Temperature-Dependent Deformation Behavior of Al–Mg–Sc Alloys Fabricated by Multi-Directional Forging at Room Temperature

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Ultrafine-grained Al–Mg–Sc alloys were fabricated by multi-directional forging (MDF) with different number of forging passes of 3, 9 and 15, i.e., to cumulative strains of ∑Δε = 1.2, 3.6 and 6.0, at room temperature. The achieved average grain sizes were 950, 680 and 360 nm at 3, 9 and 15 passes, respectively. Peak-aging treatments at 473 K for 172.8 ks were adopted for a portion of specimens after MDF in order to obtain finely dispersed Al3Sc precipitates. Grain coarsening did not take place in all the specimens during the aging. The activation volume for plastic deformation was estimated from the strain-rate jump tensile tests before and after the aging. Aging-free 3- and 9-pass specimens showed positive temperature dependence of the activation volume, while that of the aging-free 15-pass one bearing the smallest grain size exhibited a negative temperature dependence. Contrary to these results, values of the activation volume in the peak-aged specimens were approximately identical regardless of grain size or deformation temperature. These results strongly suggested that, due to the precipitation of Al3Sc, the rate-controlling process of deformation was changed from interaction between forest dislocations and mobile dislocations for the aging-free 3- and 9-pass specimens to interaction between mobile dislocations and Al3Sc precipitates, or from bowing-out of dislocations from grain boundaries for the aging-free 15-pass specimen to the interaction between the mobile dislocations and precipitates.

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1. Introduction

Bulkly ultrafine-grained (UFGed) metals and alloys with nanometer-scale grain sizes have been recently fabricated via severe plastic deformation (SPD) methods, which include accumulative roll-bonding (ARB),¹ equal-channel angular pressing (ECAP),² high-pressure torsion (HPT)³ and multi-directional forging (MDF)⁴. UFGed metals and alloys fabricated by means of SPD methods have been actively investigated in the last two decades⁵⁻⁷ and many researchers have reported the excellent mechanical properties. In this way, the unique mechanical properties of UFGed materials are being come to light, which are largely different from those of conventional coarse-grained ones. Thus, UFGed materials are highly attractive from both the practical and academic perspectives; for example, the activation volume of plastic deformation has a positive temperature dependence in coarse-grained pure face-centered cubic (fcc) metals, but it has exhibited a negative temperature dependence in submicron-grained materials.⁸⁻¹¹ Kato et al.¹⁰,¹¹ quantitatively explained the grain-size and temperature dependencies of the activation volume of the UFGed pure fcc metals with a depinning model considering the dislocation bowing-out from grain boundaries. This model is based on the decreasing in-grain dislocation sources with decreasing grain size. According to the model, the transmission of dislocations transitions from in-grain sources to grain boundaries, and then, the temperature dependence of activation volume changes. However, the effect of second-phase particles dispersed in UFGs on the temperature dependence of activation volume has not been clarified yet. When precipitates are dispersed in the UFGs, it can be assumed that the rate-controlling process of the plastic deformation would no longer be the dislocation bowing-out from the boundaries but the process of overcoming the precipitates. Therefore, dispersoids in UFGs would also change the temperature dependence of activation volume.

In the present study, UFGed Al–Mg–Sc alloys with having different grain sizes were fabricated by means of MDF. Some specimens were followed by aging to have finely dispersed Al3Sc precipitates in grain interiors and on boundaries. The differences in the temperature dependence of mechanical properties of these specimens were systematically investigated.

