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ORIGINAL ARTICLE

High temperature deformation of solution treated 7010 Al-alloy

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Abstract The deformation behavior of 7010 Al-alloy at elevated temperature was investigated. The specimens were solution treated at 723 K for 2 h and then water quenched to obtain the supersaturated solid solution (SSS). Tension tests of the SSS specimens were then conducted at temperatures of 623, 673, and 723 K at various strain rates in the range 5×10^{-5} to 2×10^{-2} s^{-1}. Stress dependence of the strain rate revealed a stress exponent, n of ~8 throughout the ranges of temperatures and strain rates used. This stress exponent is higher than what is normally observed in Al–Zn alloys under similar experimental conditions. This high value of the stress exponent implies the presence of the threshold stress. This behavior resulted from a dislocation interaction with second phase particles due to precipitation at testing temperatures. The values of the threshold stress were observed to decrease exponentially with temperature. The presence of the threshold stress resulted in high apparent activation energy (~180 kJ/mol). The true activation energy was obtained by incorporating the threshold stress in the analysis. The normalized constitutive equation was then developed for the alloy under the present experimental conditions.

1. Introduction

The high-temperature deformation behavior of aluminum and its alloys has received great interest from researchers lately (McQueen and Jones, 1975; Kaibyshev et al., 2002, 2005; Marquis et al., 2003). In aluminum alloys, the alloying element along with their chemical composition has a significant effect on their properties (McQueen and Kassner, 2005; Spigarelli et al., 2003; Mrowka-Nowotnik and Sieniawski, 2005). High strength aluminum alloys have been widely used in aeronautics and automotive applications. Dynamic recovery often occurs at high temperature deformation, which is believed to be the controlling process at that temperature range (Kassner and Perez-Prado, 2000; Mohamed and Langdon, 1974). The formation of the solute atom atmosphere hinders dislocation glide raising the flow stress, but precipitation of fine particles is a more effective source of strengthening. The stress dependence of $\dot{\varepsilon}$ is usually described by a power function; in the so-called power law regime (Kassner and Perez-Prado, 2000):

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$$\dot{\varepsilon} = A\sigma^n \exp\left(\frac{-Q_a}{RT}\right) \quad (1)$$

where σ is the stress, n is the stress exponent, Q_a is the apparent activation energy, R is the universal gas constant, T is the absolute temperature and A is a constant. The value of n and other characteristics were used to classify metals and alloys (Yavari et al., 1981; Soliman and Mohamed, 1982; Oikawa et al., 1984). For Al–Mg alloys and Al–Cu alloys (Soliman, 1995) which exhibit large solid solution hardening (high atomic-misfit parameter), the measured values of n are close to 3, and are characterized by solute drag mechanism resulting from an elastic interaction between dislocations and solute atoms due to the size effect. This type of behavior has been termed as class I or alloy class (Mohamed and Langdon, 1974; Oikawa et al., 1985). On the other hand, for pure metals and some solid solution alloys, usually with low atomic-misfit parameter, the values of n are close to 5 and the strain rate is a function of stacking fault energy. This behavior is termed as class II or pure metal class (Kassner and Perez-Prado, 2000; Mohamed and Langdon, 1974). It is believed that some form of dislocation climb is the rate controlling process. It was suggested that dislocation climb and viscous glide are two sequential processes in solid solution alloys and that the slower process controls the deformation mechanism under the imposed experimental conditions (Mohamed, 1983). It was also suggested that other viscous drag processes can operate sequentially with the solute drag mechanism, but the contribution of these processes to the total drag force depends on the alloy system (Soliman and Mohamed, 1984).

Comparing the hot formability of aluminum binary solid-solution alloys with that of aluminum alloys produced by ingot metallurgy (IM) and powder metallurgy (PM) and that of aluminum-based metal-matrix composites (Mohamed, 1998; Evangelista and Spigarelli, 2002) suggests that the latter materials are characterized by the presence of the threshold stress σ_o, resulting from the interaction of the fine dispersed particles in these alloys with the moving lattice dislocations. Under this condition, the deformation process is not driven by the applied stress but rather by an effective stress σ_e ($=\sigma - \sigma_o$). Hypereutectic Al–17Si was investigated (Spigarelli et al., 2004), and a stress exponent close to ~5 was found. Although the magnitude of the stress exponent observed was equivalent to that observed in pure Al, the apparent activation energy for creep was higher (Q = 210 kJ/mol) than the activation energy for self-diffusion in Al (Q_d = 143 kJ/mol) (Mohamed, 1983). This observation indicated that creep response should be addressed by taking into account more articulated models; they rationalised both the magnitude of the stress exponent and the apparent activation energy for creep, based on threshold-stress concept, which arises due to the interaction between fine particles and dislocations. They calculated the true activation energy, taking into account the threshold stress value, to be 160 kJ/mol, which is comparable to that of pure Al.

In the present paper, the deformation behavior of 7010 aluminum alloy at high temperatures was investigated. The purpose of the investigation is two folds: (a) to examine the presence of the threshold stress and (b) to identify the rate-controlling mechanisms for the deformation of the alloy in the presence of solute atoms and second phase particles.

2. Experimental procedure

The AA7010 alloy was solution treated at 743 K for 2 h and then quenched. The composition of the alloy in wt.% is as follows: 5.7 Zn, 2.4 Mg, 1.84 Cu, 0.29 Si, 0.15 Fe, 0.05 S, 0.03 Mn and the balance is Al. Tensile specimens were machined in preparation for high temperature tests. The tensile specimens were of 15 mm gauge length and cross-section area of 5×3 mm^2 and the tensile axis was parallel to the rolling direction. The tension tests were carried out using an Instron machine model 3385H with a computer and a split furnace containing three heating zones. Samples were soaked for 30 min. before testing in air. Three testing temperatures were used 623, 673, and 773 K; these temperatures were selected as they simulate the real industrial forming conditions of this alloy. The temperature was controlled using a thermocouple connected to the middle of the gauge section. At each temperature different constant speeds were used to impose initial strain rates in the range of 5×10^{-5} to 10^{-2} s^{-1}. The strain rates quoted thereafter represent the initial strain rates calculated from the initial gauge length of the specimens.

3. Results

3.1. True stress–strain curves

The true stress–true strain diagrams shown in Fig. 1(a)–(c) are for temperatures 623, 673, and 723 K, respectively, for different initial strain rates. High strain rate sensitivity was exhibited by the alloy at these temperatures. The true stress–strain response can be divided into three regions: namely strain hardening, steady-state stress, necking and/or cracking.

3.2. Stress dependence of strain rate

The stress dependence of the strain rate is shown in Fig. 2 by plotting the steady state stress vs. strain rate using a double logarithmic scale. The plot was carried out at various temperatures of 623, 673, and 723 K. For the range of strain rates and temperatures tested the data points fall on line segments with slope of n (stress exponent) of ~8.5.

3.3. Ductility

The ductility (elongation at fracture e_f %) is plotted as a function of temperature at various initial strain rates $\dot{\varepsilon}$ (Fig. 3). Ductility increases with increasing temperature and also increases with increasing strain rate.

3.4. Apparent activation energy

The apparent activation energy Q_a (Eq. (1)) can be rewritten for constant strain rate as

$$Q_a = nR\left[\frac{\partial \ln \sigma}{\partial (1/T)}\right]_{\dot{\varepsilon}} \quad (2)$$

The apparent activation energy was calculated by plotting σ vs. $(1/T)$ using a semi-logarithmic scale at three different strain rates (Fig. 4). The activation energy is obtained from the average slope of the data points, which is equal to $\left(\frac{Q_a}{2.3nR}\right)$, which