the precipitates are sheared, resulting in the stress drop.

However, tensile straining of the (0 1 1)[2 1 1] oriented C specimen activates the (1 1 1)[1 0 1], (1 1 1)[0 1 1], (1 1 1)[1 1 0], (1 1 1)[0 1 1], and (1 1 1)[1 1 0] slip systems (Table 1), neglecting the slip systems for which the Schmid factor is $\sqrt{6}/18$. Thus, the Schmid factors for the active slip systems are different unlike the (0 0 1)[1 1 0] oriented S specimen. The slip systems for which the Schmid factor is $\sqrt{6}/9$ are first activated. If the slip systems for which the Schmid factor is $\sqrt{6}/9$ are activated by cross slip before the stress on the dislocation pile-up on the first slip systems reaches the shear yield stress of the $\delta'$ precipitates, the particles are not likely to be sheared (Fig. 4). In this case, no serration will take place. This may be the case of the flow curve of the C specimen, which is devoid of serration. This does not negate the possibility of shearing of the precipitates. If a sufficient number of precipitates are not sheared, the serration cannot be detected. Fig. 5 shows the TEM micrographs in tensioned peak-aged samples.
The smaller elongation of the peak-aged C specimen implies that dislocations on the slip systems with the smaller Schmid factors are difficult to shear δ' precipitates without cracking because the precipitates associated with the smaller Schmid factors are subjected to higher tensile stresses than those with the larger Schmid factors.

3.6. Texture effect on serrated flow in reversion-treated specimen

For the reversion-treated specimen, serrated flows are observed even in the C specimen (curve 3 in Fig. 3) as well as in the S specimen (curve 4 in Fig. 3). Since the T1 precipitates form as thin plates on the {1 1 1} matrix planes, gliding dislocations are likely to meet the T1 precipitates on four different {1 1 1} planes, that is, (1 1 1), (1 1 1), (1 1 1), and (1 1 1) as schematically shown in Fig. 6. A gliding dislocation bypasses the precipitates lying on its glide plane by multiple cross slip or by leaving loops around them because of their thin plate-like characteristics, while some parts of it are likely to meet the precipitates lying on at least one of the remaining three different {1 1 1} planes. These precipitates are likely to be sheared, because of their thin cross-sections as shown in Fig. 5e. The distribution of the {1 1 1} planes is not sensitive to the orientation of specimen. Therefore, unlike the peak-aged specimens, the serration is observed even in the reversion-treated center-layer specimens.

![Fig. 6](image)

Fig. 6. (a) Corners A, B, C, D of tetrahedron and their opposite faces α, β, γ, δ. (b) α-plane is on glide plane (dark plane), while other planes (β, γ, δ) are at 70.5° to glide plane.

![Fig. 7](image)

Fig. 7. Stress–strain curves of solution-treated center-layer specimens of 2090 Al–Li alloy plate at various (a) strain rates and (b) temperatures. (c) Plot of ln(δ) vs. lnεc. (d) Plot of lnεc vs. 1/T.
3.7. Activation energy for serrated flow

The treatment of dynamic strain aging for serrated flow in substitutional fcc metals, which incorporate the deformation-produced vacancies, solute concentration, and mobile dislocations generated during plastic deformation, can be described by the following equation [22].

\[ \dot{\varepsilon} = K C^n \exp \left( -\frac{Q_M}{RT} \right) \varepsilon_c^m \]  

(1)

where \( \dot{\varepsilon} \) is the plastic strain rate, \( C \) the solute concentration (wt.%), \( Q_M \) the effective activation energy for the thermally activated process responsible for serrated flow, \( \varepsilon_c \) the critical strain for the onset of serration, \( R \) the gas constant, and \( K, n, \) and \( m \) are the empirical constants.

