Materials Science & Engineering A 648 (2015) 164–177

Contents lists available at ScienceDirect

Materials Science & Engineering A

journal homepage: www.elsevier.com/locate/msea

Static softening following multistage hot deformation of 7150 aluminum alloy: Experiment and modeling

Fulin Jiang [a,b], Hatem S. Zurob [b,*], Gary R. Purdy [b], Hui Zhang [a,c,*]

[a] College of Materials Science and Engineering, Hunan University, Changsha 410082, China
[b] Department of Materials Science and Engineering, McMaster University, 1280 Main Street West, Hamiltion, Ontario, Canada L8S 4L7
[c] Hunan Province Key Laboratory for Spray Deposition Technology and Application, Hunan University, Changsha 410082, China

ARTICLE INFO

Article history:
Received 14 August 2015
Received in revised form
11 September 2015
Accepted 14 September 2015
Available online 18 September 2015

Keywords:
Aluminum alloy
Hot deformation
Static softening mechanism
Modeling
Recovery
Precipitation

ABSTRACT

Previous studies have demonstrated that the static softening kinetics of 7150 aluminum alloy showed typical sigmoidal behavior at 400 °C and softening plateaus at 300 °C (F.L. Jiang, et al., Mater. Sci. Eng. A, vol. 552, 2012, pp. 269–275). In present work, the static softening mechanisms, the microstructural evolution during post-deformation holding was studied by optical microscopy, scanning electron microscope, electron back-scattered diffraction and transmission electron microscopy. It was demonstrated that recrystallization is essentially absent during post-deformation holding, and that static recovery was the main contribution to static softening. Strain induced precipitation and coarsening caused softening plateaus at 300 °C. In order to better understand the static softening mechanism, physically-based modeling, which integrated recovery and multicomponent particle coarsening modeling, was employed to rationalize the experimental results.

1. Introduction

Al–Zn–Mg–Cu super high strength alloys have attracted much attention because of their high strength-to-density ratio, hot workability and excellent comprehensive mechanical properties. They are among the most important structural materials used in the aerospace industry. Although the properties of 7xxx alloys are attractive, the manufacturing processing of these alloys is quite difficult due to their high alloying element content [1,2]. During hot working processes, such as hot rolling, complex microstructural changes take place after each deformation pass [3]. The effect of multi-pass deformation on the microstructure evolution is even less well-understood due to the large number of variables involved [4]. During interpass holding intervals of multistage schedules, static recovery, precipitation and recrystallization may occur. These processes, which strongly depend on the pass strains, temperatures and inter-pass times, have a clearly influence on what follows mechanically and structurally. Hot working stages with suitable holding intervals are managed to combine deformation and annealing textures for enhanced planar anisotropy or for the production of less fibrous grains to avoid delamination corrosion and to promote outstanding mechanical properties [1,5].

Published experimental results indicated that double plateaus appeared on the static softening curves of 7150 aluminum alloy during holding intervals of multistage hot deformation at 300 °C, as shown in our recent literature [6]. Further, typical sigmoidal softening curves are observed at 400 °C. Sigmoidal static softening curves are commonly observed in steels; these curves are usually interpreted as being largely due to static recrystallization, with a minor softening contribution (about 20–30%) due to concurrent static recovery [2,7]. Static softening plateaus have also been widely studied and attributed to coupled softening from precipitation, recrystallization, recovery as well as the reduction in solid-solution hardening due to precipitation in microalloyed steels [7–10], and in Mg alloys as well [11]. A double plateau for steel was reported by Medina et al. [7] when the temperature was equal to or less than 1050 °C and reconfirmed in several subsequent articles [8–10]. Several researchers have presented explanations for softening kinetics and the formation of these plateaus in steels. Medina et al. [7,8] suggested that the formation of the "double hump" is due to the appearance of two types of precipitates. Once the kinetics of the first precipitation (first plateau) ended and recrystallization continued, the second precipitation began, and once this was completed, recrystallization started again. Zurob et al. [9] presented explanations for three possible softening vs. time curve shapes, depending on the evolution of the stored dislocation energy and precipitation hardening which give rise to the softening and hardening contributions that occasionally result in approximately constant hardness or a plateau on the

* Corresponding authors.
E-mail addresses: zurobh@mcmaster.ca (H.S. Zurob), zhanghui63hunu@163.com (H. Zhang).

http://dx.doi.org/10.1016/j.msea.2015.09.063
0921-5093/

softening curve. Gómez et al. [10] also worked on the influence of (Al, Nb, V) precipitates on the recrystallization inhibition in microalloyed steels and found that the duration of plateaus followed the order: Nb > V > Al. Maghsoudi et al. [11] suggested that the effect of precipitation of $Mg_{17}Al_{12}$ particles on the grain boundary migration was the main reason for the observed plateau in AZ61 magnesium alloy. Physically-based models for static softening, which integrated the complex interactions between recovery, recrystallization and precipitation during the hot-deformation of microalloyed steels, have been developed by Zurob et al. [9,12,13].

Due to the high-stacking-fault-energy of aluminum and high alloying element content in present alloy, which, respectively, lead to high level of recovery and complex precipitation reactions, the static softening mechanism of Al–Zn–Mg–Cu alloys will likely be different from that of microalloyed steels [1,2]. Thus a clear understanding of the softening mechanisms and interactions of precipitation, recrystallization and recovery requires further investigation for aluminum alloys. It is also well-accepted that modeling the microstructural evolution of aluminum alloys during thermomechanical processing is highly desirable in order to predict product properties and/or to design process variables based on requirements for the properties [14]. To do so, having sound physically-based models is of interest for both academic research and industrial practice. In this work, microstructural evolution during post-deformation holding was studied by optical microscopy (OM), scanning electron microscopy (SEM), electron backscattered diffraction (EBSD) and transmission electron microscopy (TEM) to explore the static softening mechanism. Physically-based modeling, which integrated recovery and multicomponent particle coarsening modeling, was developed in order to rationalize the experimental static softening results and to better understand the underlying mechanisms.

