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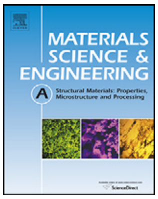

Steady state creep and creep recovery behaviours of pre-aging Al–Si alloys

A.M. Abd El-Khalek

Department of Physics, Faculty of Education, Ain Shams University, Cairo, Egypt

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ABSTRACT

The creep and creep recovery of pre-aging Al–1 wt.%Si and Al–1 wt.%Si–0.1 wt.%Zr–0.1 wt.%Ti alloys have been investigated at room temperature under different constant stresses. The aging temperature dependence of steady creep rate, ε_{st}, and the recovery strain rate, π, show that under the same test conditions first alloy yields creep or creep recovery rates much higher as compared with those of second alloy. The stress exponent n was found to change from 2.5 to 7.43 and 4.57 to 11.99 for two alloys, respectively, characterizing dislocation slipping mechanism. The activation energies of steady state creep of the two alloys were found to be 78.4 kJ/mol and 32.8 kJ/mol for Al–Si and Al–Si–Zr–Ti alloys, respectively. The microstructure of the samples studied was investigated by optical and transmission electron microscopy (TEM).

1. Introduction

Creep may be defined as the continuous deformation of a material with time when subjected to a constant stress or load. Its characteristics are known to depend on the applied stress, the temperature, and the microstructure of the sample [1,2]. It plays an important role in the mechanical behaviour of materials. Extensive deformation under constant stresses and without risk of fracture [3] has been observed in a wide variety of alloy systems. Some experimental changes in the mechanical properties of materials were observed while undergoing a phase transformation [4].

Creep and creep recovery characteristics [5–7] were previously found to be closely related to the internal microstructure of the testing material. Creep characteristics were previously used to determine deformational and relaxation mechanisms associated with the dissolution of precipitates formed in binary solid solutions.

Al–Si alloys are widely used in the automotive and aerospace industries, where they have been steadily replacing many conventional ferrous alloys due to their excellent combination of properties such as good fluidity, low coefficient of thermal expansion, high strength-to-weight ratio, good corrosion resistance and minimum energy requirement for recycling. Mechanical properties of aluminum can be improved by alloying, strain hardening, thermal treatment or by combination of the three techniques. Pure aluminum can be strengthen by alloying with small amounts of Si (up to 1.7 wt.%). Al–1 wt.%Si wires were used extensively for making interconnections in the semiconductors industry. These can subsequently experience mechanical shock and cyclic stress during processing and operation. Hence, many studies [8–10] of structure and mechanical properties of Al–Si alloys were carried out. Both strength and ductility of alloys were observed to be functions of the size and distribution of the silicon particles in the aluminum matrix. In the Al–Si alloys the solubility of Si in aluminum is negligible below 523 K [11], therefore separate phase exist. Above 523 K, the process of Si precipitation cease and, consequently, a reduction in the microhardness of these alloys is observed above 573 K [12].

Alloying of Al–Si with zirconium and titanium was proved [13] to slow down the aging rate and to improve the mechanical properties. Both Zr and Ti have strong grain refining effect on Al–Si alloys and Al–Zn–Mg alloys but the effect is not additive [14,15].

This work is intended to provide some additional information about the steady state creep and creep recovery behaviours of Al–1 wt.%Si and Al–1 wt.%Si–0.1 wt.%Zr–0.1 wt.%Ti alloys at aging temperature range 623–723 K.

E-mail address: Asaad_abdelkahlk@yahoo.com.

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2. Experimental procedure

Two aluminum alloys, Al–1 wt.%Si (hereafter termed alloy A) and Al–1 wt.%Si–0.1 wt.%Zr–0.1 wt.%Ti (hereafter termed alloy B) were prepared by melting 99.99% pure aluminum, 99.99% silicon, Al–5 wt.%Zr and Al–10 wt.%Ti master alloys in a clean graphite crucible in an induction furnace under argon atmosphere. The obtained ingots were homogenized at 823 K for 48 h to reduce segregation, the ingots were then rolled with an intermediate anneal to wires of diameter of 4×10^{-4} m for creep study and sheets of thickness of about 3×10^{-4} m for electron microscope investigation. The compositions determined by chemical analysis for wires and sheets were found to be very close to the alloy compositions required. Samples of both alloys were solution-treated at 823 K for 2 h, to produce the α (solid solution)-phase and the quenched in water kept at room temperature (R.T. ∼ 300 K). Both wire and sheet samples were aged for 2 h at different temperatures (623, 653, 683, 723 K) and then quenched again in water at R.T. The test length of the specimens was initially 50 mm. Creep and creep recovery tests were conducted at room temperature using a conventional constant load creep testing machine described elsewhere [16].

Electropolishing for disc samples of 3 mm diameter was performed in a twin-jet electropolishing unit using a 10% perchlorid acid–90% ethanol solution at 273 K. The voltage and current conditions were ∼24 V and ∼13 mA. Microstructures were examined using a JOEL-100 transmission electron microscope operating at 100 kV.

3. Experimental results

The samples were crept at room temperature for Al–1 wt.%Si (alloy A) and Al–1 wt.%Si–0.1 wt.%Zr–0.1 wt.%Ti (alloy B) until the steady state creep stage was attained. The creep curves were preformed under different constant applied stresses 54.2, 60.5, 63.4 and 65.9 MPa for both alloys A and B. The creep recovery test followed immediately by sudden release of the applied stress. The steady state strain rate, $\varepsilon_{\mathrm{st}}$, was determined from the liner part of the first creep curve whereas the recovery strain rate, π, was determined from the relaxed strain over 10 min period as relaxation time.

