PLASTIC AND SUPERPLASTIC BEHAVIOUR OF Cd-2WT. % Zn AND EUTECTIC Cd-Zn ALLOYS

M. M. Mostafa, H. A. El-Sayed, M. S. Sakr and M. R. Nagy
Physics Department, Faculty of Education, Ain Shams University

Abstract

The structure and dynamic deformation behaviour of Cd-2wt. % Zn alloy (A 1) and eutectic Cd – Zn alloy (A 2) under different applied stresses and thermal cycling rates are investigated. By studying the dynamic creep, it is established that the dynamic deformation of the first alloy (A1) is plastic, due to the presence of an initial large grain size (d_{in} = 100 μm) and grain growth during phase transformation (d_f=500 um). While for the second alloy (A 2) is superplastic due to the presence of an ultra-fine grain structure (d_{in} = 1 μm) and the realization of typical superplastic grain boundary diffusion mechanisms.

Introduction

Some investigators [1-3] consider the decisive mechanism to be grain boundary glide, and others, the development of diffusion processes [4]. In the previous work [5] it has been shown that subgranular slip is observed during superplastic flow. It is probable that the possibility of deformation without necking occurs when a certain favourable combination of above mentioned mechanisms is achieved. Besides this, it is important in our understanding of the nature of superplasticity to ascertain the role of these above mechanisms.

It is now well established that when materials are deformed superplasticity, deformation is often accompanied by grain growth [6]. Over a wide range of strain rate, covering that which is generally referred to as regions I and II of superplastic behaviour [1], the grains grow at a much greater rate than that shown during stress free annealing at the same temperature. This phenomenon, which is referred to as strain enhanced grain growth [6-9], has been widely observed in quasi-single phase metallic systems [7, 10, 11].

In both experiment and theory, superplastic alloys are generally regarded as being purely strain rate, temperature rate and temperature sensitive [12-17]. The influence of structure of superplastic material for eutectic alloy is confined to the strong effect of grains size, and hardening during inelastic strain is associated with growth [12, 18].

The aim of the present work is to study the effect of thermal cycling and applied stress on plastic, superplastic and grain growth during phase transformation (0.8 T-eutectic) for Cd–2 wt. % Zn (A1) and eutectic Cd-Zn (A 2) alloys respectively.

Experimental Technique :

Cd - 2 wt. % Zn (A 1) and eutectic Cd-Zn (A 2) alloys were prepared from high purity cadmium and Zinc (99.99%). The samples were in the form of wires of diameter 1×10^{-3}m and length 3×10^{-2}m. They were annealed for two hours at 553 K & 523 K for the first and second alloys respectively, and slowly cooled to room temperature with a relatively low temperature rate of $T = 2 \times 10^{-2} K. s^{-1}$. The initial grain diameters were 0.1 mm and 1 μm for Cd-2 wt. % Zn (A 1) and eutectic Cd-Zn (A 2) alloys respectively.

Samples maintained under different tensile stresses were thermally cycled across the phase transformation temperature for alloy (A 1) and 0.8 T_e (T_e eutectic temperature) for alloy (A 2) with different cycling rates.

The structure of the test samples was determined after the temperature cycle using a D-500 x-ray diffractometer.

Experimental results :

Figure (1 a) and (1 b) shows the dependence of the tensile deformation on the thermal cycling conditions across the transformation temperature for alloy (A 1) and across (0.8 T_e) for alloy (A 2). The tensile strain increased with increasing number of temperature cycles until reaching to total tensile strain E_{tot} about 30% for Cd-2 wt. % Zn which is in the range of plastic strain, and 300%

