International Journal of Plasticity 54 (2014) 178–192

Contents lists available at ScienceDirect

International Journal of Plasticity

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

Characterization of stress–strain relationships in Al over a wide range of testing temperatures

Tamás Csanádi [a], Nguyen Q. Chinh [a,*], Jenő Gubicza [a], György Vörös [a], Terence G. Langdon [b,c]

[a] Department of Materials Physics, Eötvös University Budapest, H-1117 Budapest, Pázmány P. sétány 1/A., Hungary
[b] Departments of Aerospace & Mechanical Engineering and Materials Science, University of Southern California, Los Angeles, CA 90089-1453, USA
[c] Materials Research Group, Faculty of Engineering and the Environment, University of Southampton, Southampton SO17 1BJ, UK

ARTICLE INFO

Article history:
Received 18 April 2013
Received in final revised form 16 August 2013
Available online 4 September 2013

Keywords:
A. Dislocations
A. Microstructures
A. Strengthening mechanisms
B. Constitutive behavior
B. Stress–strain relationship

ABSTRACT

The stress–strain relationships characterizing plastic deformation of aluminum are described over a wide range of testing temperatures by applying both a widely used phenomenological relationship and a well-known dislocation-based model. It is shown that over the whole range of testing temperatures the trapping of mobile dislocations and the annihilation of forest dislocations are controlled by the same thermally-activated dislocation motion, thereby leading to a simplified model which uses only two parameters to describe the multiplication and the annihilation rates of dislocations. The temperature dependence of these two microscopic quantities is explained. Furthermore, correlations between the characteristics of macroscopic and microscopic descriptions were established over a wide range of testing temperatures for pure Al.

1. Introduction

For more than fifty years, considerable interest has been devoted to understanding the plastic behavior of simple face-centered cubic (fcc) metals such as pure aluminum. Thirty years ago, Frost and Ashby (1982) summarized both the theoretical models and experimental data in the form of sets of deformation mechanism maps covering a very wide range of crystalline materials including pure aluminum. These maps divided the flow behavior into a regime of diffusion-controlled creep dominating flow at high temperatures and a regime of thermally-activated flow dominating at low temperatures, thereby suggesting a transition from low to high temperature behavior in the vicinity of $\sim 0.5T_m$, where T_m is the absolute melting point of the material.

On the basis of many different experiments, it is now well established that the work hardening of single crystals may be divided into three distinct stages, designated as I, II and III (Nabarro et al., 1964; Berner and Kronmüller, 1965; Kovács and Zsoldos, 1973; Kocks and Mecking, 2003). Investigations using torsion experiments (Zehetbauer and Seumer, 1993; Zehetbauer, 1993; Les et al., 1997) have suggested the occurrence of two additional stages, denoted as IV and V in the case of polycrystalline metallic materials. Later, applying severe plastic deformation (SPD) techniques, which are favored methods for producing bulk ultrafine-grained materials (Valiev et al., 2000; Valiev and Langdon, 2006; Zhilyaev and Langdon, 2008), it was found that the flow stress tends practically to saturate (Chinh et al., 2004, 2005, 2010; Csanádi et al., 2011) in accordance with the behavior characteristic for stage V. Consequently, it is now possible to investigate the flow process over a wide range of strain at different temperatures and this is important from both a materials science and an engineering point of view.

* Corresponding author. Tel.: +36 1 3722845.
E-mail address: chinh@metal.elte.hu (N.Q. Chinh).

0749-6419/$ - see front matter
http://dx.doi.org/10.1016/j.ijplas.2013.08.014

According to the many experimental data, several equations were developed for the description of the stress–strain ($\sigma - \varepsilon$) relationships. There are both phenomenological (Hollomon, 1945; Voce, 1948; Chinh et al., 2004, 2005; Farrokh and Khan, 2009; Csanádi et al., 2011) and dislocation-based (Kocks, 1976; Estrin and Mecking, 1984; Kubin and Estrin, 1990; Malygin, 1990; Lukác and Balík, 1994; Estrin et al., 1998; Nes, 1998; Barlat et al., 2002; Tóth et al., 2002; Beyerlein and Tomé, 2008; Austin and McDowell, 2011; Fan and Yang, 2011; Gao and Zhang, 2012; Bertin et al., 2013; Hansen et al., 2013; Li et al., 2013) models. In the case of the latter, different microstructural approaches were suggested, including the evolution of the total dislocation density (Kocks, 1976; Estrin and Mecking, 1984; Malygin, 1990; Lukác and Balík, 1994; Beyerlein and Tomé, 2008, Fan and Yang, 2011), the separate evolution of the mobile and forest dislocations (Kubin and Estrin, 1990; Barlat et al., 2002; Wang et al., 2009; Austin and McDowell, 2011; Gao and Zhang, 2012; Hansen et al., 2013; Li et al., 2013), the separate evolution of the polar and non-polar dislocations (Bertin et al., 2013), the development of the dislocation densities in cell walls and grain interiors (Estrin et al., 1998; Roters et al., 2000; Tóth et al., 2002) and/or the subgrain size and misorientation angle (Nes, 1998) as structural parameters. Recently, the plastic deformation of polycrystals was described also by considering the grain boundary effects (Lim et al., 2011) and the grain size gradient (Li and Soh, 2012) in the case of nanostructured materials. As all of the macroscopic and microscopic models reasonably describe the experimental data, despite their different approaches, there should be a correlation between the main characteristics of these various models. A knowledge of any possible connections would certainly help in achieving a deeper understanding of the mechanisms of plastic deformation, and also it would contribute towards an understanding of the physical meanings of the parameters used in the different models.