2. Experimental Procedure

A hot-rolled Al–Mg–Sc alloy plates, provided by UACJ Corporation, were used in this study, and the chemical composition is listed in Table 1. The plates were fabricated by the following procedure. Ingots with dimensions of

Table 1  Chemical composition of an Al-Mg-Sc alloy used in this study (mass%).

| Si  | Fe  | Mg  | Sc  | Ga  | V  | Al |
|-----|-----|-----|-----|-----|----|----|
| 0.041 | 0.136  | 2.93 | 0.210 | 0.010 | 0.015 | Bal |

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170 mm length, 200 mm width, and 60 mm thickness were prepared by book mold casting and homogenized at 813 K for 14.4 ks. Then, their surfaces were milled down by 5 mm and hot rolled at a starting temperature of 788 K to a thickness of approximately 20 mm. The hot-rolled plate was cut into rectangular-shaped pieces followed by solid-solution treatment at 863 K for 7.2 ks and MDF at room temperature (RT). In the MDF processing, constant forging-pass strains are repeatedly applied from three directions (the axes perpendicular to the three surfaces of the rectangular-shaped pieces). During MDF, the dimensions of the pieces are unchanged by employing an appropriate ratio of the specimen dimensions and the fitting forging pass strain, $\Delta \varepsilon$. Thus, the forging can be repeated by rotating its direction by 90°, and unlimited strain can be theoretically given to the specimen. In this study, $\Delta \varepsilon$ was set as 0.4 in true strain and dimensions of the rectangular-shaped pieces as 15.0, 22.4 and 18.3 mm which axes were parallel to the rolling direction (RD), transverse direction (TD) and normal direction (ND), respectively. MDF processing was conducted on an Amsler-type universal mechanical testing machine with an initial strain rate $\dot{\varepsilon}$ of $1.0 \times 10^{-3}$ s$^{-1}$ up to forging passes of 3, 9, and 15, which correspond to cumulative strains of $\Sigma \Delta \varepsilon = 1.2$, 3.6 and 6.0, respectively. Some specimens were aged at 473 K for various periods of time up to 302.4 ks at maximum. Hereafter, the specimens are labeled using characters of ST (solution-treated), PA (peak-aged) and X (the number of MDF passes); for example, “the ST-9” indicates the specimen solution treated and followed by 9 passes of MDF, and “the PA-9” peak aged of the ST-9. The specimens used in all the tests were cut out from the center of the rectangular-shaped pieces.

The microstructure of the specimens was investigated using an optical microscopy (OM) and a transmission electron microscopy (TEM). For the OM observation, the specimens were mechanically polished using SiC papers first, an alumina suspension with an abrasive size of 0.05 µm next, and finally a colloidal silica suspension to have mirror-like surface. They were finished by etching with a hydrofluoric acid solution (distilled water:hydrofluoric acid (46%) = 10:1 in volume) at RT for 30 s. The TEM observations were conducted using an FEI TECNAI G2 microscopy under an accelerating voltage of 200 kV. Thin foils for TEM observations were prepared by twin-jet electro-polishing with a perchloric acid/methanol solution (1:9 in volume) at 243 K and 28 V.

Measurements of hardness and electric resistance, and tensile tests were also performed. The hardness was evaluated with a micro-Vickers hardness tester (Akashi, HM-102). The measurement was repeated 10 times for each specimen under a loading of 4.9 N for 10 s, and then the average values were taken. The electrical resistance was assessed through the four-terminal method by using a resistance meter (Hioki Electric, RM3545-01) with a current of 1 A at RT. The measured values were converted into the specific resistance based on the cross-sectional area of the specimens and the distance between the terminals. For the tensile tests, dog-bone-shaped specimens with a gage section of 5 mm length, 1.5 mm width and 0.6 mm thickness were employed. They were cut from the rectangular-shaped pieces to have the tensile axis perpendicular to the final forging axis of MDF.

