Superplastic Behavior in Al–Li–Mg–Cu–Sc Alloy Sheet*

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Superplastic properties of an Al–Li–Mg–Cu–Sc alloy subjected to hot rolling at 573 K with a reduction of 90% were examined in the temperature range of 673–798 K and an initial strain rate of $1.4 \times 10^{-2}\,s^{-1}$. It was found that partially recrystallized microstructure evolved under the hot rolling resulting in slight anisotropy of superplastic behavior. The maximum elongation-to-failure of 415% appeared in the rolling direction at 723 K while in the transverse direction the maximum ductility of 385% was achieved at a higher temperature of 773 K. Microstructure evolution and cavitation during superplastic deformation were examined.

Keywords: *aluminium–lithium alloy, superplasticity, hot rolling, mechanical properties, microstructure*

1. Introduction

Aerospace industry has a substantial interest in the fabrication of complex parts from new aluminum–lithium alloys via superplastic forming.[1] Recently a new commercial Al–Li–Mg–Cu–Sc alloy, designated in Russia as 1443 aluminum alloy and denoted as 1443 Al herein, was developed for application as an advanced sheet material for fuselage skin of civil airplanes.[2] The 1443 Al is a unique high strength aluminum–lithium alloy exhibiting an optimal combination of yield strength and crack propagation resistance.[2] Besides, it has improved workability and can be processed into thin sheets with a thickness less than 1 mm. So the 1443 Al can be suitable for superplastic forming if a simple and inexpensive processing route for achieving superplasticity in this material will be developed. In addition, the 1443 Al contains $Al_3(Sc, Zr)$ dispersoids and, therefore, it is expected that this alloy is capable of achieving superplastic ductilities at high strain rates that promotes the commercial application of this material.

It is known[1,3] that aluminum alloys exhibit superplastic properties in two different microstructural conditions. The first one is a fully recrystallized microstructure with the mean grain size less than 10 µm.[1,4] In this condition most of aluminum–lithium alloys exhibit superplastic behavior.[1,5] However, the formation of fully recrystallized structure in these alloys requires the fairly complex thermomechanical processing (TMP) routes. Uniform and fine-grained structure can be produced via TMP of sheet material[1,5] or by intense plastic straining.[6,7] In contrast, superplasticity in aluminum–lithium alloys with unrecrystallized structure subjected to extensive warm working,[8–10] which is the second microstructural condition, can be achieved via a simple processing route. In this condition a fine-grained structure is evolved during the early stages of superplastic deformation after the warm working at 573 K.[10] Nowadays, the last processing is considered as the simplest route of TMP to make an aluminum alloy superplastic.

Thus, the aim of present study is to demonstrate a feasibility of achieving superplastic properties in the 1443 Al by subjecting this material to extensive hot rolling.

2. Experimental Procedure

An ingot of the 1443 Al with a chemical composition of Al–1.9Li–1.0Mg–1.7Cu–0.03Sc–0.08Zr (in mass %) was manufactured by direct chill casting followed by solution treatment at 803 K for 20 hours. Then it was cut into rectangular preforms having dimensions of $60 \times 40 \times 20\,mm^3$.

Hot rolling of the 1443 Al preforms was carried out in a duo rolling mill with 300 mm-diameter rolls using a speed of 100 mm/s and 10–20% nominal reduction per pass to the total reduction in thickness of 90%. Prior to rolling the preforms were soaked at 573 K for 15 minutes. The time of interpass reheatings was 5 minutes.

Flat samples with a gauge size of $6 \times 3 \times 2\,mm^3$ cut both along the rolling and the transverse directions of the sheet were tensile tested using a Shimadzu AG-G machine at temperatures in the range of 673–798 K and an initial strain rate of $1.4 \times 10^{-2}\,s^{-1}$. Each sample was held at the testing temperature for about 20 min in order to achieve thermal equilibrium.

Microstructure analysis was carried out using an optical microscope Olympus BX-60 and a transmission electron microscope Hitachi H-650 operating at an accelerating voltage of 200 kV. Grain boundary misorientation distributions were obtained from electron backscattering diffraction pattern (EBSP) using JEOL JXA8100 electron probe microanalyzer with OIM software provided by TexSEM Lab., Inc. Volume fraction of cavities was measured on as-polished and unetched samples, which were deformed to failure, using the standard point-count technique.[11]

3. Results and Discussion

Figure 1 shows typical microstructure of the hot rolled 1443 Al. One can see that hot rolling at 573 K resulted in the formation of partially recrystallized structure. Fine recrystal-

