SUBGRAIN FORMATION AND EVOLUTION DURING THE DEFORMATION OF AN Al–Mg–Sc ALLOY AT ELEVATED TEMPERATURES

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Introduction

Scandium is the only alloying element to form a thermally stable, coherent $L1_2$ phase, Al_3Sc, in aluminum (analogous to in Ni-based superalloys). The Al_3Sc precipitate is unusually resistant to coarsening. Despite the relatively low solubility of Sc and, hence, limited volume fraction of the Al_3Sc phase, it produces a significant strengthening effect. In fact, Al_3Sc is the most potent strengthener, on an equal atomic fraction basis, known in Al-base systems [1]. The Al_3Sc precipitate is also extremely effective in stabilizing substructure, thus allowing the use of strain-hardening and grain-boundaries strengthening to enhance the strength of Al alloys. In addition, the effectiveness of the Al_3Sc precipitate in pinning grain boundary can be utilized to produce fine-grained aluminum for forming operations, such as superplastic forming.

The high-temperature properties of the binary Al–Mg and the ternary Al–Mg–Sc alloys have been studied by Sawtell and Jensen [2]. The tensile elongation value for the Al–4Mg–0.5Sc alloy was quite remarkable (~1000%). The high elongation was apparently attributed to a fine distribution of Al_3Sc particles in the alloy. However, the exact role of Al_3Sc on the microstructural evolution in the Al-Mg alloy was not clear. Also, the strain rate sensitivity value was found to vary with strain rate and temperature, suggesting there might exist a change in deformation mechanisms. Several years ago, research in Russia also led to the development of an Al–6Mg–0.3Sc alloy (designated Al 1570). The purpose of this paper is to demonstrate the effectiveness of Al_3Sc in stabilizing the substructure/structure in aluminum and to relate the microstructural evolution to the formability of this alloy.

Experiments

The alloy used in the present study is the Russian Al 1570 (composition in weight%: Al–5.76Mg–0.32 Sc–0.3Mn–0.1Fe–0.2Si–0.1Zn). The alloy was initially produced by ingot casting. The as-received material was in a sheet form with a thickness of 1 mm. The sheet was cold rolled to 0.1 mm with two intermediate annealings each at 250°C for 30 min. The first annealing was performed after 50% reduction in thickness (i.e. final thickness = 0.5 mm), and the second was performed when the thickness was 0.2 mm. The final thickness of the sample was 0.09 mm, which represents a total reduction of 91%.

Tensile specimens were machined from the final sheet. Tensile tests were conducted in air at temperatures between 300 and 500°C, and at strain rates between 10^{-5} and 1 s^{-1}, using a screw-driven Instron machine equipped with a radiant furnace. The variation of the target temperatures was controlled to within ±1°C.

Microstructures of both the grip and gauge sections of tested specimens were examined using a JOEL-200CX transmission electron microscope operated at 200 KV. TEM foils were first sliced from the specimens, and were finally prepared by twin-jet electropolishing in a solution of 60% ethanol, 35% butyl alcohol, and 5% perchloric acid at 15V and -10°C.

Experimental Results

Microstructure

The microstructure of a sample before testing is shown in Figure 1. This microstructure is typical of a heavily deformed metal; it consists of a cellular structure with the cell size ranging from 100 nm to as large as 2 μm. The cellular structure was readily recovered upon annealing.

Shown in Figure 2 is a TEM dark-field image from the grip region of a sample tested at 350°C at a strain rate of 10^{-5} s^{-1}. The microstructure has a strong <011> rolling texture and contains low-angled grains (i.e. subgrains). The 350°C annealing (~16 h) apparently causes recovery and results in the formation of subgrains from the initial cellular structure. The average subgrain size is about 1 μm. Examination of the gage region (dynamically annealed) of the tested sample indicates that, except for having a slightly larger subgrain size (~2 μm), the microstructure is virtually the same as that in the grip region. The overwhelming presence of subgrain boundaries is attributable to the fact that the alloying of Mg to Al greatly reduces its stacking fault energy [3]. Dislocation recovery is expected to be difficult in an alloy with a low stacking fault energy and, thus, enhances the formation of subgrains in the alloy. The microstructure also revealed that fine ($L1_2$-type) Al_3Sc precipitates with particle size ranging from 10 to 100 nm were found to be uniformly distributed within Al grains. The spacing of the Al_3Sc particles ranged from 50 to 100 nm.