The tests were conducted on an Instron-type universal mechanical testing machine (Shimadzu AG-10kNX Plus) at both 77 K and RT at $\dot{\varepsilon} = 1.0 \times 10^{-3}$ s$^{-1}$. Furthermore, strain-rate jump tests were performed at RT, 200 K and 77 K by repeatedly changing the strain rate between $1.0 \times 10^{-4}$ and $1.0 \times 10^{-3}$ s$^{-1}$. The refrigerants used for the tests at 200 K and 77 K were methanol chilled to near its melting point and liquid nitrogen, respectively. In this case, the specimens were preliminarily soaked in each refrigerant for 600 s and kept soaking during the measurements. For the test at 200 K, the refrigerant temperature was often measured, and further liquid nitrogen was appropriately added to maintain the desired temperature (approximately ±2 K).

3. Results

3.1 Microstructure and mechanical properties

Figure 1 shows the optical micrographs of a hot-rolled and solutionized Al–Mg–Sc specimen with grains elongated along RD. The average grain-boundary spacings perpendicular to ND, RD and TD were about 30, 153 and 54 µm before the solution treatment and approximately 96, 225 and 102 µm afterward, respectively. This indicates grain growth during the solution treatment.

The TEM observation of the hot-rolled specimen revealed the presence of $\text{Al}_3\text{Sc}$ precipitates with an average radius of around 40 nm, while no precipitate was confirmed after solution treatment, which became single-phase fcc. Figure 2 displays a bright-field TEM micrograph of the ST-15 specimen, indicating that the initial coarse grains were dramatically fragmented after MDFing. The corresponding
selected-area-diffraction pattern (SADP) (inset of Fig. 2), taken using a selected-area aperture of 6 µm in diameter, revealed nearly continuous diffraction rings which suggests random-orientation distribution of UFGs surrounded by high-angle grain boundaries. The grain size decreased with increasing the number of MDF passes. The average (sub)grain diameter \( d \) of the ST-3, ST-9 and ST-15 specimens was approximately 950, 680 and 360 nm, respectively. The grain size was measured regardless of the grain-boundary character, i.e., whether they were low-angle or high-angle ones.

Figure 3 presents the age-hardening curves of specimens aged at 473 K. The hardness before aging increased with the number of MDF passes. The average (sub)grain diameter \( d \) of the ST-3, ST-9 and ST-15 specimens was approximately 950, 680 and 360 nm, respectively. The grain size was measured regardless of the grain-boundary character, i.e., whether they were low-angle or high-angle ones.

Figure 4(a) displays a bright-field TEM micrograph and the SADP of the PA-15 specimen, showing that the UFGed structure is still maintained after the peak aging. The average value of \( d \) was about 360 nm, which is almost the same with that of the ST-15 specimen. This indicates that grain coarsening hardly occurred during aging at 473 K. Detailed TEM observations revealed that the precipitates often formed on the boundaries and pinned the grain-boundary migration (Fig. 4(b)). These precipitates were identified as the Al\(_3\)Sc phase based on the SADP analyses. The Al\(_3\)Sc precipitates, therefore, effectively stabilize the UFGed structure during aging. The fine spherical Al\(_3\)Sc precipitates were dispersed also in grain interior as exemplified in Fig. 4(c). Therefore, the hardness increase by aging can be attributed to the formation of these Al\(_3\)Sc precipitates. Table 2 summarizes average grain size of \( d \) and average radius \( r \) of the in-grain precipitates for both the ST and PA specimens. The value of \( d \) appears identical among the PA and the corresponding ST specimens regardless of the MDF passes. Moreover, the value of \( r \) among the three PA specimens looks almost the same.

Figure 5 displays the stress-strain curves attained by tensile tests. All the curves exhibited serrations as shown in
the insets in Fig. 5. Solute Mg atoms in an Al matrix form Cottrell atmosphere around the dislocations.\(^{12-14}\) The occurrence of serrations could be, therefore, due to the impediment and separation of mobile dislocations by and from the Cottrell atmosphere.\(^{14-17}\) Table 3 summarizes the mechanical properties, and Fig. 6 illustrates the relationship between \(d\) and the 0.2% proof stress \(\sigma_{0.2}\) of the specimens. All the specimens exhibited a linear relationship between \(\sigma_{0.2}\) and \(d^{-1/2}\). Thus, it is evident that the increase in \(\sigma_{0.2}\) by grain refinement follows the Hall-Petch relation. Moreover, due to the strengthening by the in-grain Al\(_3\)Sc precipitates, the PA specimens showed higher \(\sigma_{0.2}\) values compared to the ST ones.

