Materials and Design 107 (2016) 277–289

Contents lists available at ScienceDirect

Materials and Design

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

Physically based constitutive analysis and microstructural evolution of AA7050 aluminum alloy during hot compression

S. Wang [a], J.R. Luo [a], L.G. Hou [a,*], J.S. Zhang [a], L.Z. Zhuang [a,b,**]

[a] State Key Laboratory for Advanced Metals and Materials, University of Science and Technology Beijing, Beijing 100083, PR China
[b] Tata Steel, 1970 CA IJmuiden, Netherlands

ARTICLE INFO

Article history:
Received 20 April 2016
Received in revised form 29 May 2016
Accepted 7 June 2016
Available online 8 June 2016

Keywords:
7050 aluminum alloy
Constitutive model
Dynamic recrystallization
Grain boundary sliding
Dislocation creep

ABSTRACT

The hot compression tests of AA7050 aluminum alloy were conducted under conditions of 603–693 K and 0.001–10 s^{-1}, and the related microstructures were observed. Physically based constitutive analysis was conducted to describe the flow behaviors, which can relate the microstructural evolution with flow behaviors for high stacking fault energy (SFE) and/or precipitation-strengthened alloys. A revised model considering the coupling effects of lattice diffusion and grain boundary diffusion was proposed to characterize the transition of diffusion mechanisms under different deformation conditions. The main diffusion mechanism is determined as lattice diffusion at 633–693 K and grain boundary diffusion at 603 K. The microstructural evolution can be reflected by the deviation of creep exponent n' from the theoretical value ($n'=5$). The reasons for the creep exponent $n'>5$ could be related to the change of internal stress and creep rate by dynamic precipitates at lower temperatures. At higher strain rates, it could be related to the impediment of dislocations motion by defects and the change of rate controlling mechanism. The operation of grain boundary sliding (GBS) may lead to $n'<5$ at higher temperatures and lower strain rates. Moreover, the mechanisms of dynamic recrystallization under wide conditions and high-strain-rate superplasticity were discussed.

1. Introduction

The Al-Zn-Mg-Cu aluminum alloys have been widely used for structural applications in automobile and aerospace industries due to their high strength-to-density ratio, high toughness and good corrosion resistance. The good mechanical properties and desired microstructural characteristics can be resulted from appropriate deformation conditions, such as temperature, strain rate and degree of deformation. The influence of these deformation parameters on microstructural evolution can be characterized by relating the dynamic recrystallization (DRX) with flow behaviors in the form of constitutive analyses for low stacking fault energy (SFE) materials [1] or by relating power dissipation and flow instability with flow behaviors in the form of processing maps [2]. Therefore, a deeper understanding of flow behaviors can play an important role in optimizing the design of metal-forming processes and controlling the microstructural evolution during hot working.

As summarized in Table 1, two types of constitutive models are often used to describe the flow behaviors [3]: (1) phenomenological constitutive models; (2) physically based constitutive models. The Arrhenius-type model [4] and Johnson-Cook model [5], as two widely used phenomenological models, have been successfully applied to predict the flow behaviors of many materials [6,7]. However, the materials constants in these models have less metallurgical meaning and cannot reflect the microstructural evolution. As a physically based model, the Zerilli-Armstrong model [8] has been widely employed to describe the dislocation mechanism for BCC and FCC metals. In this type model, the coupling effects of strain-hardening, strain-rate hardening and thermal softening on the flow behaviors are considered [9], however less information on microstructural evolution can be obtained. For the Estrin and Mecking [10] and Bergström [11] model, another widely-used and physically-based model, its greatest advantage is to relate the flow stress with the volume fraction of DRX. Essentially, this model considering the influence of the work-hardening and dynamic softening on flow stress is based on the variation of dislocation density [12]. However, the hypothesis that the flow stress is only influenced by the work-hardening and dynamic restoration (*i.e.* dynamic recovery (DRV) and/or DRX) limits its application in some alloys, such as precipitation-strengthened alloys. In addition, the DRX volume fraction equation used in this model is based on the Avrami equation which is suitable for discontinuous DRX (DDRX, in the form of nucleation and growth). For high SFE metals/alloys (*e.g.* aluminum alloys), the main DRX mechanism during hot deformation is continuous DRX (CDRX), which occurs by progressive subgrain rotation without grain boundary migration

* Corresponding author.
** Correspondence to: L.Z. Zhuang, Tata Steel, 1970 CA IJmuiden, Netherlands.
E-mail addresses: lghou@skl.ustb.edu.cn (L.G. Hou), linzhongzhuang@yahoo.com (L.Z. Zhuang).

http://dx.doi.org/10.1016/j.matdes.2016.06.023
0264-1275/

Table 1
The advantages and disadvantages of four widely-used constitutive models.