The grain structure was quite stable. For example, the microstructure from the grip region of a sample tested at 475°C at a strain rate of 10^{-2} s^{-1} is presented in Figure 3. Again, it consists of primarily low-

Figure 1. Sample microstructure prior to deformation.

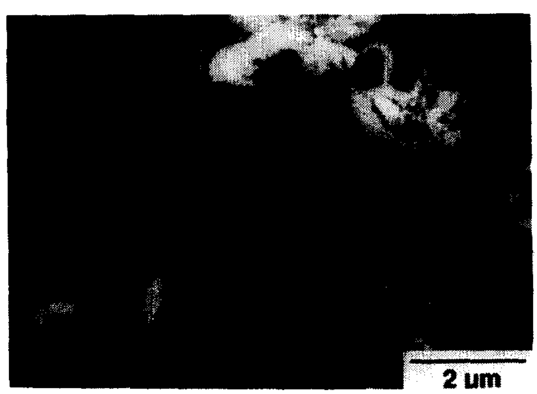

Figure 2. TEM dark-field image from the grip region of a sample tested at 350°C at a strain rate of 10^{-5} s^{-1}.

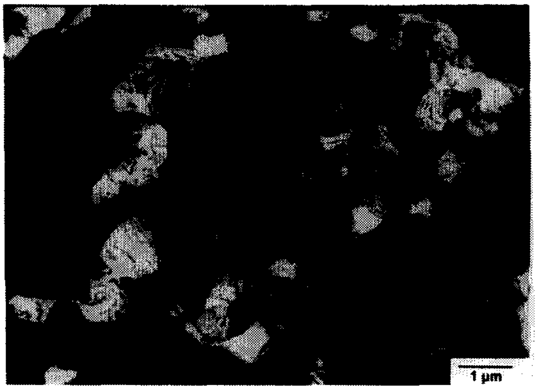

Figure 3. TEM dark-field image from the grip region of the sample tested at a strain rate of 10^{-2} s^{-1} and 475°C.

Figure 4. Microstructure of the gage region of the sample tested at a strain rate of 10^{-2} s^{-1} and 475°C. Stress axis is indicated and cavities at grain triple junctions are marked.

angled grain boundaries with boundary misorientation angles of less than 4–5°. The subgrain size is about 1 µm, which is slightly larger than that observed at 350°C (~1 µm). In comparison, the gage region exhibits a different structure (Figure 4). This shows a recrystallized microstructure in which the majority of grain boundaries are high-angled. The average grain size (~6 µm) is slightly larger than that of the subgrain size in the grip region. From a superplasticity point of view, a 5 µm grain size is considered to be fine for an aluminum alloy. The fine microstructure is evidently a result of the extremely fine (10–20 nm) and uniform distribution of the $L1_2$ Al_3Sc precipitates. In fact, these precipitates not only effectively pin high-angled grain boundaries but also subgrain boundaries, as illustrated in Figure 5.

Mechanical Properties

The true stress-true strain rate curves for Al 1570 tested at a true strain rate of 10^{-2} s^{-1} at different temperatures (350–500°C) are shown in Figure 6. At all temperatures there is an immediate hardening

Figure 5. Al_3Sc precipitates effectively pin subgrain boundaries.

