

Work hardening characteristics in Al base alloys with 12.6 and 45 wt.% Zn

F. Abd El-Salam^a, M.M. Mostafa^a, L.A. Wahab^b,
M.T. Mostafa^a, Sh.M. Abd El-Aziz^{a,*}

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

^b National Centre for Radiation Research and Technology, Nasr-City, Cairo, Egypt

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Abstract

The stress–strain curves were obtained for Al–Zn alloys of 12.6 wt.% Zn (alloy I) and 45 wt.% Zn (alloy II) with elements of purity (99.99). The monotonic shift of these curves towards lower flow stress and higher ductility was interrupted at the transformation temperatures 483 K (alloy I) and both 543, 603 K (alloy II). By increasing deformation temperature, Young's modulus, Y , yield and fracture stresses, σ_y and σ_f , respectively, fracture time, t_f , the coefficient of parabolic work hardening, χ , decreased while fracture strain, ϵ_f , and dislocation slip distance, L , increased. From the obtained X-rays diffraction patterns the lattice strain, ϵ , crystallite size, η , and dislocation density, ρ , were obtained at different deformation temperatures around transformation.

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1. Introduction

Zinc is sufficiently available, but it is well known for the problems related to wet ability and corrosion and in very liable to oxidation [1]. Al is one of the suitable candidates which improve the oxidation resistance of Zn [2].

In metals and alloys, the hardening induced by forming a structure in which dislocation mobility is reduced may be due to the intersection of dislocations, interaction of dislocations with impurity atoms, the formation of second phase particles and ordering [3].

Theoretical investigation of binary Al–Zn system suggests that it has the tendency for the self-coordination leading to segregation.

Also, the study of concentration fluctuations in the long wavelength demonstrates that a degree of like–atoms pairing exists in Al–Zn system [1].

The formation of precipitates greatly increases the strength of an alloy because Frank–Read dislocation sources are supposed to cause hardening of polycrystals due to the subsequent formation of dislocation pile-ups at grain boundaries.

The solution of Al–Zn alloys by rapid quenching from the α -phase field produces a very fine structure with grains of very high dislocation density [4] (Fig. 1).

The sequence and mechanism of structure transformation in supersaturated solid solutions, was found [5] to start with the formation of the spherical and the ellipsoidal Guinier–Preston Zones (GPZ), which are rich in solute atoms, coherent with the matrix and differ from each other in the thermal stability.

The transformation of GPZ, produces the rhombohedral transition phase (R-phase) which while losing some coherency yields the Zn rich fcc α' -phase and finally the Zn-rich stable precipitates. These thermally induced structure variations affect largely the hardening parameters of the alloy. So, the aim of the present work is to investigate the temperature and structure dependence of the tensile characteristics of Al–Zn system including (12.6 and 45) wt.% Zn.

2. Experimental procedure

The Al–Zn alloys of 12.6 wt.% Zn (alloy I) and 45 wt.% Zn (alloy II) were prepared from highly pure Al and Zn (99.99%).

The ingots were homogenized at 643 K for 24 h then drawn into wires of diameter 0.7 mm.

* Corresponding author. Tel.: +20 123649582; fax: +20 25524403.
E-mail address: Shereen_moh@yahoo.com (Sh.M. Abd El-Aziz).

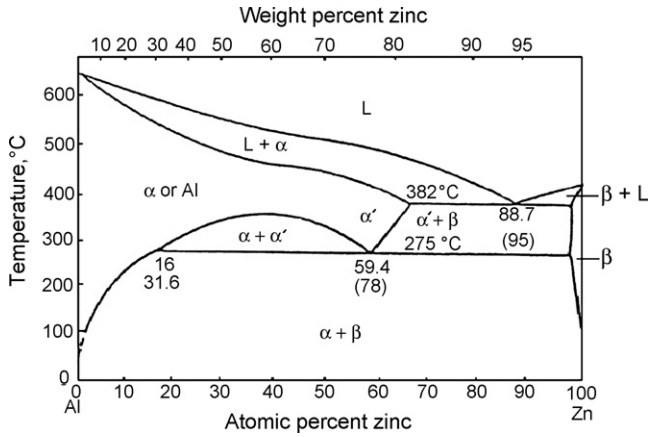


Fig. 1. Al–Zn phase diagram.

The samples were 5 cm long. Both alloys were annealed at 643 K for 2 h then rapidly quenched to room temperature (~ 300 K) to retain the fcc α -solid solution formed at 643 K.

Stress–strain tests were applied to the samples of both alloys in the temperature range 300–523 K for samples of alloy I, and up to 643 K for alloy II till fracture. A conventional-type tensile testing machine described elsewhere [6] was employed. The applied stress was gradually increased (with 30 s internal between two successive loadings) and the elongation was recorded with an accuracy of $\pm 10^{-5}$ m, immediately after applying the stress.

