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Superplasticity in an Al–Mg–Zr–Sc alloy produced by equal-channel angular pressing

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Abstract

The high-temperature plastic deformation of an Al–4.5 wt.% Mg–0.2 wt.% Zr–0.2 wt.% Sc alloy subjected to equal-channel angular pressing was studied, as well as its microstructure. The alloy exhibits high strain rate superplasticity in the temperature range of 573–798 K. Exceptionally high elongations to failure were achieved when testing at an initial strain rate of $4.5 \times 10^{-2}\ s^{-1}$—2130% at 773 K and 1950% at 798 K. At these temperatures, the strain rate sensitivity parameter m exceeded the value of 0.6. The grain size after pressing was <1 μm. The microstructure exhibited exceptional stability, with the grain size <10 μm at a temperature as high as 798 K, due to the presence of $Al_3(Zr_xSc_{1-x})$ precipitates.

Keywords: Aluminum alloys; Equal-channel angular pressing; High strain rate superplasticity

1. Introduction

Ultrafine-grained materials are known to exhibit attractive mechanical properties, namely high strength at low temperatures and superplastic behavior at elevated temperatures. At present, a number of commercial applications have already been found for superplastic alloys, see, e.g. ref. [1]. However, superplastic forming cannot be used for high volume production due to the relatively low strain rates, at which the high ductilities are achieved in the materials available at present. The key to overcoming this problem is the refinement of the microstructure, resulting in the increase of the optimum strain rate for superplastic deformation. Ultrafine-grained materials can exhibit high strain rate superplasticity (HSRSP), which has been defined as superplasticity occurring at strain rates of $10^{-2}\ s^{-1}$ or higher [2]. Severe plastic deformation (SPD) can be used to prepare polycrystalline materials with a grain size smaller than 1 μm. The most widespread SPD method is equal-channel angular pressing (ECAP) [3], which can be used to prepare relatively large bulk samples of ultrafine-grained material with low residual porosity.

When exploring the possibilities of ECAP as a means for the preparation of materials, which exhibit HSRSP, Al–Mg alloys have been in the center of interest. Unfortunately, at elevated temperatures that are necessary for the occurrence of superplastic behavior in such materials, rapid grain growth can be expected due to high grain boundary mobility. For this reason, a variety of additions have been used to stabilize the ultrafine-grained microstructure. Zr and especially Sc additions have proved to be very effective in this respect [5–7]. Recent investigations [8] suggest, that using a combination of Zr and Sc could bring even better results, ensuring a stable microstructure at temperatures exceeding 700 K in ultrafine-grained Al–Mg alloys. The aim of the present investigation was to explore this possibility for an ECAPed Al–4.5% Mg alloy.

2. Experimental material and procedures

The experiments were conducted using an alloy containing, in wt.%, Al–4.5 Mg–0.2 Zr–0.2 Sc. The material was processed by ECAP with six passes at 523 K, using route B_C. The die consisted of two channels with the cross-section of 14 mm × 14 mm, intersecting at 90°. After ECAP, samples were cut from the billets with their tensile axis parallel to the pressing direction. Two different tensile specimen sizes were used, with gauge lengths of 17 and 7.5 mm and cross-sections of 1 mm × 8 mm and 1 mm × 6 mm, respectively. Prior to testing, the samples were polished.

A series of tensile tests were conducted at selected temperatures in the interval of 573–823 K to determine the strain rate

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sensitivity parameter m (defined as $\partial \log \sigma / \partial \log \dot{\varepsilon}$) of the studied material. The method of abrupt strain rate changes was used, as it best complies with the requirement for constant structure. Consequently, an optimum initial strain rate of 4.5×10^{-2} 1 s^{-1} was chosen, and the samples were pulled to failure with a constant rate of crosshead displacement at temperatures from the interval of 573 to 798 K. The samples were placed into the furnace approximately 1 h prior to deformation, in order to approach the state of thermodynamic equilibrium.

The material for the microstructural observations using transmission electron microscopy (TEM) was taken from the non-deformed grip sections of the samples that had been tested at 573–798 K. The material in its initial state (as ECAPed) was examined as well. The foils for TEM observations were prepared by twin-jet polishing.

In addition to TEM observations, the microstructure was studied using the method of optical microscopy (OM). The samples were polished and consequently strained to 10% elongation.