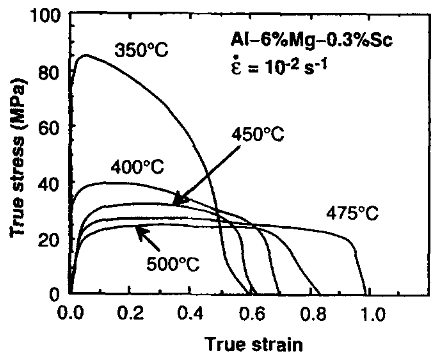

Figure 6. True stress-true strain rate curves for Al 1570 tested at a true strain rate of 10^{-2} s^{-1} at different temperatures (350–500°C).

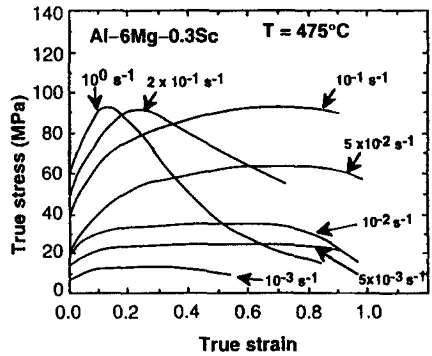

Figure 7. True stress-true strain rate curves for Al 1570 tested at 475°C and at different strain rates.

upon loading. It is noted that, at temperatures lower than 450°C, that this hardening was followed by a continuous softening. In contrast, at temperatures higher than 475°C, the hardening was followed by an apparent steady-state flow. The elongation value is a function of testing temperature and exhibits a maxi mum at 475°C.

The true stress-true strain rate curves for the alloy tested at 475°C and at different true strain rates are depicted in Figure 7. Except for strain rates faster than 2×10^{-1} s^{-1}, a region of steady state flow appears at all strain rates. Also noted is that, within the strain rates from 1.1×10^{-3} to 1.1 s^{-1}, the maximum tensile elongation occurs at strain rates of about $10^{-2} - 10^{-1}$ s^{-1}.

Experimental results obtained at various temperatures are summarized in Figure 8. At the top of Figure 8, the flow stress (at a fixed strain of 0.2) is shown as a function of strain rate. At the bottom of Figure 8, the elongation-to-failure is shown as a function of strain rate. The strain rate sensitivity value, m, in the equation $\sigma = k\dot{\varepsilon}^m$, is noted to increase with testing temperature. At 350°C, m is about 0.35 and increases to 0.45 at 475°C. Tensile elongation approximately follows the expected trend, i.e. a higher m value results in a larger elongation. The elongation is always less than 200%.

Discussion

Al–Mg alloys are known to exhibit Class 1 solid solution behavior, namely, deformation is controlled by solute-drag on gliding dislocations [4-7]. In the present study, the Al–Mg–Sc alloy possesses a fine-grained structure, thus, grain boundary sliding is also expected to operate under certain test conditions. Since solute-drag and grain boundary sliding are two independent mechanisms, the resultant deformation in the alloy is the summation of contributions from both mechanisms. That is, the deformation strain rate can be described by the equation:

$$\varepsilon_{total} = \varepsilon_{gbs} + \varepsilon_{drag} = = AD_{gbs}\sigma^2 + BD_L\sigma^3 \quad (1)$$

where ϵ_{total} is the total strain rate, ϵ_{gbs} and ϵ_{drag} are the strain rates caused by grain boundary sliding and solute drag, respectively, D_{gbs} and D_L are the grain boundary and lattice diffusion coefficients, respectively, σ is the flow stress, and A and B are material constants.

According to Equation (1), depending upon the test conditions, the strain rate sensitivity value m should have an upper bound value of 0.5 and lower bound value of 0.33. The experimental results (Figure 8) indeed showed that m ranges from 0.33 to 0.5. Specifically, at 350°C the m value is about 0.35. At this temperature, although the grain size appears to be fine, the grains are primarily subgrains (Figure 2). Sub-