2. Experimental procedure

The experiments were carried out on a commercial 7150 aluminum alloy containing 6.38Zn, 2.32Mg, 2.11Cu, 0.09Zr, 0.06Si, 0.08Fe, 0.053Ti (mass, %). The semi-continuous chill cast ingot was homogenized at 465 °C for 24 h followed by air cooling. The homogenized ingot was machined to make cylindrical compression samples with length 15 mm and diameter 10 mm. The flat ends of the specimen were recessed to a depth of 0.2 mm deep in order to entrap the lubricant of graphite mixed with machine oil during deformation so that friction at the specimen/die interface would be minimized. The samples were heated to deformation temperature at a heating rate of 10 °C/s and held at that temperature for 3 min by feedback-controlled AC current before compression. Then the samples were subjected to first-pass deformation, using a strain of 0.4 at strain rates of 0.01 s^{-1} and 0.1 s^{-1} and temperatures of 300 °C and 400 °C, and subsequent isothermal holding for different times (i.e. 0, 1, 10, 100, 250, 500 and 1000 s). The isothermal holding stage was followed by either second-pass reloading, performed at the same strain rate to a total strain of 0.8, or immediate water quenching to preserve the high temperature microstructure. Microstructural examination was performed in the central regions of the thermo-mechanically processed samples on a plane containing the compression axis using optical microscopy (OM), scanning electron microscope (SEM), electron back-scattered diffraction (EBSD) and transmission electron microscopy (TEM). The samples were prepared by conventional grinding and polishing methods and then examined using an Axiovert-40 metallographic microscope, an FEI QUANTA200 environmental SEM and a JEOL JSM-7000F equipped with EBSD. TEM observations were carried out on JEOL JEM-3010 transmission electron microscope operating at 200 kV after electropolishing samples in a solution of 30% HNO_3 and 70% methanol at 25 V and at −30 °C. Microstructure quantification was carried out by Image-Pro Plus software. Numerical modeling studies utilized Thermo-Calc and Maple software.

3. Static softening mechanisms

3.1. Microstructure of as-received 7150 AL alloy

Fig. 1(a) shows the OM of the as-received and homogenized 7150 aluminum alloy. Equiaxed grains are observed. The average grain size is approximately 69 μm. The strong contrast at the grain-boundaries is due to preferentially etched coarse particles which are present along grain boundaries. The coarse particles along boundaries might be undissolved intermetallic particles, or the result of precipitates that formed preferentially along the boundaries during cooling [15,16]; they are also imaged by SEM in Fig. 1(b). In addition, a large number of coarse constituent particles with size below 1 μm are observed in Fig. 1(b). These constituent particles are probably formed by precipitation during cooling after homogenization. Due to the quenching sensitivity of current alloy, precipitation nucleation, growth or coarsen process rapidly during cooling [16]. The homogenized ingot used in this work was cooled in the air with a relative slow cooling rate, which could explain the coarse particles appearing in Fig. 1(b). In addition, a few sparse fine rounded precipitates were also observed by TEM (Fig. 1(c)). Separate coarse rodlike particles are also observed by TEM in Fig. 1 (d), which likely corresponds to the particles seen in Fig. 1(b). Those particles in Fig. 1(b)–(d) are η ($MgZn_2$), T ($Al_2Mg_3Zn_3$), S (Al_2CuMg), Al_7Cu_2Fe and Mg_2Si second phases which are usually present in Al–Zn–Mg–Cu alloys according to EDS identification and previous works [15–17].

3.2. Microstructure analysis during post-deformation

Fig. 2(a)–(d) presents the selected micrographs of deformed 7150 alloy samples with second-pass reloading (total strain of 0.8). Elongated grains are shown in all deformed samples, even with 1000 s duration. There are no clear fine (recrystallized) grains as well. Such characteristics correspond to a typical recovery microstructure. In order to get better sense of the grain evolution, quantitative metallographic studies were carried out according to the methodology introduced by Orsetti Rossi and Sellars [18]. Fig. 2(e) shows the final quantitative metallography results of deformed 7150 alloy samples ($\varepsilon = 0.8$) under various durations (of holding time, t). Compared with the average grain size of as-received alloy (original), little decrease of grain size is shown for the deformed samples. It seems at first glance that the average grain size decreases gradually with increasing holding time. However, this variation is not significant when considering the statistical errors. Therefore, the grains were simply elongated during the various double stage deformation stages; this implies the absence of recrystallization.

Fig. 3(a) and (b) show the high-angle and low-angle (sub)grain boundaries of the sample with 1000 s isothermal holding after first-pass deformation ($\varepsilon = 0.4$) at 400 °C and 0.1 s^{-1} as assessed by EBSD. The figures reveal that the sample contains large elongated grains containing some internal low-angle boundaries, but bounded predominantly by high-angle grain boundaries. Most of the low-angle grain boundaries terminate at the original high-angle grain boundaries which could be seen in OM as well (Fig. 3 (d)). The average grain size (high-angle) is approximate those of the original grains (69 μm, Fig. 1(a)). The corresponding misorientation distribution was plotted in Fig. 3(c) utilizing the Channel 5 software. A high percentage of low-angle boundaries