Model types	Model descriptions	Advantages and/or disadvantages
Arrhenius-type [6]	$\sigma = \frac{1}{\alpha}\,,\ln\{(\dot{\varepsilon}Q/ART)^{1/n} + [(\dot{\varepsilon}Q/ART)^{2/n}+1]^{0.5}\}$	These three models can be applied for many materials successfully with simple forms, but they cannot reflect the microstructural evolution during deformation.
Johnson-Cook [7]	$\sigma = (A + B\varepsilon^n)(1 + Cln(\dot{\varepsilon}/\dot{\varepsilon}_r))[1-(T-T_r)/(T_m-T_r)]$	
Zerilli-Armstrong [9]	$\sigma = C_0 + C_2\varepsilon^{0.5}\,,\exp(-C_3T + C_4T\,,\ln\dot{\varepsilon})$ for FCC	
	$\sigma = C_0 + C_1\,,\exp(-C_3T + C_4T\,,\ln\dot{\varepsilon}) + C_5\varepsilon^n$ for BCC	
Estrin and Mecking and Bergström [12]	$\sigma=\sigma_{rec}=[\sigma_{sat}^2+(\sigma_0^2-\sigma_{sat}^2)e^{-\Omega\varepsilon}]^{0.5}$ $\varepsilon\le\varepsilon_c$	It can be used to relate the DRX with flow behaviors for low SFE metals/alloys, but it may be unsuitable for high SFE and/or precipitation-strengthened metals/alloys.
	$\sigma=\sigma_{rec}-(\sigma_{sat}-\sigma_{ss})X_{drx}$ $\varepsilon\ge\varepsilon_cX_{drx}=1-\exp[-k_d((\varepsilon-\varepsilon_c)/\varepsilon_p)^{n_d}]$	

(The definition of materials constants in these models can be referred to the corresponding references.)

(GBM) [13,14]. Thus, this model can reflect microstructural evolution during deformation for low SFE materials, but it may be unsuitable for metals/alloys with high SFE such as aluminum alloys.

In recent years, the hot deformation behaviors of Al-Zn-Mg-Cu alloys have been extensively studied, including constitutive analysis [15], hot workability [16], dynamic precipitates [17] and DRX mechanisms [18], *etc*. However, less researches were focused on relating the microstructural evolution with constitutive analysis for such high SFE metals/alloys and/or precipitation-strengthened alloys. So it is useful and necessary to construct a physically constitutive model which can reflect the microstructural evolution for metals/alloys such as Al-Zn-Mg-Cu alloys. Cabrera et al. [19–22] proposed a physically based constitutive model with a creep exponent ($n = 5$) to describe the flow behaviors successfully as long as the deformation mechanism is controlled by dislocation glide and climb. In this model, the Young's modulus (E) and the self-diffusion coefficient (D) are taken as function of temperatures (T), and the relationship can be expressed as:

$$\dot{\varepsilon}/D(T) = B[\sinh(\alpha\sigma/E(T))]^5 \tag{1}$$

where $\dot{\varepsilon}$ and σ are the strain rate (s^{-1}) and flow stress (MPa), respectively. B and α are materials constants, and 5 represents the theoretical value of the creep exponent n. In this study, a revised model based on Eq. (1) was constructed in an attempt to determine the main diffusion mechanisms (grain boundary diffusion or lattice diffusion) under different deformation temperatures. Simultaneously, the microstructural evolution during hot deformation of AA7050 aluminum alloy was characterized and related with the variation of the creep exponent.

