Recrystallization Behavior in an Al–Cu–Mg–Fe–Ni Alloy with Trace Scandium and Zirconium

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The Vickers hardness variation of a cold-rolled Al–Cu–Mg–Fe–Ni alloy to which trace scandium and zirconium were added had been measured during annealing. The microstructures of alloy were observed using optical microscope and transmission electron microscope (TEM). It was shown that the dispersed $\text{Al}_3(\text{Sc}, \text{Zr})$ particles could pin dislocations, stabilize substructures and impede the movement of boundaries. Therefore, the recrystallization of the Al–Cu–Mg–Fe–Ni alloy was retarded. The mechanical properties of the alloy with $\text{Al}_3(\text{Sc}, \text{Zr})$ particles at both ambient and elevated temperatures were improved. Subgrain coalescence was involved in the recrystallization process.

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1. Introduction

Aluminum alloy (AA) structural materials resistant to elevated temperature (from 373 to 473 K or even higher) are required to obtain the speed and payload necessary to make aircraft economically feasible.¹⁾ A wrought aluminum alloy commonly favored for elevated temperature applications is AA2618 (Al–2.4%Cu–1.5%Mg–1%Fe–1%Ni) which is based on Al–Cu–Mg system. Compared with other 2XXX series alloys, the mechanical properties of AA2618 at both ambient and elevated temperatures were improved due to the precipitation of GP (Cu, Mg) zones (also known as GPB zones) and the formation of the semi-coherent phase S' and the stable particles of the Al_3FeNi phase.²⁾

During the past a few years, there have been several publications concerning the effect of the addition of transition metals, such as scandium, to aluminum alloy.³⁾ It is reported that the influence of scandium on the microstructure evolutions and mechanical properties of aluminum and its alloys is similar to that of other transition metals, but more pronounced. Yelagin⁴⁾ found that the zirconium additions in aluminum formed a metastable phase Al_3Zr ($\text{L}1_2$ structure with a lattice parameter of 0.408 nm) with crystal lattice very close to that of phase Al_3Sc ($\text{L}1_2$ structure with a lattice parameter of 0.410 nm). In addition, scandium and zirconium were similar in nature and could form a continuous solid solution with each other. As a result, an $\text{Al}_3(\text{Sc}, \text{Zr})$ phase of variable compositions could form. This $\text{Al}_3(\text{Sc}, \text{Zr})$ phase also had a $\text{L}1_2$ type structure and was characterized by high structural and dimensional match with the aluminum matrix like the Al_3Sc phase. The objective of the present investigation is to study the influences of $\text{Al}_3(\text{Sc}, \text{Zr})$ on the recrystallization behavior of cold-rolled AA2618.

2. Experimental Methods

The composition (mass%) of the alloy used in the experiments was as follows: Al–2.23%Cu–1.21%Mg–0.93%Fe–1.09%Ni–0.30%Sc–0.30%Zr (Alloy A). The behavior of

AA2618 (Alloy B) was also studied for comparison. The cast billets of 16 mm thick were homogenized at 773 K for 10 hours, then hot-rolled at 723 K into plates of 7 mm thick, and cold-rolled to 4 mm thick. After annealed at 723 K for 1 hour, the plates were finally cold-rolled into sheets of 2 and 1 mm in thickness. The cold reductions were 50 and 75%, respectively. The specimens were annealed at different temperature (from 423 to 773 K) for 1 hour. It is well known that the recrystallization of aluminum alloys depends on several major factors including the amount of prior deformation and annealing time.⁵⁾ In the current study, the recrystallization temperature was measured under the condition of a great prior deformation and a confined annealing time (3.6 ks).

The hardness measurement was performed at room temperature using a HVA-10A vickers hardness tester. Microstructures of alloys were observed using a Nephoto-2 optical microscope and H-800 transmission electron microscope. Tensile test was performed on an Instron 8032 servo hydraulic testing system with computerized data acquisition. The tensile specimens were kept under test temperature for about 900 s, prior to the tensile test at a displacement rate of 0.0167 mm/s.

3. Results

The hardness variations of Alloys A and B at different temperatures are shown in Fig. 1. It is worth noting that the hardness of Alloy B decreases significantly in the temperature range of from 473 to 573 K. It is evident that in this temperature range, Alloy B has been recrystallized after annealing, as shown in Figs. 2(a) and (b). The recrystallized grains are shown clearly in Fig. 2(b) as the alloy is annealed at 623 K for 3.6 ks. But under the same condition, the recrystallization features in Alloy A (Figs. 2(c) and (d)) is not clearly identified, as in Alloy B. The hardness of Alloy A does not decrease as much as Alloy B (Fig. 1).

In order to obtain the temperature at which Alloy A begins to recrystallize, the hardness difference of Alloy A under different deformations is measured as a function $\Delta HV =$

$HV\varepsilon_1 - HV\varepsilon_2^{5)}$ (where $\varepsilon_1 = 75\%$, $\varepsilon_2 = 50\%$). The results are shown in Fig. 3. The temperature range (from 673 to 723 K) in which the value of ΔHV decreased greatly may be treated as the beginning of primary recrystallization.

