Materials Science and Engineering A 393 (2005) 157–163

MATERIALS SCIENCE & ENGINEERING A

www.elsevier.com/locate/msea

Controlling mechanisms of deformation of AA5052 aluminium alloy at small strains under hot working conditions

R. Carmona*, Q. Zhu[1], C.M. Sellars, J.H. Beynon

IMMPETUS—Institute for Microstructural and Mechanical Process Engineering: The University of Sheffield, Mappin Street, Sheffield S13JD, UK

Received 5 April 2004; received in revised form 4 October 2004; accepted 2 November 2004

Abstract

The objective of the present work was to investigate the thermomechanical behaviour of a commercial Al–Mg alloy and to understand the deformation mechanisms taking place at small strains (<0.2) and high temperatures (≥300 °C) at strain rates relevant to hot working conditions (0.01–10 s^{-1}). The experimental approach addresses different effects of testing machines on the load–displacement measurements. The results show distinct sorts of behaviour depending on the Zener–Hollomon parameter (Z): normal continuous increase in the stress with strain for high values of Z and a rise followed by a drop in stress with increasing strain for low values of Z. For steady-state behaviour, the relationship between stress and strain rate follows the typical power law with an exponent of 3 at low values of Z and an exponential relationship at higher values. These observations are similar to those of creep. The difference in stress response may be explained by different controlling mechanisms, namely climb of dislocations for the normal hardening behaviour and solute drag in the case where a drop of stress is observed in the stress–strain curve.

Keywords: Hot working; Small strain deformation; Aluminium alloys; Flow stress; Deformation mechanism; Zener–Hollomon parameter

1. Introduction

To date, interest in thermomechanical processing has been mainly focused on large strain plastic deformation. Recently, however, the small strain behaviour has received attention, because some regions in the hot formed workpiece may undergo very little deformation during certain stages of processing [1]. Such is the case for large section rolling or forging, where localised regions of the sections may receive very little or no deformation in certain passes. As a result, the mechanical properties in these localised regions of products may be poor, because the deformation may not be enough to trigger recrystallisation. Mechanical equations of state are available for aluminium alloys for large strain plasticity [2] but these equations cannot be extrapolated with confidence to small plastic strains. The processes undergone by the material when the strains are small have not yet been addressed and there is a knowledge gap in this area of study of high temperature deformation at hot working strain rates.

Creep studies of Al–Mg alloys show that, under some deformation conditions, this type of alloy presents special features, such as an exponent of 3 in the stress–strain rate relationship instead of 5 usually found for pure Al or other Al alloys [3,4]. Yavari and Langdon [3] and Zhu et al. [4] carried out their tests not only under traditional creep conditions (constant stress and low strain rate), but also under constant cross-head displacement rate (close to constant strain rate). Al–Mg alloys showed a transition between the exponents 3 and 5, which was interpreted as a breakaway of the dislocations from their solute atmospheres, with a changeover in

* Corresponding author. Present address: EADS-CASA, Ensayos Estructurales, Paseo John Lennon, s/n, E-28906 Getafe, Madrid, Spain
Tel.: +34 659 68 5627.
E-mail address: rcarmona@coiim.es (R. Carmona).
[1] Present address: Holset Engineering, St. Andrew's Road, Huddersfield HD16RA, UK

0921-5093/$ – see front matter © 2004 Elsevier B.V. All rights reserved.
doi:10.1016/j.msea.2004.11.010

Nomenclature

$\boldsymbol{b}$	Burgers vector
E	Young's modulus
D	diffusion coefficient: $D = D_{\mathrm{o}} \exp\left(\frac{-Q_{\mathrm{diff}}}{RT}\right)$
D_{o}	frequency factor
f_{v}	volume fraction of second phase particles
G	shear modulus
k_{B}	Boltzmann's constant: 1.381×10^{-23} J/K
M	Taylor factor (≅3 for fcc polycrystals)
n	exponent relating strain rate and stress
n'	exponent relating velocity of dislocations and effective stress
Q_{def}, Q_{diff}	activation energies for deformation and diffusion, respectively
R	universal gas constant: 8.314 J/(mol K)
r	second phase particle radius
T	absolute temperature
W_{m}	binding energy between solute atoms and dislocations
Z	Zener–Hollomon parameter $Z = \dot{\varepsilon} \exp\left(\frac{Q_{\mathrm{def}}}{RT}\right)$
δ	subgrain size
ε	strain
$\dot{\varepsilon}$	strain rate
ρ_{i}	internal dislocation density
θ	subgrain misorientation
ν	velocity of moving dislocations
σ	equivalent tensile stress
α, β, C	material constants