Taking the logarithm of both sides of Eq. (1), we have

\[ \ln \dot{\varepsilon} = \ln K + n \ln C - \frac{Q_M}{RT} + m \ln \varepsilon_c \]  

(2)

Fig. 7a shows the strain rate effect on the onset-strain of serrated flow of the solution-treated specimens. Fig. 7c gives the plot of \( \ln \dot{\varepsilon} \) versus \( \ln \varepsilon_c \) at 25°C. From the slope of the plot, \( m = 1.9716 \) is obtained. Fig. 7b shows the effect of temperature on the onset-strain of serrated flow of solution-treated specimens at a strain rate of \( 2 \times 10^{-3} \) s\(^{-1}\). From the plot of \( \ln \varepsilon_c \) versus \( 1/T \) in Fig. 7d, it is obtained that \( Q_M/mR = 1.4590 \). Thus, the effective activation energy for the thermally activated process responsible for serrated flow is calculated to be \( Q_M = 23.92 \) kJ/mol (0.2478 eV).

In the substitutional fcc alloy system, \( Q_M \) is believed to be the activation energy for vacancy motion [22]. The energy required to move a vacancy in pure aluminum is 0.75 eV [4,23] and the binding energy exists between a vacancy and a lithium atom is 0.25 eV [24]. Therefore, the activation energy for the movement of a lithium atom in aluminum by bulk lattice diffusion should be 0.5 eV [23]. The activation value measured in this study, 0.2478 eV, is one half of the activation energy for the movement of lithium in aluminum, 0.5 eV.

The lower activation energy for solute diffusion can be attributed to pipe diffusion along the dislocation cores. Since the diffusivity of solute atoms along dislocations by a vacancy mechanism can be significantly higher than that in the matrix due to the high concentration of vacancies in the dislocation cores. It has been concluded that the activation energy in the region of dislocation core is in the range of 40–70% of the activation energy for lattice diffusion [25]. Therefore, the most probable mechanism for the initiation of serrated flow in present solution-treated 2090 Al–Li alloy specimens is DSA involving dislocation pipe diffusion as the rate-controlling process, as it is suggested that the serration is attributed to the DSA caused by pipe diffusion of Li atoms for the serrated flow behavior in Al–Li alloy [2].

Eq. (2) predicts that the plot of \( \ln \varepsilon_c \) versus \( 1/T \) should have a positive slope because \( Q_M/mR \) has a positive value. Therefore, the determination of the apparent activation energy for serrated flow is limited to the temperature range in which a positive slope of the \( \ln \varepsilon_c - 1/T \) plot is obtained. However, tensile tests of the peak-aged and reversion-treated specimens give rise to negative slopes, that is, \( \varepsilon_c \) increases with increasing temperature as shown in Figs. 8c and 9c [26].

For aluminum and other substitutional alloys, if a negative slope in the \( \ln \varepsilon_c \) versus \( 1/T \) relation occur, the following relation may be applied [27]:

\[ \varepsilon_c^{\beta} \dot{\varepsilon}^\gamma = K \exp \left( -\frac{Q'}{RT} \right) \]  

(3)
where $\alpha$, $\beta$, $\gamma$, and $K$ are constants, and $Q' = E_I + E_m$ is the activation energy for formation and movement of vacancies. Taking the logarithm of both sides of Eq. (3), the following equation is obtained

$$\ln \varepsilon_c = \frac{\ln K - \gamma \ln \dot{\varepsilon}}{\alpha \beta R} - \frac{Q'}{\alpha \beta RT}$$

(4)

From the plots of $\ln \varepsilon_c$ versus $1/T$ in Figs. 8c and 9c, we obtain $Q'/\alpha \beta R = 0.6226$ and 3.6855, respectively. Because reported values of $\alpha$ and $\beta$ lie between 0.6 and 1.2 for Al alloy, taking $\alpha \beta = 1$ [27], $Q'$ is calculated to be 5.18 kJ/mol (0.054 eV) from Fig. 8c and 30.6 kJ/mol (0.32 eV) from Fig. 9c.

Fig. 9. Stress–strain curves of reversion-treated surface-layer specimens of 2090 Al–Li alloy plate at various (a) strain rates and (b) temperature [26]. (c) Plot of $\ln \varepsilon_c$ vs. $1/T$ [26].