The TEM micrograph of Alloy A in the as-cast condition (Fig. 4) shows that the $Al_3(Sc, Zr)$ particles (with double-arc contrast) are coherent with the matrix. The $[111]$ orientation diffraction pattern from $Al_3(Sc, Zr)$ particles shows the superlattice spots due to the ordered nature of the $L1_2$ structure. Based upon the previous investigation³⁾ and the Al-Sc-Zr phase diagram,⁶⁾ the chemical formula of $Al_3(Sc, Zr)$ composite particles might be more accurately written as $Al_3Sc_{1-x}Zr_x$, where x is a variable which depends on the content of zirconium in alloy. The incubation time for $Al_3(Sc, Zr)$ precipitation is quite short (about 10^2 seconds) and it is approximately 10^4 faster than other precipitation in Al-transition metal alloys (such as Al_6Mn or Al_3Zr precipitation in Al-Mn or Al-Zr alloy, respectively). The $Al_3(Sc, Zr)$ particles precipitate during aging or hot-rolling as in some Al-Sc and Al-Mg-Sc alloys.⁷⁾

The TEM micrographs of Alloy A (Fig. 5(a)) show that the dislocations are pinned and the movement of boundaries are impeded by the $Al_3(Sc, Zr)$ particles at a high temperature. Subgrain coalescence is involved in the recrystallization process as shown in Fig. 5(b).

The tensile properties for Alloy A at 293, 473, 523 and 573 K are compared with those of Alloy B in Fig. 6. It is worth noting that the yield strength and ultimate strength of Alloy A at both ambient and elevated temperatures are obviously greater than those of Alloy B.

4. Discussion

It is well known that the second phase in the alloy will influence the process of recrystallization.⁸⁾ In the alloy AA2618, the typical hard particles Al_9FeNi act as preferential sites for increasing dislocation density during cold work because the particles produce a turbulent and complex deformation pattern around themselves. As a result, the particles provide preferential sites for the nucleation of recrystallization during aging because of the increased strain energy in the matrix around these particles. But the $Al_3(Sc, Zr)$ particles which are dispersed and coherent with the matrix have the ability to prevent the recrystallization from happening. The dislocation density around $Al_3(Sc, Zr)$ particles is not very high because of their coherency. The force applied by $Al_3(Sc, Zr)$ particles to retard the recrystallization can be expressed as:⁹⁾

$$P_Z = 3f\gamma/2r, \quad (1)$$

where γ is the boundary energy, f the volume fraction and r the average size of the dispersoids. And the critical nucleus radius for recrystallization R_C is given by:⁹⁾

$$R_C = 2\gamma/\Delta G_V, \quad (2)$$

where ΔG_V is the difference of Gibbs free energies between the new and old phases. Based upon Ref. 9) and eqs. (1) and (2), R_C can be expressed as:

$$R_C = 2\gamma/[(\rho + \Delta\rho)\mu b^2/2 - 3f\gamma/2r] \quad (3)$$

where ρ is the dislocation density, and $(\rho + \Delta\rho)\mu b^2/2$ may be taken as the driving force for recrystallization. When

Fig. 1 Hardness-temperature curves of alloy plates.

Fig. 3 Hardness difference of Alloy A under different deformations.

Fig. 2 Optical morphologies of alloys at different temperature. (a) Alloy B, 473 K (b) Alloy B, 623 K (c) Alloy A, 473 K and (d) Alloy A, 673 K.

Fig. 4 $\text{Al}_3(\text{Sc,Zr})$ particles in Alloy A ingot.

Fig. 5 TEM morphologies of Alloy A at (a) 723 and (b) 773 K.

Fig. 6 Tensile properties of alloys at different temperatures.

the $\text{Al}_3(\text{Sc,Zr})$ particles are fine and dispersed, $3f\gamma/2r > \Delta\rho\mu b^2/2$, and R_c becomes larger. So the particles can impede the recrystallization.

The movement velocity of boundaries is expressed as:⁹⁾

$$V = m[(\rho + \Delta\rho)\mu b^2/2 - 3f\gamma/2r] \quad (4)$$

where m is a constant. When the value of f/r becomes big (for example, for the $\text{Al}_3(\text{Sc,Zr})$ particles), the speed of recrystallization will be slowed.

Furthermore, $\text{Al}_3(\text{Sc,Zr})$ particles have extreme thermostability. Low-temperature annealing (below 623 K) does not affect both the density and the mean size of the particles.⁶⁾ With the temperature increasing, $\text{Al}_3(\text{Sc,Zr})$ particles will be coarsened under the Ostward ripening process like other precipitates. The mean radius r_1 can be expressed as:

$$r_1^3 = r_0^3 + Kt \quad (5)$$

where r_0 is the mean particle radius at the onset of coarsening and K is a constant given by:

$$K = 8DS\sigma\Omega/9kT \quad (6)$$

where D is the solute diffusion coefficient, S the equilibrium concentration, σ the particle/matrix interfacial energy and Ω the atomic volume. The $\text{Al}_3(\text{Sc,Zr})$ particles have a low value of σ since they are coherent with the matrix. As a result, the $\text{Al}_3(\text{Sc,Zr})$ particles obtain a low coarsening rate. It can be shown in Fig. 5 that the $\text{Al}_3(\text{Sc,Zr})$ particles are only 30 ~ 40 nm in diameter after being aged at 723 ~ 773 K. The same phenomenon can be observed in the ternary Al–Sc–Zr alloy.⁴⁾

Therefore, the recrystallization of Alloy A is retarded by

the thermally stable phase $\text{Al}_3(\text{Sc}, \text{Zr})$ through their impeding both grain nucleation and growth. More detail discussions can be found in Ref. 10) regarding the effects of $\text{Al}_3(\text{Sc}, \text{Zr})$ on the tensile properties of Alloy A.

5. Conclusion

The recrystallization temperature of AA2618 added with Sc and Zr is about 200 K higher than that of AA2618. The thermally stable $\text{Al}_3(\text{Sc}, \text{Zr})$ particles can impede the nucleation of grains for recrystallization and retard the grain growth. Therefore, the recrystallization can be suppressed with the addition of Sc and Zr.

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