the acting mechanisms for the control of dislocation motion, from viscous drag controlled glide to high temperature climb or cross-slip. However, that investigation was focused only on the steady-state behaviour. In fact, much earlier, Jonas et al. [5] reviewed the whole spectrum of conditions in creep and hot working, and established the interdependence of stress and strain rate by a hyperbolic sine function, which becomes a power law at lower stresses and an exponential law at higher stresses.

It is clear that there is a need to understand fully the whole deformation behaviour, including transient stress–strain curves, and the microstructural processes taking place at small strains in thermomechanical processing. Current knowledge of creep or large plastic deformation under hot working conditions does not give sufficient insight into the deformation behaviour at low strain. In the present study, the deformation behaviour at strains up to about 0.15 for Al–Mg alloys under hot working conditions is investigated in detail. Aluminium alloy AA5052 was used because it represents a group of aluminium alloys widely used in the transportation, marine and automotive industries. The controlling mechanisms of hot deformation behaviour are discussed using internal state variables, i.e. the internal dislocation density (ρ_{I}) the subgrain size (δ) and the misorientation between subgrains (θ).

2. Description of the stress as a function of the internal state variables

The flow stress of the material (σ) can be described as [6]:

$$\sigma = \sigma_{\mathrm{e}} + \sigma_{\rho_{\mathrm{i}}} + \sigma_{\delta} + \sigma_{\mathrm{d}} + \sigma_{\mathrm{p}} \tag{1}$$

where σ_{e} is the effective stress (also known as friction stresses, σ_{f}) $\sigma_{\rho_{\mathrm{i}}}$ the athermal stress due to interactions of dislocations within the subgrains, σ_{δ} and σ_{d} are the long range athermal stresses from the subgrain boundaries and from the grain boundaries, respectively, and σ_{p} is the particle hardening term.

The athermal stress arising from interaction of internal dislocations can be expressed as [7]:

$$\sigma_{\rho_{\mathrm{i}}} = \alpha M G \boldsymbol{b} \sqrt{\rho_{\mathrm{i}}}, \tag{2}$$

where α is a material parameter, M the Taylor factor, G the shear modulus, $\boldsymbol{b}$ the Burgers vector and ρ_{i} is the internal dislocation density.

The stress arising from the grain boundaries (σ_{d}) is negligible when the grain size is large in comparison with the size. The value of the stress arising from the subgrain boundaries (σ_{δ}) depends mainly on the subgrain size (δ) but it is also a function of the Taylor factor, the shear modulus and the Burgers vector [8]. The precipitation hardening stress (σ_{p}) is given by a relationship between different constants, such as the shear modulus and the Burgers vector, the volume fraction of precipitates and the particle radius [9].

The strain rate ($\dot{\varepsilon}$) is related to the dislocation density and to the velocity of dislocations (ν) as [7]:

$$\dot{\varepsilon} = \frac{\boldsymbol{b}}{M} \rho_{\mathrm{i}} \nu, \tag{3}$$

where ν is given by [10]:

$$\nu = C_1 \exp\left(-\frac{Q_{\mathrm{def}}}{k_{\mathrm{B}} T}\right) \sinh\left(\frac{C_2 \sigma_{\mathrm{e}}}{T}\right), \tag{4}$$

where Q_{def} is the deformation activation energy, and C_1 and C_2 are material constants. Eq. (4) can be re-written as a power law [11]:

$$\nu = \beta \sigma_{\mathrm{e}}^{n'}, \tag{5}$$

where n' is an exponent reflecting the sensitivity of the relationship between velocity of dislocation motion and applied stress, which is dependent on the magnitude of the stress. At low stresses, $n' = 1$ for Al–Mg and other strongly solute hardened alloys and $n' = 3$ for pure Al and other weakly solute hardened alloys. At high stresses, n' increases with increasing stress.