Figure 7 shows the results of tensile tests at 77 K. For all the specimens, the flow stress levels and elongations increased compared to those observed at RT. Besides, no serration appeared. Since the diffusion rate of solute Mg atoms exponentially decreases along with decreasing temperature, the Cottrell atmosphere hardly formed at cryogenic temperatures.\(^{18}\)

**3.2 Strain-rate jump tests**

For the strain-rate jump tests, \(\dot{\varepsilon}\) was first set to \(1.0 \times 10^{-3}\) s\(^{-1}\) (\(\dot{\varepsilon}_1\)), and then the crosshead speed was suddenly changed up to \(1.0 \times 10^{-2}\) s\(^{-1}\) (\(\dot{\varepsilon}_2\)) during the tests. After straining to about 0.5%, the crosshead speed was suddenly returned down to \(1.0 \times 10^{-3}\) s\(^{-1}\) and then another straining of 0.5% was added again. The cycle of strain-rate change was repeated until the flow stress reached the tensile strength. When the strain rate is increased, the flow stress generally increases as well.\(^{19}\) Actually, the increase in flow stress at 200 and 77 K corresponded to the increase in strain rate also in the present tests. Nevertheless, at RT, the flow stress decreased as the strain rate increased and vice versa. Miyajima et al. have reported that the deformation of a UFGe\(_{3}\) Al–Mg alloy at RT is mainly controlled by the dynamic strain aging effect from the solute Mg atoms\(^{20}\) and the flow stress decreases with increasing strain rate. Therefore, this reversed behavior of flow stress observed at RT can be ascribed to the dynamic strain aging by the Mg atoms. Thus, only the results of the tests at 77 and 200 K are described from now on.

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**Table 2** Average grain size \(d\) and radius \(r\) of Al\(_3\)Sc precipitates in PA-specimens. Average grain sizes of ST specimens are shown for comparison. Also shown are data for an OA-15 specimen (15-pass specimen over-aged at 473 K for 302.4 ks).

| Specimen | \(d\) [nm] | \(r\) [nm] |
|----------|-----------|-----------|
| PA 3-pass | 950 ± 20  | 3.5 ± 0.4 |
| 9-pass   | 680 ± 20  | 4.0 ± 0.2 |
| 15-pass  | 360 ± 10  | 3.8 ± 0.5 |
| OA 15-pass| 370 ± 10  | 6.8 ± 0.6 |
| ST 3-pass | 940 ± 20  | –         |
| 9-pass   | 670 ± 20  | –         |
| 15-pass  | 360 ± 20  | –         |

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**Table 3** Fracture strain \(\varepsilon_f\), 0.2% proof stress \(\sigma_{0.2}\), and tensile stress \(\sigma_{UTS}\) of ST- and PA-specimens. The data were obtained from tensile tests at an initial strain rate of \(1.0 \times 10^{-3}\) s\(^{-1}\) at room temperature.

| Specimen | \(\varepsilon_f\) [%] | \(\sigma_{0.2}\) [MPa] | \(\sigma_{UTS}\) [MPa] |
|----------|---------------------|----------------------|---------------------|
| 3-pass   | 3                   | 273                  | 330                  |
| ST 9-pass| 5                   | 291                  | 358                  |
| 15-pass  | 3                   | 332                  | 383                  |
| PA 9-pass| 8                   | 318                  | 353                  |
| 15-pass  | 5                   | 371                  | 412                  |

