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Enhanced ductility and deformation mechanisms of ultrafine-grained Al–Mg–Si alloy in sheet form at warm temperatures

W.J. Kim* and S.J. Yoo

Department of Materials Science and Engineering, Hong-Ik University, 72-1, Sangsu-dong, Mapo-gu, Seoul 121-791, Republic of Korea

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Ultrafine-grained Al–Mg–Si alloy processed by high-ratio differential speed rolling exhibited significantly enhanced tensile ductility at warm temperatures (493–553 K). The dominant deformation mechanism of the alloy was found to be pipe-diffusion-controlled grain boundary sliding associated with a stress exponent of 4. According to the analyses based on Hart's necking criterion, in addition to increased strain rate sensitivity, increased strain hardening by accelerated dynamic grain growth notably contributes to the ductility of ultrafine-grained aluminum.

Keywords: Severe plastic deformation; Aluminum alloys; Ductility; Ultrafine-grained microstructure

Severe plastic deformation (SPD) techniques offer the possibility of improving the ductility of aluminum through effective grain refinement to the (sub-)grain size at the submicrometer level. Sabirov et al. [1] reported enhanced ductility at room temperature in ultrafine-grained (UFG) 6083 Al in tension and attributed it to the increased strain rate sensitivity m by grain refinement. May et al. [2] investigated the strain rate sensitivity of commercial purity aluminum with ultrafine grain size in the temperature range 298–523 K in compression and found that, compared with coarse-grained Al, the m value of UFG Al increased more with increasing temperature.

In this paper, the tensile ductility behavior of UFG 6061 Al sheets fabricated using high-ratio differential speed rolling (HRDSR) was examined at temperatures in the range 493–553 K (0.58–0.65 T_m, where T_m is the absolute melting point of the 6061 alloy) to understand the rate-controlling deformation mechanisms and the factors for improving the tensile ductility of UFG aluminum at warm temperatures. It has been well proved that the deformation mechanism of UFG aluminum alloys is either dislocation glide or grain-boundary-diffusivity-controlled grain boundary sliding (GBS) at high temperatures >573 K [3]. The tensile ductility behavior and deformation mechanisms of UFG Al at warm temperatures, however, have rarely been systematically studied, though there was some speculation that the increased m value by grain refinement might be the result of the increased contribution of GBS or coble creep to plastic flow [2]. Deformation of UFG 6061 Al at warm temperatures is important, as the alloy does not possess special elements such as Zr and Sc that form thermally stable intermetallic compounds, such that the high strength gained through SPD can be easily lost during deformation at high temperatures owing to rapid recovery or/and grain coarsening.

The initial material was commercially available 6061 plates received in T6 condition. The plates were solid-solutionized at 803 K for 4 h and then cooled to room temperature following two different routes: rapid quenching into room-temperature water to produce solute-supersaturated material (WQ, water quenched), and slow cooling in the furnace to produce fully annealed material (FC, furnace cooled). The diameters of the upper and lower rollers used for the HRDSR process were identical, and the speed ratio of the upper to lower roller was 3. The rolls and samples were preheated to the same temperature of 423 K before HRDSR. The plates with a width of 100 mm were rolled from 2 to 0.6 mm (70% in thickness reduction) in a single pass.

Tensile elongation tests were carried out at 493, 523 and 553 K for different strain rates under constant cross-head speed conditions. Before tensile testing, the samples were heated to the test temperature within

* Corresponding author. Tel.: +82 2 320 1468; fax: +82 2 325 6116; e-mail: kimwj@wow.hongik.ac.kr

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5 min. The test temperature was held constant for an additional 5 min before tensile testing. Strain rate change (SRC) tests were conducted at 523 K for the HRDSR processed Al by varying the cross-head speed repeatedly between 0.3 and 0.06 mm min^{-1}.Tensile specimens with a gauge length of 5 mm parallel to the rolling direction were used for the tensile tests, and scratch marks on the gauge region were used to evaluate the tensile elongation prior to fracture.

The samples of interest were electropolished, and the grain structure was recorded by orientation imaging microscopy (OIM) using the electron backscattered diffraction (EBSD) technique integrated in an Amray scanning electron microscope (controlled and analyzed using TSL software). An area 12 × 34 μm was scanned on the normal direction plane, and the step size was taken to be 0.04 μm in all scans. The grain tolerance angle was 5° to get grain size distribution and evaluate the average grain size.