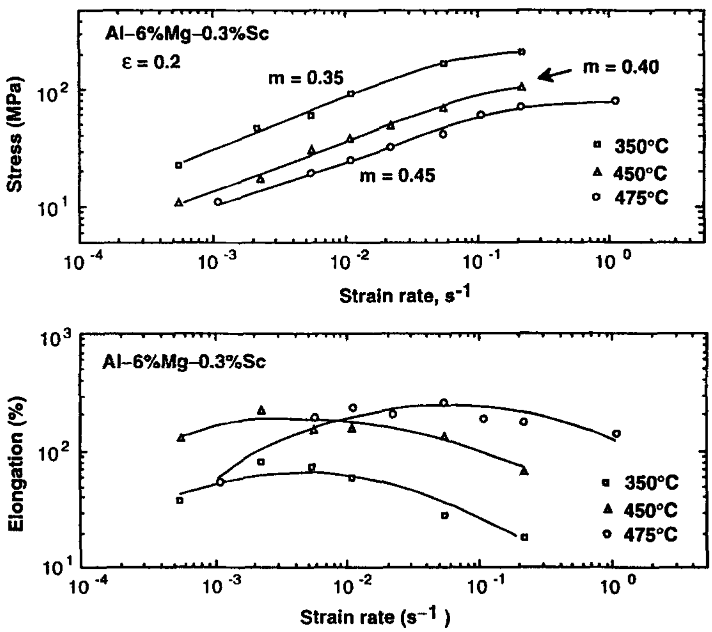

Figure 8. (Top) the flow stress (at a fixed strain of 0.2) is shown as a function of strain rate, and (Bottom) the elongation-to-failure is shown as a function of strain rate.

grain boundaries are generally immobile with respect to grain boundary sliding [8]. As a result, the grain boundary sliding process is not expected to prevail at 350°C and deformation would be mainly controlled by dislocation glide through the lattice (i.e. $m \sim 0.33$). This is consistent with the previous result that the solute-drag mechanism in Al-Mg usually takes place at intermediate temperatures around 300°C [5].

At 475°C the m value is about 0.45. At this high temperature, as a result of the pinning effects of the Al_3Sc particles, fine subgrains are still thermally stable under static conditions (Figure 3). Under stresses, however, high-angled grain boundaries rapidly evolve (Figure 4). These deformation-induced, high-angled, boundaries are readily able to slide and dominate the overall deformation in the sample. This results in a high strain rate sensitivity value, in the proximity of 0.5, reflecting the grain boundary sliding mechanism. Other microstructural evidence for the prevalent grain boundary sliding at high temperature is given in Figure 4, in which cavities at grain triple junctions are readily observed. These cavities were formed because grain boundary sliding was not properly accommodated. In contrast, only a limited amount of cavity formation was observed at grain triple junctions in samples deformed at 350°C. This is because dislocation glide is primarily an intragranular process which is not expected to lead to cavity formation at triple junctions.

A final comment is noted about the elongation: although the total elongation was less than 200%, it should be pointed out that these data were obtained from testing extremely-thin samples (90 μm). Tensile elongation is expected to be strongly sensitive to surface defects on thin samples. In fact, our most recent data measured from thick samples (~2 mm) indicated that elongation can be over 700% at 475°C.

Summary

The microstructure and mechanical properties of a commercial Al–6Mg–0.3Sc (Russian Al 1570) were characterized. The presence of Sc results in the uniform distribution of fine $L1_2$ precipitates which stabilize the grain substructure/structure in the alloy. At an intermediate temperature of 350°C, subgrains are formed and they are stable even under a dynamic condition (i.e. under stress). The deformation of Al 1570 at this temperature is controlled by solute drag on gliding dislocations. Thus, the strain rate sensitivity value is about 0.33. At a high temperature of 475°C, fine subgrains are still preferentially formed under static conditions (i.e. without stress). But, under a dynamic condition the low-angled subgrain boundaries quickly convert into high-angled grain boundaries and lead to extensive grain boundary sliding. Therefore, the dominant deformation mechanism of the alloy at 475°C is grain boundary sliding, and the alloy exhibits a strain rate sensitivity value close to 0.5.

Acknowledgment

This work is partially supported by the Army Research Office and performed under the auspices of the U.S. Department of Energy by LLNL (contract No. W–7405–Eng–48).

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