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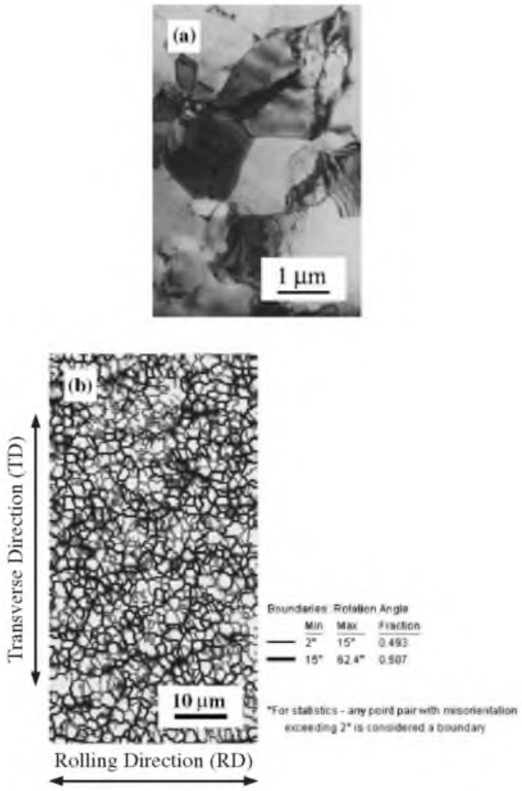

Fig. 1 Typical microstructure of the 1443 Al after hot rolling at 573 K: (a) TEM; (b) grain boundary misorientation map.

lized grains with sizes zanging from 1 to 3 µm (Fig. 1(a)) appeared mainly near the coarse particles of the secondary phases. EBSP analysis data show that fraction of the volume recrystallized was 65–70% (Fig. 1(b)). The boundary misorientation distribution was found to be distinctly bimodal. Almost 50% of boundaries had low-angle origin (Fig. 1b).

The tensile deformation curves of the samples cut along the rolling and the transverse directions of the sheet are given on Fig. 2. Extensive strain hardening took place at the beginning of deformation followed by steady-state flow stage. An increase in the testing temperature resulted in decreasing flow stress and a reduction in initial work hardening. The steady state flow stress decreased gradually from 45 MPa to about 15 MPa in both directions with increasing temperature from 673 to 798 K.

The temperature dependence of total elongation at a strain rate of $1.4 \times 10^{-2}\,s^{-1}$ is shown on Fig. 3. It is seen that a well-defined maximum in elongation-to-failure appeared for both types of samples. However, the samples cut along the rolling direction exhibited a higher ductility of 415% at 723 K, while the maximum elongation-to-failure of the samples cut along the transverse direction was observed at a higher temperature of 773 K (Fig. 3).

OM analysis of the grip sections of deformed samples revealed the microstructural evolution of the hot rolled 1443 Al under static annealing (Figs. 4(a) and 5(a), Table 1). Initial recrystallized grains were essentially stable under static annealing up to 773 K (Table 1). At the same time, the bands of coarse grains with sizes ranging from 10 to 50 µm and unrecrystallized areas (Figs. 4(a) and 5(a)) were observed

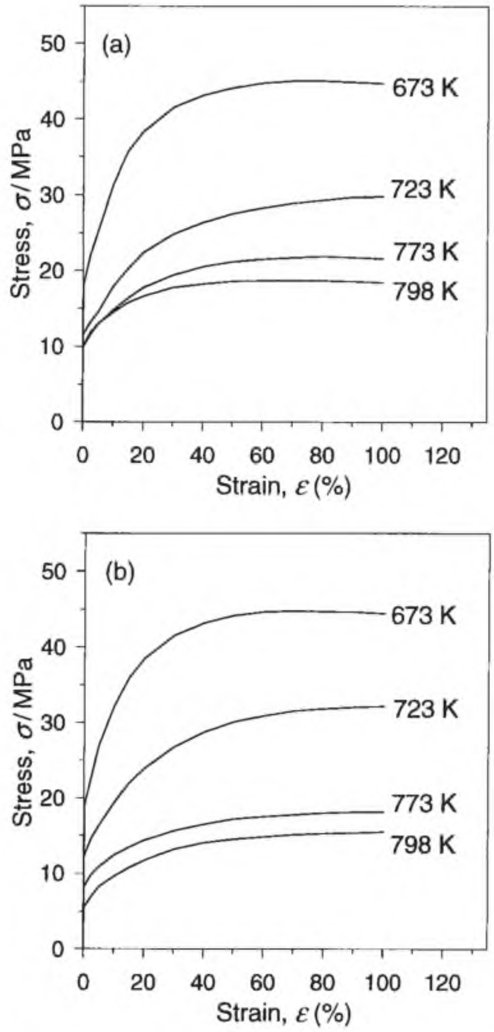

Fig. 2 Stress-strain curves of the 1443 Al sheet samples cut along (a) the rolling direction and (b) the transverse direction. Initial strain rate was $1.4 \times 10^{-2}\,s^{-1}$.

