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Deformation structures and strengthening mechanisms in an Al-Cu alloy subjected to extensive cold rolling

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ABSTRACT

The deformation structures, mechanical properties and strengthening mechanisms of a 2519 aluminum alloy subjected to cold rolling up to a total reduction of 80% ($\varepsilon \sim 1.61$) in the supersaturated solid solution condition were studied. The formation of cell structure and a one hundred-fold increase in lattice dislocation density up to $\sim 2.1 \times 10^{15}\ m^{-2}$ after a 40% reduction leads to increase in yield stress (σ_{YS}) and ultimate tensile strength (σ_{UTS}) from 135 to 453 MPa, from 350 to 503 MPa, respectively. The formation of fully lamellar structure and further increase in lattice dislocation up to $\rho_d \sim 5 \times 10^{15}\ m^{-2}$ take place at an 80% reduction. As a result, σ_{YS} and σ_{UTS} increase to 560 and 590 MPa, respectively. Yield stress at reductions ≥ 40% are significantly higher than that in an AA2519T87 alloy. Subdivision of initial grains by lamellar boundaries due to deformation banding provide high efficiency of dislocation strengthening due to the accumulation of an extremely high density of lattice dislocations in supersaturated solid solution and retention of sufficient elongation-to-failure of 9% and 5% after reductions of 40% and 80%, respectively.

1. Introduction

AA2519-T87 alloy is widely used in the aerospace industry as a structural material in cryogenic tanks and shells, as well as armor plates for combat vehicles, due to its attractive combination of high strength with good ductility, fracture toughness, weldability and satisfactory corrosion resistance [1,2]. Sheets from AA2519 alloy are produced through T87 thermomechanical processing which includes solution heat treatment followed by quenching, cold rolling with a reduction of ~ 7% and subsequent peak aging [1–3]. High strength and sufficient ductility of alloy is provided by a dispersion of θ′-phase nucleated on dislocations [3]. Cold rolling reduction strongly affects ageing response and, therefore, final mechanical properties [3]. A considerable amount of studies address the microstructure evolution of Al alloys during cold rolling and the effect of the deformation structure on mechanical properties of Al [4–6]. To date, there is limited understanding of the microstructural evolution of Al alloys during intense plastic straining in condition of supersaturated solid solution [3,7–13]. It is known that solutes in aluminum alloys inhibit dynamic recovery, retard grain boundary mobility and provide solid solution strengthening. Cu solutes have the greatest effect on these processes because they hinder dislocation climb [8,12–14] and therefore the deformed materials can reach a higher supersaturating density of dislocations. The microstructural evolution in Al-Cu alloy and pure Al during cold rolling is distinctly different. At reductions ≤ 40%, the uniform dislocation distribution evolves in Al-Cu alloys, while in pure Al the well-defined microshear bands aligned along $\{111\}_\alpha$ planes with an inclination angle appear in grains with stable orientations after reductions ≥ 10% [5,8,11,16–18]. The microshear bands are restricted to a grain, i.e. these bands are grain-scale bands. At large strain (reductions ≥ 70%), the formation of full-fledged shear bands crossing an entire sample may occurs [7,8]. As a rule, the formation of microshear bands and shear bands is associated with decreasing the work hardening rate, strong deterioration of ductility and the appearance of mechanical properties anisotropy. Thereby, three sequential processes occur in Al-Cu alloys in the supersaturated solid solution condition during cold rolling [8,18,19]: (i) the accumulation of separate lattice dislocation within interiors of initial grains; (ii) the formation of microshear bands; (iii) the formation of shear bands. It is worth noting that in Al-Cu alloys, deformation banding and continuous dynamic recrystallization take place under equal channel angular pressing (ECAP) at room temperature [9,16].

The aim of the present study is twofold. The first aim is to elucidate the evolution of microstructure during cold rolling in an Al-5.6%wt. Cu alloy under supersaturated solid solution condition. The second aim is to establish a relationship between deformation structures and mechanical properties by analyzing the strengthening mechanisms.

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2. Material and experimental procedure

The AA2519 alloy, with a chemical composition of Al-5.64Cu-0.33Mn-0.23Mg-0.15Zr-0.11Ti-0.09V-0.08Fe-0.08Zn-0.04Sn-0.01Si (in weight %), was manufactured by semi-continuous casting followed by homogenization annealing at 510 °C for 24 h. Next, ingots with dimensions of 100 × 120 × 200 mm^2 were swaged at a temperature of ~ 400 °C (isothermally) in three mutually orthogonal directions up to total strain of ~ 1 imposed using this procedure. Plates of 45 mm width, 250 mm length and thicknesses of 3.0, 3.33, 3.75, 5, 7.5, 10 and 15 mm were machined from a central part of the forged ingot along a prior casting axis. These plates were subjected to solution heat treatment at 535 °C for 1 h and immediately quenched in cold water (~ 20 °C). Next, the plates with different thicknesses were rolled at room temperature with a constant rolling speed of ~ 2 m/min and reductions of 10%, 20%, 40%, 60%, 70% and 80% (logarithmic strain, ε, of 0.11, 0.22, 0.51, 0.92, 1.20 and 1.61, respectively) up to a final thickness of 3 mm. The time before quenching and rolling was ~30 min; therefore, the effect of natural aging on microstructure was insignificant [18].