Charnock [28] obtained the value of $Q'$ ranging from 58 to 116 kJ/mol (0.6 and 1.2 eV) for aluminum and other substitutional alloys. King et al. [27] measured $Q'$ for vacancies in as-quenched Al–Zn–Mg alloy as 19.7 kJ/mol, and suggested that the lower $Q'$ value might be due to the quenched-in vacancies present at the start of the tests that made the $E_I$ term unimportant. Therefore, $Q' \approx E_m$, which is almost the same as the $E_m$ value for vacancies in an Al–7 wt% Mg alloy, 19.3 kJ/mol, measured by Harris [29]. Furthermore, they suggested that the serrated yielding in Al–Zn–Mg alloys was attributed to the DSA caused by pipe diffusion of solute atoms.

Fig. 10 shows the DSC analysis results of heat-treated 2090 Al–Li alloy samples. Labeled A, B, C, D, and E effects are attributed mainly to the dissolution of solute clusters, $\delta'$ precipitation, $\delta'$ dissolution, $T_1$ precipitation, and $T_1$ dissolution, respectively [15,30–33]. F effect in the reversion-treated sample may be due to the formation of solute clusters. The quenched-in supersaturated vacancies could be annihilated during the reversion-treatment at 275°C, and the decrease in excess vacancies retards the kinetics of clustering, resulting in a formation of solute clusters at a temperature range of 80–120°C rather than during quenching or at room temperature as suggested for Al–Li alloys [30,31]. Thus, the $E_I$ term cannot be neglected in the case of reversion-treatment.

Since the activation energy for formation of vacancies, $E_I$, in Al–Cu alloys is about 0.67 ± 0.06 eV [34], the activation energy for movement of vacancies in Al–Li alloy $E_m = 0.5$ eV [23], the value of $Q'$ is estimated to be about 1.17 eV, which is apparently larger than $Q' = 0.32$ eV obtained for reversion-treated specimen. This indicates that the starting of serrated flows in the reversion-treated specimen cannot be attributed to the DSA, but seem to result from the shearing of precipitates.

For peak-aged 2090 Al–Li alloy specimens, since the value of $Q' = 5.18$ kJ/mol (0.054 eV) cannot be comparable with the activation energy for serrated flow of Al–Li alloy caused by the DSA involving bulk diffusion or pipe diffusion of solute atoms. Therefore, the starting of serrated flow is thought to be attributed to shearing of $\delta'$ precipitates.
3.8. Stress drops and elapsed time in serrated flow

Fig. 11 shows the stress drop versus tensile strain relations for the solution-treated, peak-aged, and reversion-treated specimens. Fig. 12 shows the elapsed time (an interval of time between serrations) versus strain relations of the solution-treated, peak-aged, and reversion-treated specimens. For the solution-treated specimens, the stress drop and elapsed time tend to increase with increasing tensile strain. However, the peak-aged and reversion-treated specimens show little correlation between the stress drop and elapsed time and the tensile strain.

Since the stress drop versus strain relations and the elapsed time versus strain relations behave similarly, they can expect some relations between the stress drop and the elapsed time as shown in Fig. 13. The stress drop tends to increase with elapsed time in the solution-treated specimens, and no correlation can be seen in the peak-aged and reversion-treated specimens.

If serrated flow is dominated by DSA, on the one hand, the diffusion rate of solute atoms increases with strain and consequently the moving dislocations are more strongly locked at larger strain [35]. The interaction between the solute atoms and moving dislocations increases with increasing the total strain, which will result in an increase of serration amplitude [36]. On the other hand, since serrated flow is diffusion-controlled and it is time and temperature dependent [37], at the higher strain there is greater difficulty in unlocking. As a result, the time for the lock-and-unlock process between moving dislocations and their atmospheres is longer, and the elapsed time of serrations increases. A study on the serrated flow in Al-Mg alloy by Pink and Grinberg [38] indicated that the stress drop (\(\Delta \sigma\)) is proportional to the time interval (\(\Delta t\)) for every individual event of the solute diffusion to the dislocations at constant strain, strain rate, and temperature, i.e., \(\Delta \sigma = \Delta \sigma 0.31\text{–}1.5\). Behnood and Evans [39] attributed the serrated flow in solution-treated Al-Li specimens to the interaction between dislocations and precipitates, which resulted in the shear of \(\delta\) precipitates, because quenching does not suppress precipitation of \(\delta\).