2. Experimental materials and procedures

The compression specimens with 10 mm in diameter and 15 mm in height were machined from commercial AA7050 aluminum alloy plate according to ASTM: E209. Before compression tests, all samples were treated at 748 K for 2 h, and then quenched into room-temperature water immediately. Hot compression tests were conducted on a Gleeble-3500 thermo-simulator at temperature range of 633–693 K and strain rate range of 0.001–10 s^{-1}. Before compression, each specimen was heated to the preset temperature with a heating rate of 2 K/s and held at that temperature for 5 min to minimize thermal gradients. Thin graphite sheets were used to reduce frictions, and all the deformed specimens after compression were quenched into room-temperature water immediately for microstructural observation. The microstructures of some samples were observed on the center of the axial section by scanning electron microscopy (SEM), transmission electron microscopy (TEM) and electron backscattered diffraction (EBSD) technique. The SEM samples were firstly mechanical-polished and then etched with the Keller solution. TEM samples were prepared by twin-jet electro-polishing using the solution of HNO_3 and methanol (1:3 in volume) and conducted on a Tecnai G2 F30 S-TWIN TEM. The EBSD samples were electro-polished in a solution of 5% perchloric acid and 95% ethanol at 30 V for 20 s. The EBSD data were analyzed through HKL Channel 5 software. The low angle grain boundaries (LAGBs, grain boundary orientation angle: 2°–15°) were marked by thin red lines for 2°–5° and thin fuchsia lines for 5°–15°, and the high angle grain boundaries (HAGBs, grain boundary orientation angle >15°) were marked by thick black lines in all the EBSD restructured maps. The Kernel average misorientation (KAM) maps represent the local misorientation which means an average misorientation of a point with all of its neighboring points in a grain. The average misorientation of a point was calculated with a provision that misorientation exceeding some tolerance value (5°) are excluded.

3. Results and discussion

3.1. Flow behaviors

The flow stress curves of AA7050 aluminum alloy under the temperature range of 603–693 K and the strain rate range of 0.001–10 s^{-1} are gained from Ref. [13] and displayed in Fig. 1. Generally, the flow stresses increase with increasing strain rates or decreasing deformation temperatures. The flow stress curves of aluminum alloy show an obvious difference to that of low SFE materials such as austenitic steel [13]. At lower deformation temperature (603 K and 0.001 s^{-1} in Fig. 1(a)), the DRV may be the main softening mechanism. However, the flow stresses increase to peak stress and then decrease to a steady stress continuously, which is similar to the flow curves of low SFE materials undergoing DRX. In the case of DRX (*e.g.* 693 K and 0.001 s^{-1} in Fig. 1(a)), the flow stress curve keeps constant after reaching peak stress, which is the typical feature of low SFE materials under condition of DRV. With microstructural observation, the DRX is easy to occur at higher temperatures and low strain rates due to higher rate and more times for dislocation motion. In addition, at higher strain rates (>1 s^{-1}), the deformation heating can be produced and promote the occurrence of DRV and/or DRX [23]. So the flow curves show dynamic softening at lower and higher strain rates. At intermediate strain rates (*e.g.* 0.1 s^{-1} in Fig. 1(c) and 1 s^{-1} in Fig. 1(d)), the flow stresses increase with increasing strains (from 0.35 to 0.65 for 0.1 s^{-1} and from 0.1 to 0.65 for 1 s^{-1}), which may be related to the combined effects of less DRV and/or DRX and the Orowan strengthening mechanisms caused by dynamic precipitates [13].

3.2. Original physically based model with considering lattice diffusion

The activation energy represents the level of an energy barrier to be surmounted in some atomistic mechanisms such as diffusion, deformation, microstructures and so on [24]. Based on the result of Arrhenius model analysis [13], the apparent hot working activation energy, which assumed that the microstructure remained constant and ignored the underlying atomic mechanisms during hot deformation, was calculated as 200 KJ mol^{-1} for AA7050 aluminum alloy. This value is higher than the self-diffusion activation energy of aluminum, no matter the activation energy of lattice diffusion (142 KJ mol^{-1}) or grain boundary diffusion (84 KJ mol^{-1}) [25]. Some researchers have pointed out that the variation of Young's modulus with temperatures may be one of the reasons caused the deviation of activation energy from self-diffusion activation energy and the creep exponent n from the theoretical value