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Fig. 5 Stress-strain curves of (a) ST-specimens and (b) PA-specimens. Tensile tests were conducted under an initial strain rate of \(1.0 \times 10^{-3}\) s\(^{-1}\) at room temperature.
The results of strain-rate jump tests for the ST and PA specimens are shown in Fig. 8 and Fig. 9, respectively. When the strain rate was changed, the flow stress variation was relatively more significant at 77 K than at 200 K. The curves at 77 K exhibited a sawtooth-like shape. In contrast, at 200 K, a sharp increase in flow stress was then followed by gradual decrease, as described in the circle in Fig. 8. This behavior is typically observed when dislocations detach from the Cottrell atmosphere. Therefore, the characteristic change in the flow stress occurred at 200 K can be attributed to the Cottrell atmosphere of the Mg atoms.

The activation volume \( V^* \) was calculated using the results of the first strain-rate jump in Figs. 8 and 9, immediately after the macroscopic yielding (at a plastic strain of approximately 0.5%), as follows: \[ V^* = M_f k T (\Delta \ln \dot{\varepsilon})_T = M_f k T [\ln (\dot{\varepsilon}_2/\dot{\varepsilon}_1)] \] (1)

where \( \Delta \sigma \) is the flow stress change caused by the change in strain-rate from \( \dot{\varepsilon}_1 \) to \( \dot{\varepsilon}_2 \), \( k \) is the Boltzmann constant, \( T \) is the deformation temperature, and \( M_f \) is the Taylor factor, whose general value for fcc metals is 3.06. Figure 10 summarizes the temperature dependence of \( V^* \) for each specimen, derived from eq. (1). Here the \( V^* \) was normalized using the magnitude of the Burger’s vector \( (b = 0.287 \text{ nm}) \), whose value was obtained from the lattice constant \( (a = 0.4063 \text{ nm}) \) experimentally determined via X-ray diffractometry. As shown in Fig. 10(a), the \( V^*/b^3 \) of the ST-3 and ST-9 specimens exhibited positive dependence on the temperature, while that of the ST-15 one demonstrates a negative temperature dependence. On the other hand, the \( V^* \) values...
for the PA specimens were all about $300b^3$ regardless of the number of MDF passes and temperature (Fig. 10(b)).

4. Discussion

The $V^*$ of pure fcc metals strongly depends on the grain size, and its temperature dependency varies when the grains are fragmented to submicron order ($50$–$500$ nm).4–8) More concretely, the temperature dependence of $V^*$ changes from positive to negative with grain refinement.8–11) With a depinning model of dislocations bowing-out from the grain boundaries, Kato et al. quantitatively explained the grain-size and temperature dependencies of $V^*$ in UFG fcc metals.10,11) According to this model, the interactions between mobile and forest dislocations control the deformation rate when the dislocation sources are within the grains, but the dislocation bowing-out from grain boundaries becomes the rate-controlling process when the grain refinement strongly reduces such in-grain sources. As shown in Fig. 10(a), for the ST specimens, the temperature dependence of $V^*$ changed from positive to negative with increasing the number of MDF passes, i.e., with decreasing grain size. This is in good accordance with the model proposed by Kato et al. That is, with decreasing grain size of the ST specimens, dominant dislocation source transitions from grain interior to grain boundary and, therefore, the deformation-rate-control process changes from the interaction between mobile and forest dislocations within grains to bowing-out of dislocations from grain boundaries.

In contrast, the $V^*$ of the PA specimens, which had fine Al$_3$Sc precipitates within the grains, exhibited almost no temperature dependence (Fig. 10(b)). Thus, the presence of precipitates should also cause a change in the rate-controlling process of deformation.23) This is because the dislocations are interacted with and pinned by the in-grain precipitates irrespective of dislocation sources and, therefore, the dislocation overcoming of precipitates becomes the deformation-rate-controlling process. In this case, the $V^*$ should depend on the precipitate radius $r$ and the interprecipitate spacing $\lambda$. Here, the discussion will be proceeded assuming that the Al$_3$Sc precipitates are distributed at the points in a regular hexagonal shape on a certain plane for simplicity schematically described Fig. 11. The value of $\lambda$ can be estimated as follows,24) based on the $r$ and volume fraction $f$ of the Al$_3$Sc precipitates:

$$\lambda = r[(8\pi/3\sqrt{3}f)^{1/2} - 2].$$

By considering the effect of Sc addition to Al on the resistivity ($\Delta\rho_{Al} = 34$ $\mu\Omega$m/at% $^{23}$), the amount of Sc consumed to form the Al$_3$Sc phase was derived from the resistivity variation $\Delta\rho$ before and after aging. Then, the value of $f$ could be evaluated using the molar volume of the Al$_3$Sc phase ($1.0425 \times 10^{-5}$ m$^3$/mol$^{-1}$). The $\Delta\rho$ values of the specimens forged to 3, 9 and 15 passes were approximately 4, 5, and 5 $\mu\Omega$m, and the corresponding $f$ values were estimated to be 0.30%, 0.37% and 0.37%, respectively. Finally, the $\lambda$ values were derived using eq. (2) with the $r$ values listed in Table 2. The results are summarized in Table 4. For the PA specimens, the $\lambda$ values were nearly identical regardless of grain size $d$.

Table 4 Volume fraction of Al$_3$Sc precipitates $f$, average precipitate radius $r$, and inter-precipitate spacing $\lambda$ for PA-specimens. Also shown are data for an OA-15 specimen (15-pass specimen over-aged at 473 K for 302.4ks).

| Specimen | $f$ [%] | $r$ [nm] | $\lambda$ [nm] |
|----------|---------|----------|----------------|
| 3-pass   | 0.30    | 3.5      | 133            |
| PA 9-pass| 0.37    | 4.0      | 136            |
| 15-pass  | 0.37    | 3.8      | 130            |
| OA 15-pass| 0.56   | 6.8      | 186            |

Fig. 10 Temperature dependence of normalized activation volume $V^*/b^3$ of (a) ST- and (b) PA-specimens. Also shown in (b) is data for an OA-15 specimen (15-pass specimen over-aged at 473 K for 302.4ks).

Fig. 11 Schematic illustration of the distribution model of precipitate particles.
On the other hand, it is reported that the Orowan mechanism is dominant (Fig. 12(a)), the $V^*$ value can be expressed as

$$V^* \equiv S^* b = 2r\lambda - (\lambda^2/4)[(2\theta - \sin 2\theta)/\sin^2 \theta]$$

where, $S^*$ is the activation area schematically indicated by the gray hatch in Fig. 12(a) and $\theta$ is the bowing-out angle of the dislocation. The $V^*$ values for the all PA specimens were experimentally estimated by the strain-rate jump tests to be about 300$b^3$ (see section 3.2 and Fig. 10(b)). In this case, based on eq. (3) and the values in Table 4, $\theta$ must be 8–9°. On the other hand, it is reported that the Orowan mechanism is dominant when the radius of Al$_3$Sc precipitates exceeds 2.4 nm in an Al–2.0 mass% Mg–0.2 mass% Sc alloy, which contained the same volume fraction of precipitates as in the present alloy.28) In all the PA specimens, the $r$ values were larger than the critical radius ($\approx 2.4$ nm) (Table 4). Hence, the Orowan mechanism should be dominant in the PA specimens. Moreover, if the dislocations overcame the precipitates from the state of bowing-out with $\theta \approx 90°$ (Fig. 12(b)), the $V^*$ value can be expressed as

$$V^* \equiv S^* b = (\pi/8)[(\lambda + 2r)^2 - \lambda^2]b.$$

Based on the $\lambda$ and $r$ values in Table 4, the resulting $V^*$ is approximately 9800$b^3$, which is over one order of magnitude larger than the experimental values (Fig. 10). Thus, the model illustrated in Fig. 12(b) cannot reasonably explain our experimental results.