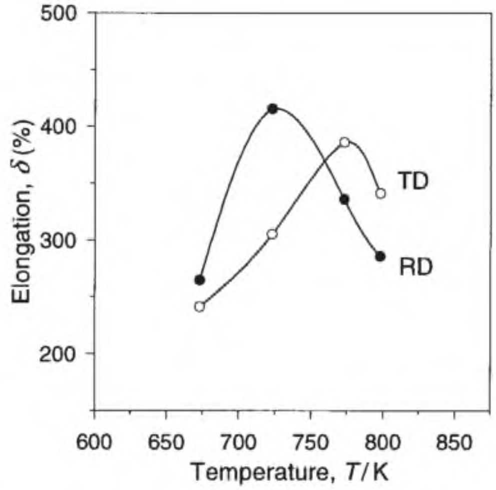

Fig. 3 Temperature dependence of the elongation-to-failure of the 1443 Al sheet samples cut along the rolling (RD) and the transverse (TD) directions. Initial strain rate was $1.4 \times 10^{-2}\,s^{-1}$.

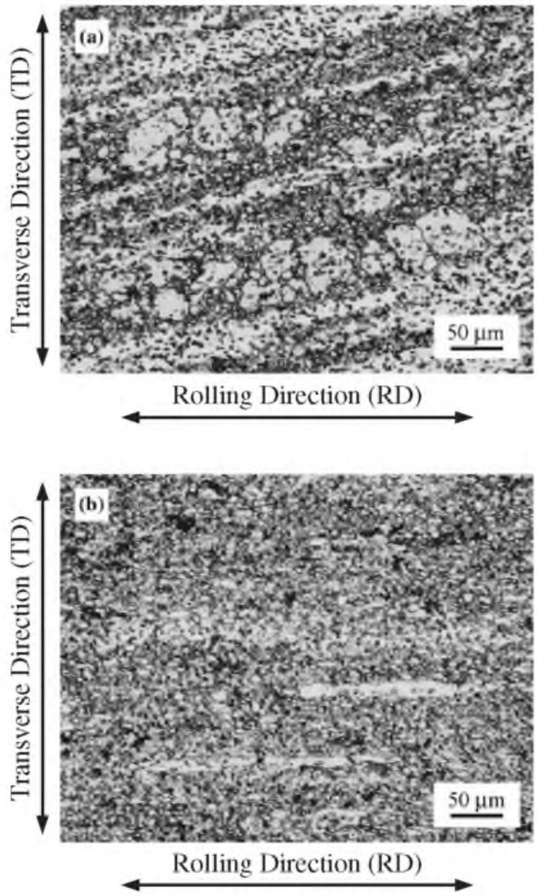

Fig. 4 OM photomicrographs of (a) grip and (b) gauge sections of deformed samples cut along the rolling direction of the sheet ($T = 723$ K; $\dot{\varepsilon} = 1.4 \times 10^{-2}$ s^{-1}; $\delta = 415\%$). Tensile direction is horizontal.