3. Experimental results

Fig. 1 shows the true stress versus true strain rate curves in a double logarithmic scale for temperatures from the interval of 573 to 798 K. A sigmoidal behavior is evident at all temperatures, however, region III has not been reached even at the highest strain rates. The plot in Fig. 1 also documents the decrease of the flow stress with increasing temperature in the interval of 573–798 K. The dependence of m on true strain rate for selected temperatures is shown in Fig. 2. Already at the temperature of 573 K, m reaches a maximum value of 0.46 at a strain rate in the vicinity of 10^{-2} s^{-1}. With increasing temperature, the values of m rise to the maximum of 0.67 at 798 K, while the optimum strain rate remains close to 10^{-2} s^{-1} in the entire temperature interval. At 823 K however, the curve is displaced towards substantially lower strain rates, with its maximum situated near the strain rate of 10^{-3} s^{-1}. Also, at this highest testing temperature, the overall level of the flow stress was higher than at 798 K.

Based on the results presented above, a temperature range of 573–798 K and an initial strain rate of 4.5×10^{-2} s^{-1} were chosen for the tensile tests at a constant rate of crosshead displacement, Fig. 3 shows the corresponding curves. At temperatures ≥773 K and strains higher than ∼1200%, the error on stress measurement was unacceptable due to the extremely low loads. The material exhibited very little strain hardening and the overall level of the flow stress was low, with the exception of the sample strained at 573 K. Fig. 4 documents the recorded ductilities and the appearance of the fractured samples. At the temperatures of 773 and 798 K, exceptionally high ductilities of 2130 and 1950%, respectively, were achieved. Only the sample pulled to failure at the lowest testing temperature of 573 K showed evidence of necking. Even at this temperature however, an elongation to failure of 410% was recorded.

OM observations revealed that grain boundary sliding (GBS) was the dominant deformation mechanism during superplastic deformation of the studied material. This is illustrated in

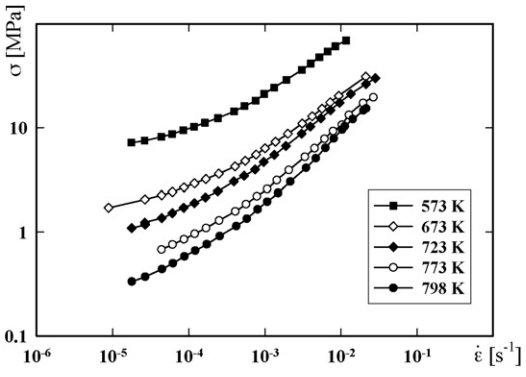

Fig. 1. Dependence of true flow stress on true strain rate in the temperature range of 573–798 K.

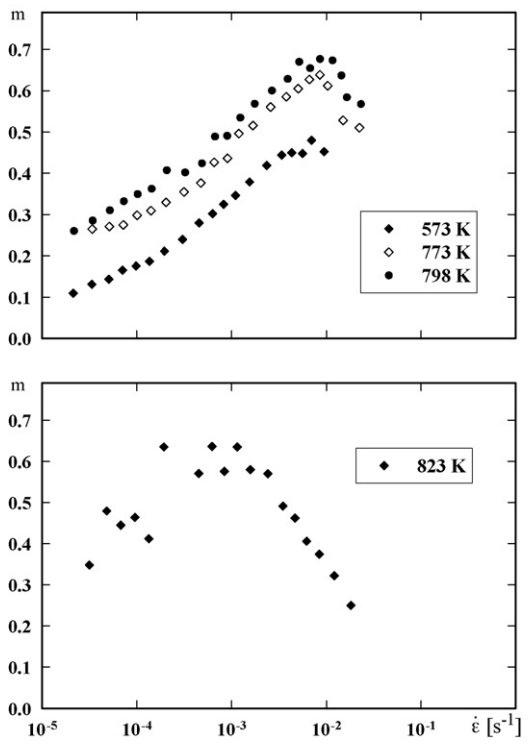

Fig. 2. Dependence of the strain rate sensitivity parameter on true strain rate for selected temperatures.

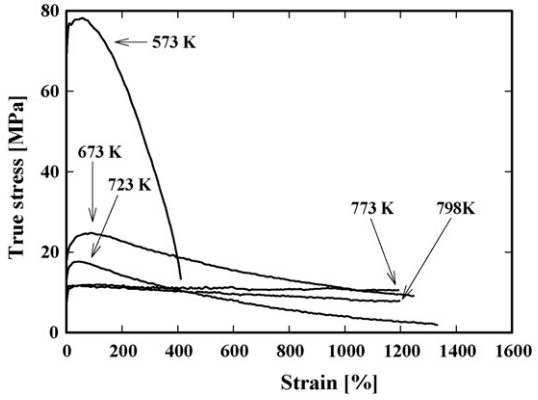

Fig. 3. True stress vs. strain curves for the samples pulled to failure at the initial strain rate of 4.5×10^{-2} s^{-1}.