On the other hand, if the Orowan mechanism is the deformation rate-controlling process, the dislocation of the critical state probably bypasses the precipitates with a slight increase in $\theta$ ($\Delta\theta$) from the semi-circular shape (Fig. 13). In this case, the $S^*$ can be expressed as

$$S^* = V^*/b = (\pi\lambda^2/8)[(1/\cos^2 \Delta\theta) + (\Delta\theta/180°) - 1] + (\lambda^2 \tan \Delta\theta)/4.$$

Using eq. (5) and the $\lambda$ values in Table 4, the $V^*$ values were considered again. In this case, the $\Delta\theta$ required by the dislocations to bypass the precipitates via the Orowan mechanism should be significantly small; thus, the value of $V^*$ was calculated in the range of $\Delta\theta = 0.1–0.5°$. The values were almost the same for all PA specimens, ranging from $150b^3$ ($\Delta\theta = 0.1°$) to $800b^3$ ($\Delta\theta = 0.5°$). Moreover, these values were in the same order as the experimental ones (Fig. 10(b)). Therefore, the $V^*$ values experimentally obtained for the PA specimens are reasonably understood from dislocation bypass of the Al$_3$Sc precipitates by the Orowan mechanism as illustrated in Fig. 13. The model discussed above is qualitatively consistent with the study by Marquis et al., in which the Orowan mechanism controlled the yield stress of an Al–Mg–Sc alloy when the radius of Al$_3$Sc precipitates is larger than 2.4 nm.28) If the precipitate size is relatively large and cannot be shear deformed by mobile dislocations, the precipitates can be considered as “strong long-range obstacles” against the dislocation motion.23) Since such obstacles are generally considered athermal,23) the strain rate and temperature dependencies of the flow stress become extremely small.29) And so, in the PA specimens with almost constant $r$ and $\lambda$ (Table 4), all the $V^*$ values were almost identical (Fig. 10(b)) and no temperature dependence appeared. To verify this conclusion, a portion of the specimen forged with 15 passes was subjected to over aging at 473 K for 302.4 ks; this specimen was referred as OA-15. Both the $r$ and $\lambda$ of the OA-15 specimen increased by the over aging compared with those of the PA-15 specimen (Table 4). The OA-15 specimen was also subjected to strain-rate jump tests to investigate the temperature dependence of its $V^*$. The results are displayed in Fig. 10(b). As for the PA specimens, the $V^*$ showed no temperature dependence. Furthermore, its values were relatively larger than those of the OA-15 specimen. This is because the value of $V^*$ monotonically increases along with $\lambda$ according to eq. (5). These results strongly support the conclusion stated above.

5. Conclusions

The temperature dependence of the deformation behavior of ultrafine-grained Al–Mg–Sc alloys, having three different
grain sizes fabricated by various numbers of multi-directional forging (MDF) passes at room temperature, was systematically investigated. The results yielded are summarized as follows.

(1) In the specimens forged within 3 and 9 passes, which had relatively large grain sizes of 950 and 680 nm, the activation volume of plastic deformation showed a positive temperature dependence. In contrast, that of the specimen prepared via 15 MDF passes, which had the smallest grain size of 360 nm, exhibited a negative temperature dependence. These results can be reasonably explained by the depinning model in which the rate-controlling process of deformation transitions with decreasing grain size from the interaction between mobile and forest dislocations to the dislocation bowing-out from boundaries.

(2) The activation volume of plastic deformation in the peak-aged specimens with dispersed fine Al3Sc precipitates showed no temperature dependence and was nearly identical regardless of the number of MDF passes, i.e., grain size. The Al3Sc precipitates acted as strong obstacles against mobile dislocations, and thus, the deformation rate was controlled by the interaction between mobile dislocations and precipitates. Therefore, dislocation bypass of the Al3Sc precipitates by the Orowan mechanism should work as a dominant rate-control process of deformation independent of grain size.

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