Specimens for mechanical characterization were machined from the rolled plates. Samples with dimensions of 3 × 10 × 15 mm were cut, grounded, electrolytically polished and applied for Vickers hardness measurements with a Wilson Wolpert 402MVD tester using a load of 2 N and dwell time of 10 s. At least 10 indentations were performed in arbitrarily selected areas for each data point to determine the average values and standard deviations of microhardness for each condition. Flat specimens ("dog bone" type) with a 32 mm gauge length and a 3 × 7 mm^2 cross section were cut with a tensile axis in the longitudinal direction relative to rolling direction. The uniaxial tensile tests to failure were carried out at ambient temperature and an initial strain rate of 1.3 × 10^{-3} s^{-1} using an Instron 5882 testing machine equipped with an automatic high-resolution contacting extensometer MFX 500. Mechanical characteristics were averaged over tests of three samples at each experimental point.

The specimens for structural characterization were cut from the central parts of the rolled plates. Technique of sample preparation for transmission electron microscopy (TEM) and scanning electron microscope has been described in detail in previous works [3,17,18]. Microstructural observations were carried out on the longitudinal plane containing the rolling direction (RD), and normal direction. TEM studies were performed using a JEOL JEM-2100 microscope and a FEI Tecnai G^2 F20 microscope with a double-tilt stage and a field emission gun operated at 200 kV. The both TEMs were equipped with energy-dispersive X-ray (EDX) spectroscopes. Orientation imaging microscopy with automated indexing for Electron Back Scattering Diffraction (EBSD) patterns was applied to examine the microstructural evolution using a high-resolution FEI Nova NanoSEM 450 field-emission-gun with TSL OIM analysis software provided by TexSem Lab Inc. High-angle boundaries (HABs) and low-angle boundaries (LABs) were defined when adjacent pixels in the map exhibit a misorientation of $> 15^\circ$ and $2^\circ < \theta \leq 15^\circ$ and are depicted in misorientation maps using red and white lines, respectively [3,17,18,20]. Areas of 200 × 200 μm^2 were scanned for each material condition. Scan steps of 0.5, 0.2 and 0.1 μm were applied to reconstruct misorientation maps for samples rolled with reductions of 0 ÷ 20, 40 ÷ 70%, and 80%, respectively.

The received misorientation maps were not subjected to any clean-up procedures. The minimal confidence index was set as 0.1. Points with lower indexes were marked in black and were not taken into account. The terms "grains" and "subgrains" are used for the definition of crystallites, which are entirely delimited by HABs and LABs, respectively. The "crystallite size", D_C, is defined as the mean separation of all boundaries (HABs and LABs). The density of appropriate boundaries, γ, was determined by the ratio of the total length of the HABs/LABs and the area of the misorientation map [3,20,21]. For comparative purposes, random misorientation distribution [22] was added as a blue line.

Average lattice dislocation density was estimated by three techniques. First, dislocation density was evaluated from EBSD data from the lattice curvature using the kernel average misorientation option [23,24] via Frank's equation:

$$\theta \approx 2\sin\frac{\theta}{2} = \frac{N\bullet \mathrm{b}}{h}, \tag{1}$$

where θ (in rad) is the misorientation, created by a wall consisting of N dislocations of height h and b is Burgers' vector magnitude. In the EBSD experiments, distance h corresponds to the step size of the scanning. The dislocation density ρ is then given by the ratio of the dislocation number to the surface area:

$$\rho = \frac{N}{S}, \tag{2}$$

where the surface area of the hexagon S as a function of a scanning step h is

$$S = \frac{\sqrt{3}}{2}h^2. \tag{3}$$

It follows that the dislocation density can be estimated from the relationship:

$$\rho_{\mathrm{EBSD}} = \frac{2\sqrt{3}}{3}\frac{\theta}{\mathrm{b}\bullet h}. \tag{4}$$

Misorientations of 0–5° were selected for the dislocation density evaluation by means of the kernel average misorientation.

Second, the dislocation density was estimated by counting the individual dislocations crossing the thin foil surface within the grain/microshear band/shear band interiors in five arbitrarily selected TEM images for each data point using the intercept relationship:

$$\rho_{\mathrm{TEM}} = \frac{1}{t}\left(\frac{n_1}{L_1} + \frac{n_2}{L_2}\right), \tag{5}$$

where n_1 and n_2 are the numbers of intercepts on sets of orthogonal secants of total lengths L_1 and L_2 on the normal image, and t is the foil thickness determined by the convergent beam electron diffraction method using Kossel-Möllenstedt fringes in two-beam condition [20,21,25].

Third, dislocation density was calculated by X-ray technique. X-ray analysis was performed using the Smart Lab Rigaku diffractometer and Cu Kα radiation source. The specimens were scanned in range from 35° to 125° with a step size of 0.02°. Coherent domain size, d_D, and microstrain ($\langle\varepsilon^2\rangle^{1/2}$) were estimated based on the Rietveld technique [26] using MAUD software to accomplish the profile fitting of the experimental X-ray diffraction scan. The dislocation density was calculated using the relationship [26,27]:

$$\rho_{\mathrm{XRD}} = \frac{2\sqrt{3}\langle\varepsilon^2\rangle^{1/2}}{d_D\bullet \mathrm{b}}. \tag{6}$$

3. Results

3.1. Microstructure before deformation

Hot forging produced a partially recrystallized structure with a low portion of HABs and average size of subgrains and (sub)grains of ~ 7 μm and grains of ~ 10 μm (Fig. 1). Areas of grains/(sub)grains alternate with large areas of subgrains. Four types of secondary phase particles were revealed by the TEM technique. The primary T-phase ($Al_{20}Mn_3Cu_2$) with a plate-like shape and average dimensions of 40 × 200 nm^2 were located along HABs (Fig. 1c). Particles of Al_6(Mn,Fe)-phase with a plate-like shape and average dimensions of ~ 20 × 140 nm^2 were occasionally observed within grains. Addition of transition metals, TM, (e.g., Ti and V in case of AA2519) leads to formation of metastable $Al_3(Zr_{1-x}TM_x)$ precipitates with cubic $L1_2$ structure.