The yield stress, σ_y , is considered to be the stress corresponding to the first deviation from linearity in the starting part of the stress–strain curve. The maximum stress, the sample can endure before fracture was taken as the fracture stress, σ_f , and the corresponding strain was considered as the fracture strain, ϵ_f .

Property variations due to composition or thermal treatment are caused by structural variations.

X-rays diffraction can provide accurate information about the internal state of a material after deformation. Philips X-ray unit (PW 3710) with generator (PW 1830) was used to obtain the diffraction patterns for both alloys, with beam of wavelength $\lambda = 0.15406$ nm. Applying the equation:

$$\frac{\beta \cos \theta}{\lambda} = \frac{1}{\eta} + \frac{2\epsilon \sin \theta}{\lambda} \quad (1)$$

where θ is the peak angle of a plane, β the diffracted X-ray breadth, η the crystallite size from which the dislocation density is obtained as $\rho = 1/\eta^2$, and ϵ is the lattice strain. Half the slope of the linear relation between $(\beta \cos \theta)/\lambda$ and $(\sin \theta/\lambda)$ refers to the lattice strain ϵ , and the inverse of the intercept on the ordinates gives the average crystallite size η , from which dislocation density ρ is obtained.

3. Experimental results

The stress–strain curves obtained for alloys I and II taken at different deformation temperatures in the temperature range specified for every alloy, is given in Fig. 2a for alloy I, and in Fig. 2b for alloy II.

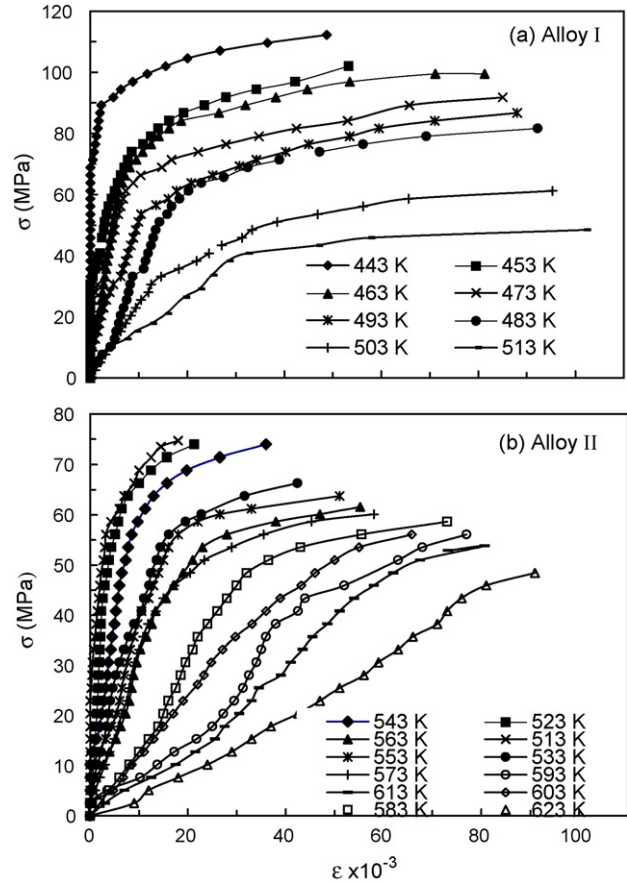


Fig. 2. The stress–strain curves at different working temperatures for: (a) alloy I and (b) alloy II.

The monotonic shift of these curves towards lower flow stress and higher ductility when increasing the deformation temperature is interrupted at 483 K for alloy I, and at 543, 603 K, for alloy II, respectively.

The softening of the deformed samples at high deformation temperatures is revealed from the behaviour of the temperature dependence of the following hardening parameters:

1. A decrease of the stress per unit strain, $\partial\sigma/\partial\epsilon$ (Yong's modulus Y), keeping the stress constant in the starting part of the stress–strain part of the stress–strain curves for both alloys, Fig. 3a.
2. The decrease of yield stress, σ_y , with increasing deformation temperature, Fig. 3b.
3. A decrease of the fracture stress, σ_f , which is the last maximum stress applied to the sample before fracture, Fig. 3c. The decrease in fracture stress points to a decrease in fracture time, t_f .
4. A decrease of the coefficient of parabolic work hardening, χ , obtained as [7]:

$$\chi = \frac{\partial\sigma^2}{\partial\epsilon} \quad (2)$$

which is given in Fig. 3d.

5. An increase in fracture strain, ϵ_f Fig. 3e.

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