From the relations between the stress drop and strain and the relations between the elapsed time and strain in the different specimens, as well as the phenomena explained in the previous sections, the serrated flow in the solution-treated specimens can be attributed to the DSA due to the interaction between moving dislocations and solute. However, because many fine \(\delta\) particles exist in the present solution-treated specimens, the serrated flow behavior must be partially attributed to the \(\delta\) particles shearing by moving dislocations. This is supported by the large scatterings in the stress drop and elapsed time data. Also, Tian et al. [40] thought that the irregular variations in stress drop and elapsed
Fig. 12. Relations between elapsed time for serration and strain for solution-treated surface-layer (S) and center-layer (C) specimens and peak-aged and reversion-treated surface-layer specimens of 2090 Al–Li alloy plate at initial strain rate of $2 \times 10^{-4}$ s$^{-1}$ at 25 °C.

Fig. 13. Elapsed time dependences of stress-drop for individual serrations obtained in tensile test of solution-treated surface-layer (S) and center-layer (C) specimens and peak-aged and reversion-treated surface-layer (S) specimens of 2090 Al–Li alloy plate at initial strain rate of $2 \times 10^{-4}$ s$^{-1}$ at 25 °C.
time during deformation of Al–Li single crystal may be due to the interaction between dislocations and δ′ particles.

In addition, the irregular changes in stress drop and elapsed time with strain as well as in the relation between stress drop and elapsed time, as shown in Figs. 11c and d, 12c and d, and 13c and d, indicate that the main mechanism of serrated flow in peak-aged and reversion-treated specimens is the shearing of δ′ or T1 precipitates.

3.9. Strain rate sensitivity

For similar degrees of serrations in the stress–strain curves as can be seen in comparison of curves (1) and (3) of Fig. 7a, the flow stresses increase with increasing strain rate, indicating positive strain rate sensitivity as usual. On the other hand, when the degree of serration decreases with increasing strain rate, the flow stress tends to decrease with increasing strain rate, indicating negative strain rate sensitivity, as can be seen in comparison of curves (2) and (4) of Fig. 7a and curves (1) and (2) of Figs. 8a and 9a.

The negative strain rate sensitivity is often attributed to DSA. However, the present result indicates that it is not correct. Since flow serrations are caused by interaction between moving dislocations and solute atoms and/or precipitate particles, the degree of serration may be proportional to the degree of pinning of dislocations, which in turn increases the flow stress.

For serrated flows the strain rate sensitivity is supposed to be determined by two competing processes: one is the usual strain rate sensitivity having a positive value and the other is a variation of serration with strain rate. When the degree of serration decreases with increasing strain rate, the strain rate sensitivity is likely to be negative, if the usual positive strain rate sensitivity is not dominant.

4. Conclusions

The serrated flow behavior of the surface and center layers of 2090 Al–Li alloy plate has been studied by tension at various temperatures and strain rates in combination with textures and microstructures. The study has lead to the following conclusions.

The texture strongly affects flow in the peak-aged specimens. Serrated flow occurs in the surface layer having the {0 0 1}{1 1 0} texture, whereas no serration is observed in the center layer with the {1 1 1}{2 1 1} texture because the probability of shearing of δ′ precipitates is lower in the center layer than in the surface layer due to easier cross slip.

The solution-treated specimens show severe serrated flows due to the dynamic strain aging involving dislocation pipe diffusion of solute Li atoms towards moving dislocations, while the serrated flows are partially attributed to the shearing of fine δ′ precipitates existing in the specimens.

The calculated activation energy for the onset of serrated flow and serration characteristics imply that the occurrence of serrated flow in the peak-aged and reversion-treated surface-layer specimens cannot be ascribed mainly to dynamic strain aging, but can be attributed to the shearing of δ′ and T1 precipitates.

For serrated flows the strain rate sensitivity is supposed to be determined by two competing processes: one is the usual strain rate sensitivity having a positive value and the other is a variation of serration with strain rate. When the degree of serration decreases with increasing strain rate, the strain rate sensitivity is likely to be negative, if the usual positive strain rate sensitivity is not dominant.

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