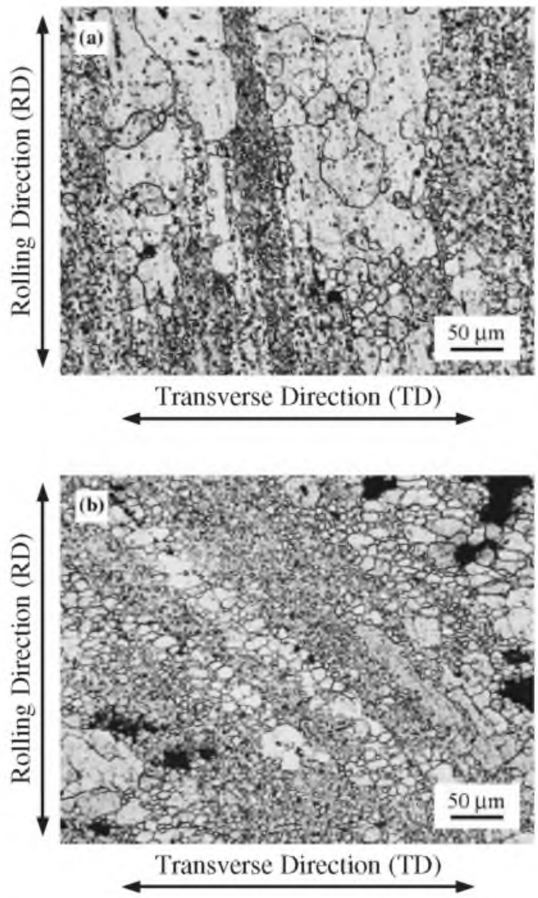

Fig. 5 OM photomicrographs of (a) grip and (b) gauge sections of deformed samples cut along the transverse direction of the sheet ($T = 773$ K; $\dot{\varepsilon} = 1.4 \times 10^{-2}$ s^{-1}; $\delta = 385\%$). Tensile direction is horizontal.

along the former rolling direction of the sheet.

Superplastic deformation led to an increase in the volume fraction of fine recrystallized grains (Figs. 4(b) and 5(b), Table 1). In the samples cut along the rolling direction, the volume fraction of unrecrystallized areas decreased to as low as 5% (Fig. 4(b) and Table 1) under the condition, at which the highest ductility was attained. In those samples no coarse recrystallized grains were found. In the samples cut along the transverse direction, the well-defined bands of coarse recrystallized grains were observed (Fig. 5(b)). It is interesting to note that superplastic deformation resulted in rotation of those bands to angles close to 45° to tensile axis (Fig. 5(b)) as compared to microstructure after static annealing (Fig. 5(a)). In addition, the size of the coarse grains in the gauge sections of the samples was less than in the grip sections (Fig. 5).

An extensive cavitation occurred during superplastic deformation at temperatures higher than 723 K (Table 1). At those temperatures, the pseudo-brittle failure occurred as a result of the nucleation, growth and interlinkage of internal voids.[1] The volume fraction of cavities near the fracture approached to 9.5% that is typical for such failure.

Thus, the results of the present study demonstrated that the 1443 Al is capable of superplastic deformation in partially recrystallized condition as the other alloy belonging to the similar system.[8] Dispersion particles prevent static recrystallization of warm worked 1443 Al, and effectively restrict

Table 1 Average size of recrystallized grains, d, after static (grip section) and dynamic (gauge section) annealing; volume fraction of unrecrystallized areas and coarse grains, V_{NR}, and fraction of cavities, V_C, in the gauge section at different temperatures and $\dot{\varepsilon} = 1.4 \times 10^{-2}$ s^{-1}. Corresponding elongations-to-failure of the samples, δ, are also indicated.

Testing conditions		Gauge section			Grip section	
T/K	δ (%)	V_{NR} (%)	d/μm	V_C (%)	V_{NR} (%)	d/μm
Rolling direction						
673	265	10	3	2.5	20	4
723	415	5	4.5	7.5	27	5
773	335	17	6	8.5	33	6
798	285	25	7.5	4.5	55	6.5
Transverse direction						
673	240	—	—	1.2	—	—
723	305	5	4.5	2.5	30	5
773	385	20	6	6.5	41	6
798	340	24	7	9.5	41	6.5

the migration of deformation-induced boundaries. As a result, the low-angle boundaries become progressively high-angle during deformation and thus able to contribute to grain boundary sliding.[1] Therefore, superplastic deformation induces high-angle boundaries via continuous dynamic recrystallization.[3] It is apparent, that extensive strain hardening at initial stages of superplastic deformation is associated with transformation of low-angle boundaries into true high-angle boundaries. The steady-state flow is attained when the fraction of fine recrystallized grains becomes enough for operation of grain boundary sliding. Some anisotropy of in-plane mechanical properties of the sheet seems to be caused by the line arrangement of coarse grains and unrecrystallized areas along the previous rolling direction. In the samples cut along the transverse direction, these bands appear to be effective stress concentrators resulting in the lower superplastic elongations at 673–723 K. Enhancing grain boundary sliding and increasing diffusion at higher temperatures (above 723 K) lead to re-orientation of the bands making them less active stress concentrators and thus, to a notable improvement of the ductility of the alloy.

4. Conclusions

(1) Hot rolling of the 1443 Al at 573 K resulted in the formation of partially recrystallized microcrystalline structure with the mean size of recrystallized grains of 3 μm and 30–35% of unrecrystallized areas containing recovered subgrains.

(2) Superplastic properties of the sheet were studied in the temperature range of 673–798 K and with the initial strain rate of $1.4 \times 10^{-2}\,s^{-1}$. The maximum elongation-to-failure of 415% was obtained at 723 K on samples cut along the rolling direction of the sheet.

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