This was the motivation for the present work and for the earlier studies reported recently (Chinh et al., 2010; Csanádi et al., 2011). Accordingly, this work may be regarded as a continuation of these recent reports where the room temperature plastic behavior of pure Al and several other fcc metals was studied over a wide range of strain. In these earlier reports (Chinh et al., 2010; Csanádi et al., 2011) the relationships were examined between the parameters of the microscopic processes adopted in the classic Kubin–Estrin (KE) model (1990) and the characteristics of a recent macroscopic description (Chinh et al., 2004). Based on this earlier analysis, the present work is focused on the plastic behavior of pure Al deformed at different testing temperatures. Specifically, the effects of thermal activation are investigated in terms of the temperature dependence of the relevant characteristics of the macroscopic (Chinh et al., 2005) and microscopic KE (Kubin and Estrin, 1990) descriptions. In order to place this report in perspective, the following section provides a brief description of the background to the analysis.

2. Background to the analysis: macroscopic description of the plastic deformation in Al

In an earlier work (Chinh et al., 2005) high purity (99.99%) aluminum samples were deformed by tensile tests over a wide range of temperature between 293 K (room temperature) and 738 K. The specimens were annealed for 30 min. at 673 K to give an initial grain size of ~190 μm and then tested using an MTS testing machine operating at a constant cross-head velocity with initial strain rate of $1.0 \times 10^{-3} s^{-1}$. Additional data were also used in this earlier analysis based on information reported previously for the same material where samples were produced by processing through equal-channel angular pressing (ECAP) (Iwahashi et al., 1998a; Komura et al., 1999). For this purpose, aluminum billets having diameters of 10 mm and lengths of 60 mm were annealed and pressed through a channel at room temperature using a solid die with an angle of 90° between the two parts of the channel and an angle of ~30° representing the outer arc of curvature where the two parts of the channel intersect. These internal angles lead to an imposed strain of ~1 on each passage through the die. Further details on the principles of processing by ECAP are given in earlier reports (Iwahashi et al., 1996, 1997, 1998b; Furukawa et al., 2001; Valiev and Langdon, 2006) and it is important to note that processing by ECAP provides the opportunity to introduce large strains into the material.

It was demonstrated that the macroscopic stress–strain ($\sigma - \varepsilon$) relationship may be fitted by a phenomenological constitutive relationship of the form (Chinh et al., 2004, 2005):

$$\sigma = \sigma_0 + \sigma_1 \left\{ 1 - \left[\exp - \left(\frac{\varepsilon}{\varepsilon_c} \right)^n \right] \right\} \tag{1}$$

where σ_o, σ_1, ε_c and the exponent n are fitting parameters and the strain ε is taken as the absolute amount of plastic strain relative to the annealed state. Physically, the first constant σ_o, is the friction stress related to the onset of plastic deformation which is then described by the three fitting parameters, σ_1, ε_c and n. Fig. 1 shows, for example, the experimental $\sigma - \varepsilon$ data obtained at room temperature and a fitted line based on Eq. (1) for two separate situations: very small initial strains as achieved in conventional tensile testing in Fig. 1(a) and a very wide range of strain including the high strains imposed in ECAP in Fig. 1(b) where the datum points represent the individual measured yield stresses. In practice, the flow stress of the specimens subjected to ECAP is documented by reading the 0.2% proof stress, $\sigma_{0.2}$ obtained in the tensile test. This approach is given in Fig. 1(b) where the strain is defined as the sum of that resulting from the tensile test and the imposed strain due to ECAP processing so that a zero strain corresponds to the initial unpressed and annealed condition. The results shown in Fig. 1 confirm the validity of the constitutive relationship given by Eq. (1) and the data in Fig. 1(b) demonstrate a smooth transition from the low strains attained in tensile testing to the high strains imposed through severe plastic deformation. An examination of Eq. (1) shows that, as the flow stress, σ, tends to a saturation value