Representative examples of creep and creep recovery curves obtained at different aging temperatures, T_{a}, for both alloys A and B are shown in Fig. 1, from which $\varepsilon_{\mathrm{st}}$ and π were determined. It is clear that amount of creep strain for alloy A are generally larger than alloy B, however, the time required to reach the same amount of strain in specimens of alloy B is much longer than in alloy A. Such differences may be best illustrated if the steady state strain rate, $\varepsilon_{\mathrm{st}}$ and the recovery strain rate, π, are plotted against the aging temperature, T_{a}, as shown in Fig. 2(a–d), from which it is seen that under the same test conditions alloy A yields creep or creep recovery rates much higher as compared with those of alloy B. The stress exponent n ($=\partial \ln\varepsilon_{\mathrm{st}}/\partial \ln\sigma$) derived from the slopes of the straight lines relating $\ln \varepsilon_{\mathrm{st}}$ and $\ln \sigma$ (Fig. 3a and b), was found of values ranging from 2.5 to 7.43 for alloy A and 4.57 to 11.99 for alloy B, depending on the aging temperature.

In the aging temperature dependence of the stress exponent n given in Fig. 3c, the stress exponent n increase with increasing aging temperature. Fig. 3c shows that alloy B is harder than alloy A.

The activation energy of steady state creep was calculated for both alloys A and B from the slopes of the straight lines relating $\ln \varepsilon_{\mathrm{st}}$ and $1000/T$ given in Fig. 4. The present results yield average activation energies of 78.4 kJ/mol and 32.8 kJ/mol for alloys A and B, respectively.

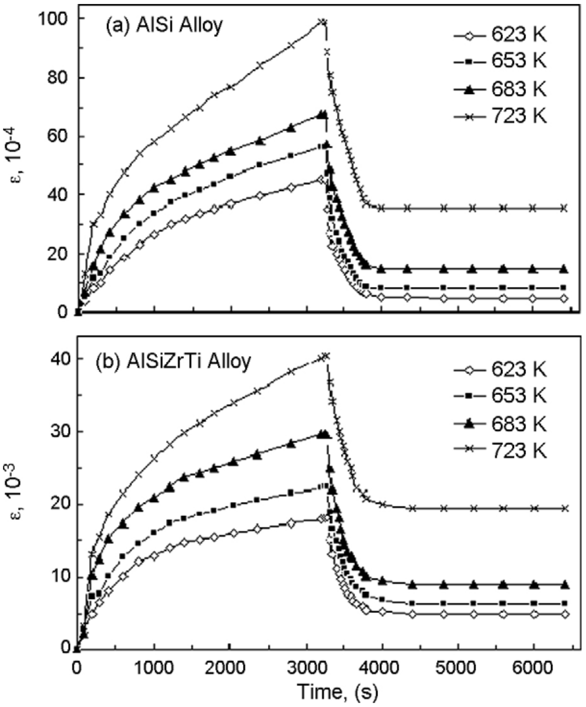

Fig. 1. Representative examples of creep and creep recovery curves obtained at different aging temperatures for: (a) Al–1 wt.%Si alloy, and (b) Al–1 wt.%Si–0.1 wt.%Zr–0.1 wt.%Ti alloy.

In order to correlate steady state creep and creep recovery, the inverse relaxation rate ratio (X^{-1}), where $X = (\pi/\varepsilon_{\mathrm{st}})$ and the residual strain ratio (γ (%) $= (\varepsilon - \pi)/\varepsilon$), were plotted versus different aging temperatures, with the applied stress σ as a parameter (Fig. 5a–d).

4. Discussion

Many properties of crystalline metallic materials are governed by lattice defects. The nucleation of precipitates in Al–Si alloys was enhanced [17,18] by pre-aging quenched specimens near room temperature. Such enhancement can be attributed to dislocation loops formed by condensation of quenched in vacancies.

The nucleation of precipitates in Al–Si alloys was enhanced by pre-aging quenched specimens near room temperature [10]. Such enhancement can be attributed to dislocation loops formed by condensation of quenched in vacancies. Observations showed [19] that Si precipitates nucleated on vacancy clusters, but their formation and the subsequent nucleation took place within a few seconds after quenching.

In reviewing the creep data of both alloys shown in Fig. 1(a and b), it is obvious that Zr and Ti additions greatly affect the work hardening of the steady state creep and creep recovery. The difference in both rates of steady creep ($\varepsilon_{\mathrm{st}}$) and creep recovery ($\pi$) for both alloys A and B are due to the variation in their grain size as seen from the optical micrographs given in Fig. 6 (a and b). As the grain size for alloy B decreases, due to the grain refining effect of Zr and Ti additions, $\varepsilon_{\mathrm{st}}$ and π decreases, since larger area of grain boundaries should be more effective as impeding agent for dislocation motion. For both alloys A and B, aging the specimens in the temperature range 623–723 K resulted in loss of strength, i.e., $\varepsilon_{\mathrm{st}}$ and π increases (Fig. 2). This due to a coarsening of Si particles with increasing aging temperature (in alloy A) as confirmed by electron micrographs (Fig. 6 c and d). Such a coarsening results in