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Dynamic recrystallization behavior of 7085 aluminum alloy during hot deformation

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Abstract: The dynamic recrystallization behavior of 7085 aluminum alloy during hot compression at various temperatures (573−723 K) and strain rates (0.01−10 s^{-1}) was studied by electron back scattered diffraction (EBSD), electro-probe microanalyzer (EPMA) and transmission electron microscopy (TEM). It is shown that dynamic recovery is the dominant softening mechanism at high Zener−Hollomon (Z) values, and dynamic recrystallization tends to appear at low Z values. Hot compression with $\ln Z$=24.01 (723 K, 0.01 s^{-1}) gives rise to the highest fraction of recrystallization of 10.2%. EBSD results show that the recrystallized grains are present near the original grain boundaries and exhibit similar orientation to the deformed grain. Strain-induced boundary migration is likely the mechanism for dynamic recrystallization. The low density of Al_3Zr dispersoids near grain boundaries can make contribution to strain-induced boundary migration.

Key words: aluminum alloy; Zener−Hollomon parameter; dynamic recrystallization; strain-induced boundary migration; Al_3Zr dispersoids

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1 Introduction

Dynamic recrystallization (DRX) often occurs in aluminum and aluminum alloys during hot deformation, and generally there are continuous dynamic recrystallization (CDRX) and discontinuous dynamic recrystallization (DDRX) [1,2]. In most aluminum alloys, DDRX was rarely observed because of its high tendency to recover [3,4]; however, some recent investigations have shown that DDRX may occur in Al−Mg, Al−Cu−Li and Al−Zn−Mg−Cu alloys [5−7]. CDRX was reported to occur at high temperatures and low strain rates [5,8,9], and at severe strain geometric dynamic recrystallization (GDRX) tends to occur [7,8]. It is known that particle-stimulated nucleation (PSN), strain-induced boundary migration (SIBM) or subgrain rotation (SGR) is the possible mechanism of DRX. PSN may lead to DDRX because second phase particles can induce a high gradient dislocation density around them during deformation, which may stimulate recrystallization nucleation and growth [10]. SIBM may occur during DRX as well, and the driving force is the difference in the density of dislocations; grain boundaries (GBs) can bulge into the regions with a high density of dislocations [10]. SGR, which often occurs in the interior of original grains, generally results in CDRX [11].

Recrystallization has great influence on the properties of aluminum alloys. For instance, in Al−Zn−Mg−Cu alloys, which are called aeronautical Al alloys [12], the occurrence of recrystallization can decrease strength, toughness and corrosion resistance and increase quench sensitivity. HAN et al [13] reported that the increase of recrystallization fraction leads to lower strength and fracture toughness of 7050 aluminum alloy. KANNAN and RAJA [14] reported that it is possible to

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enhance stress corrosion resistance of Al−Zn−Mg−Cu−Zr alloys by inhibiting recrystallization. LIU et al [15] found that recrystallization results in a larger amount of high angle boundaries and incoherent dispersoids, and consequently increases quench sensitivity of 7055 aluminum alloy. Therefore, it is essential to inhibit recrystallization so as to further improve properties of these alloys. In these aluminum alloys, there are some possible ways to inhibit recrystallization. One is to promote uniform precipitation of fine Al_3M (M=Zr, Sc, etc) dispersoids [16,17], which can hinder the migration of subgrain boundaries and grain boundaries; the second one is to decrease the amount of large second phase particles so as to minimize PSN [18]; the third one is to decrease stored deformation energy, for instance, by elevating temperature slowly during solution heat treatment or by stepped solution heat treating [13,19]. However, during hot deformation, dynamic recrystallization may be triggered due to the increased deformation energy, and this may exert great influence on the static recrystallization during subsequent solution heat treatment. Therefore, it is desirable to have better understanding of the DRX behavior of these alloys, though studies have been done on hot deformation behavior of some Al−Zn−Mg−Cu alloys, such as 7075, 7050 and 7150 alloys [2,20,21].

In this work, dynamic recrystallization behavior of 7085 aluminum alloy was investigated, and this is helpful for controlling recrystallization and obtain combination of high strength and corrosion resistance.

2 Experimental

The material was a 7085 aluminum alloy ingot with chemical composition of Al−7.59Zn−1.65Mg−1.54Cu−0.11Zr (mass fraction, %). The ingot was homogenized at 743 K for 24 h and then cooled in air. Small cylindrical specimens with 10 mm in diameter and 15 mm in height were machined from the homogenized ingot. Uniaxial compression tests were conducted on a Gleeble 3500 thermomechanical simulation unit at temperatures of 573−723 K with strain rates of 0.01−10 s^{-1}. The specimens were heated to the desirable deformation temperature with a heating rate of 3 K/s, held for 2 min and then compressed. All specimens were deformed to a true strain of 0.7 and then quenched in room temperature water immediately to freeze the as-deformed microstructure.

All deformed specimens were sectioned parallel to the compression axis along the centerline for microstructure examination. Electron back scattered diffraction (EBSD) technique was used to examine the microstructure of the specimens so as to obtain information about DRX; this was performed on a ZEISS-EVO18 scanning electron microscope (SEM), and the scanning step size was 1.75 μm and the results were analyzed using HKL Channel 5 software. The distribution of alloying elements in the grains was examined by electro-probe microanalyzer (EPMA). Some specimens were mechanically thinned to a thicknesses of about 80 μm, punched into foils of 3 mm in diameter, electropolished in 30% HNO_3 and 70% CH_3OH solution below −30 °C and then observed on a JEOL−2100F transmission electron microscope (TEM) operated at 200 kV to examine the microstructure.

3 Results and discussion

The true stress−strain curves during hot compression of 7085 aluminum alloy at strain rates of 0.01−10 s^{-1} and at temperatures of 573−723 K are presented in Fig. 1. The peak stress tends to increase with the increase of strain rate or the decrease of temperature, which is similar to previous investigations on Al−Zn−Mg−Cu alloys [2,20,21]. In general, an initial rapid increase in the flow stress can be seen with the increase of strain; however, the shape of the flow curves was changed by strain rate. At low strain rates of 0.1 s^{-1} and 0.01 s^{-1}, the flow stress tends to be constant with the further increase of strain; at 1 s^{-1}, the flow stress increases slightly; at 10 s^{-1}, the flow stress tends to decrease especially at the temperature of 573 K. Moreover, the stress−strain curves are serrated at the rates of 1 s^{-1} and 10 s^{-1}, which may indicate alternate occurrence of dynamic hardening and dynamic softening [5,21]. The fall of flow stress at the highest strain rate of 10 s^{-1} (Fig. 1(a)) shows that dynamic softening and work hardening lose balance, which may be attributed to the adiabatic heating [22].

The softening mechanism may be deduced by microstructure examination. In order to describe the influence of hot deformation parameters on microstructure, a Zener−Hollomon (*Z*) parameter was used, and therefore the effect of both temperature and strain rate can be taken into consideration [1]. The *Z* parameter can be expressed as

$$Z=\dot{\varepsilon}\exp\left(\frac{Q}{RT}\right) \tag{1}$$

where $\dot{\varepsilon}$ is strain rate, *T* is temperature, *R* is the mole gas constant, *Q* is the apparent activation energy. *Q* value was determined to be 172.0 kJ/mol [23], and the value of *Z* can be calculated under various hot deformation conditions.

Orientation imaging maps of the specimens under various hot deformation conditions are given in Fig. 2. It can be seen that grains are elongated after hot deformation, and *Z* values have significant effects on the