The EBSD data for FC-HRDSR and WQ-HRDSR 6061 Al alloys [4] are presented in Table 1. The FC-HRDSR sample with a yield stress (YS) of 276 MPa has the mean gain size d of 2 μm and a low fraction of high-angle boundaries (0.3) [4]. In contrast, the WQ-HRDSR alloy, which is considerably stronger (YS 434 MPa) than the FC-HRDSR 6061 Al, has a much smaller d (0.37 μm) and a higher fraction of high-angle boundaries (0.48).

Figure 1a shows the true stress–true strain curves of the as-received, the FC-HRDSR and the WQ-HRDSR 6061 Al at 523 K. The following are noted when the three materials are compared at the same strain rate of $1 \times 10^{-3}\ s^{-1}$. First, flow stress decreases as the grain size decreases. Second, strain hardening becomes more pronounced as the grain size decreases. Third, tensile elongation increases as the grain size decreases. The tensile elongation values of the three materials at temperatures of 493, 523 and 553 K for different strain rates, including the data shown in Figure 1a, are presented as a function of strain rate in Figure 1b. The FC-HRDSR Al exhibits enhanced ductility compared with the as-received alloy, especially at low strain rates. The ductility improvement is much more pronounced in the WQ-HRDSR alloy, where an increase in ductility by factors of 4–5 is achieved in the entire strain rate range.

The microstructures of the WQ-HRDSR and FC-HRDSR 6061 Al just before tensile loading for the elongation-to-failure tests at 523 K were investigated using EBSD to check possible microstructural change during the sample heating and holding stages. Their EBSD grain boundary maps are shown in Figure 2a and b, and the analysis results are summarized in Table 1. Notable microstructural change occurred. In the FC-HRDSR Al (Fig. 2a), the grain size increased from 2.08 to 2.97 μm, with little change in fraction of high-angle boundaries (0.3 vs 0.28). In the WQ-HRDSR Al (Fig. 2b), the increase in grain size was less (0.37 vs 0.48 μm), but the fraction of high-angle boundaries has largely increased from 0.48 to 0.66. This difference in microstructure evolution during annealing seems to be related to the difference in dislocation density within grains, as it was shown in previous work [5] that the WQ-ECAP (equal channel angular pressing) 6061 Al possessed much higher dislocation density within grains than the FC-ECAP alloy. When dislocation density is high inside the grains, transformation of low-angle to high-angle boundaries due to absorption of dislocations into the boundaries by recovery can be promoted, and the increase in grain size by grain growth can be largely canceled out by the formation of new grain boundaries due to the transformation of grain boundary misorientation angles. EBSD grain boundary maps for the grip and gauge regions of the fractured sample of WQ-HRDSR 6061 Al tested at 523 K at $1 \times 10^{-3}\ s^{-1}$ are shown in Figure 2c and d, respectively. Grain size has noticeably increased in both regions (grip region 1.2 μm, gauge region 1.9 μm), but more extensively in the gauge region where dynamic grain growth and grain elongation to the tensile axis took place. Nevertheless, the deformed region retained a reasonably small grain size as well as a high fraction of high-angle boundaries (0.5).

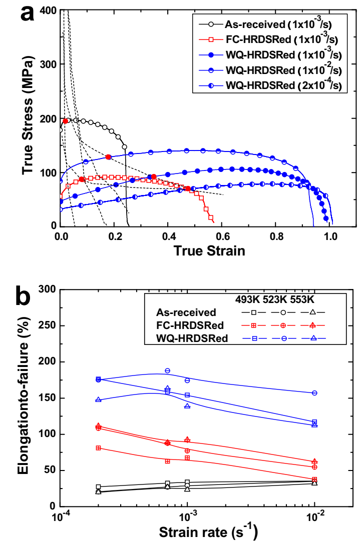

Figure 1. (a) True stress–true strain curves of the as-received, FC-HRDSR and WQ-HRDSR 6061 Al at 523 K. The dashed lines represent Hart's necking criterion: $\sigma = \frac{1}{1-m}\frac{d\sigma}{d\varepsilon}$. The intersection points are indicated by red-filled symbols. (b) Tensile elongation data of the three materials at temperatures 493–553 K given as a function of strain rate.

Table 1. EBSD data for HRDSR alloys before and after deformation.

Materials	FC-HRDS R	WQ-HRDSR	FC-HRDSR just before the tensile loading (523 K)	WQ-HRDSR just before the tensile loading (523 K)	WQ-HRDSR after fracture (grip) (523 K, 1×10^{-3}/s)	WQ-HRDSR after fracture (gauge) (523 K, 1×10^{-3}/s)
Mean grain size (μm)	2.08	0.37	2.97	0.48	1.20	1.91
Average misorientation angle	14.49	20.57	11.88	28.88	23.23	22.03
Fraction of high-angle boundaries	0.30	0.48	0.27	0.66	0.53	0.50