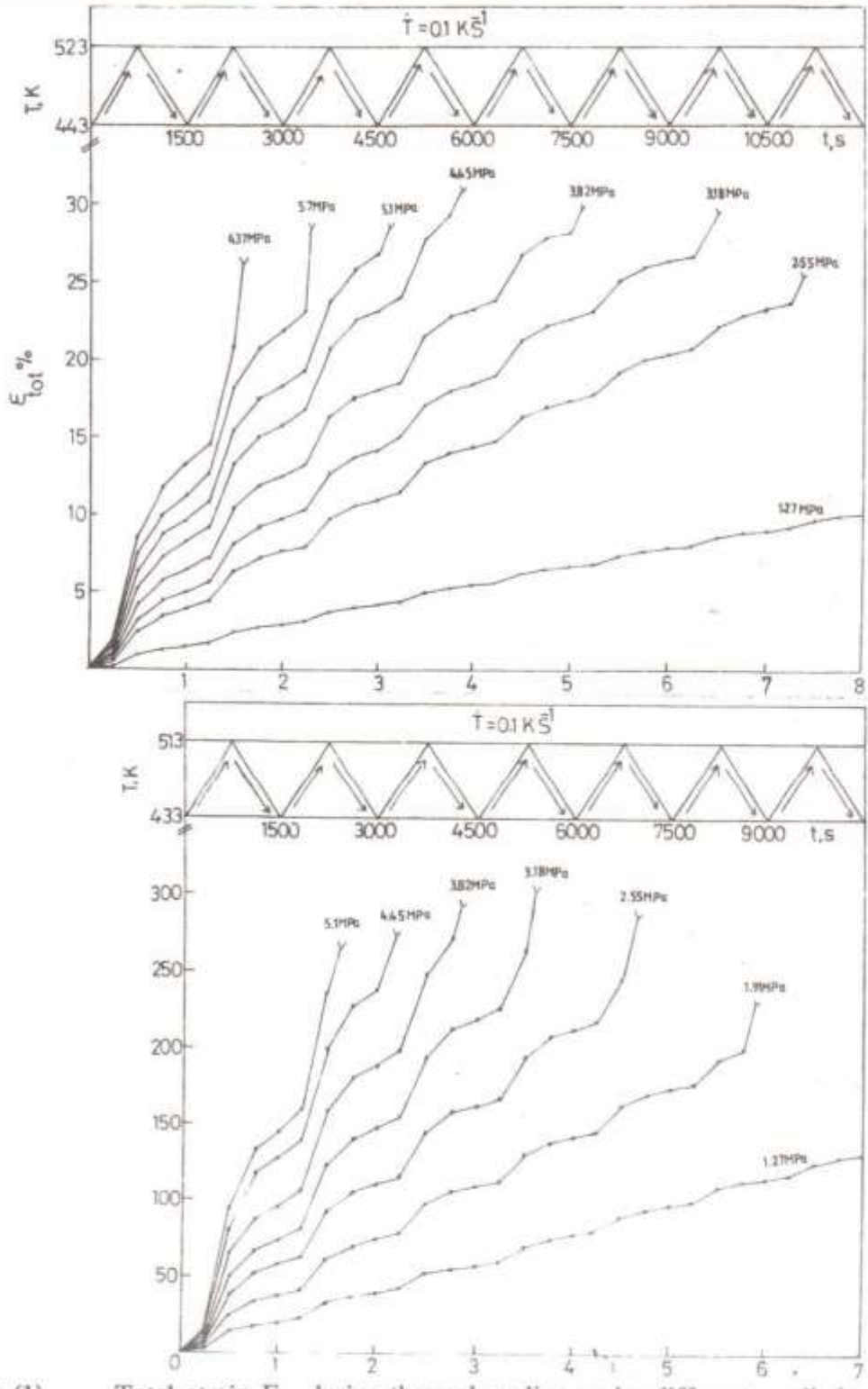

Figure (1) Total strain E_{tot} during thermal cycling under different applied stresses for (a) Cd-2 wt. % Zn alloy (A 1) and (b) eutectic Cd-Zn alloy (A 2).

for eutectic Cd-Zn alloy which is in the range of the superplastic strain respectively It can be seen that within each thermal cycle, the deformation rate changed as the sample passed the transformation temperature 0.8 T_e in the forward and reverse directions. The strain during heating was large relative to it during cooling within each thermal cycle.

The schematic diagram and the typical behaviour of the strain-time dependence for Cd-2 wt.% Zn and eutectic Cd-Zn alloys during half cycle at forward direction are shown in Fig. (2).

Figure (3) shows the relation between E_{ph1} , E_{p1} and $E_{sup\,2}$ against the number of cycles under different constant applied stresses and constant heating and cooling cycling rate for alloy (A1) and alloy (A 2). These curves indicated that E_{ph1} , E_{p1} and $E_{sup\,2}$ were increased by increasing the applied stress and the number of cycles. The obtained results of Fig. (3 A) were found to verify the following equation.

$$E = B\,t^{n} \quad \ldots\ldots\ldots\ldots\ldots \quad (1)$$

where E represents each of E_{ph1} , E_{p1} and $E_{sup\,2}$, t is the time of forward half cycles in seconds. The parameter n was found to be equal to 0.9 and the parameter B ranged from 1.1×10^{-5} to 1.23×10^{-3} (see Fig.3 B).

Figure (4 a) shows the relation between the applied tensile stress and both the plastic and superplastic strain rates during thermal cycling. The rates E_{p1} and $E_{sup\,2}$ of plastic strain and superplastic strain for alloy (A 1) and alloy (A 2) respectively are given by the plastic and superplastic deformations E_{p1} and $E_{sup\,2}$ divide by the times t_{p1} and $t_{sup\,2}$ of plastic and superplastic deformations respectively (see Fig. 2 a). The plastic deformations of alloy (A 1) is too small relative to the superplastic deformation achieved by eutectic alloy (A 2) under identical working conditions. It is clear that the plastic and superplastic strain rates increase with increasing the applied stress during the forward transformation process (heating half cycle). For both alloys, the strain rate sensitivity parameter m was determined experimentally from the slope of the linear portion ln, Figure (4 b), and was found to be m=0.7

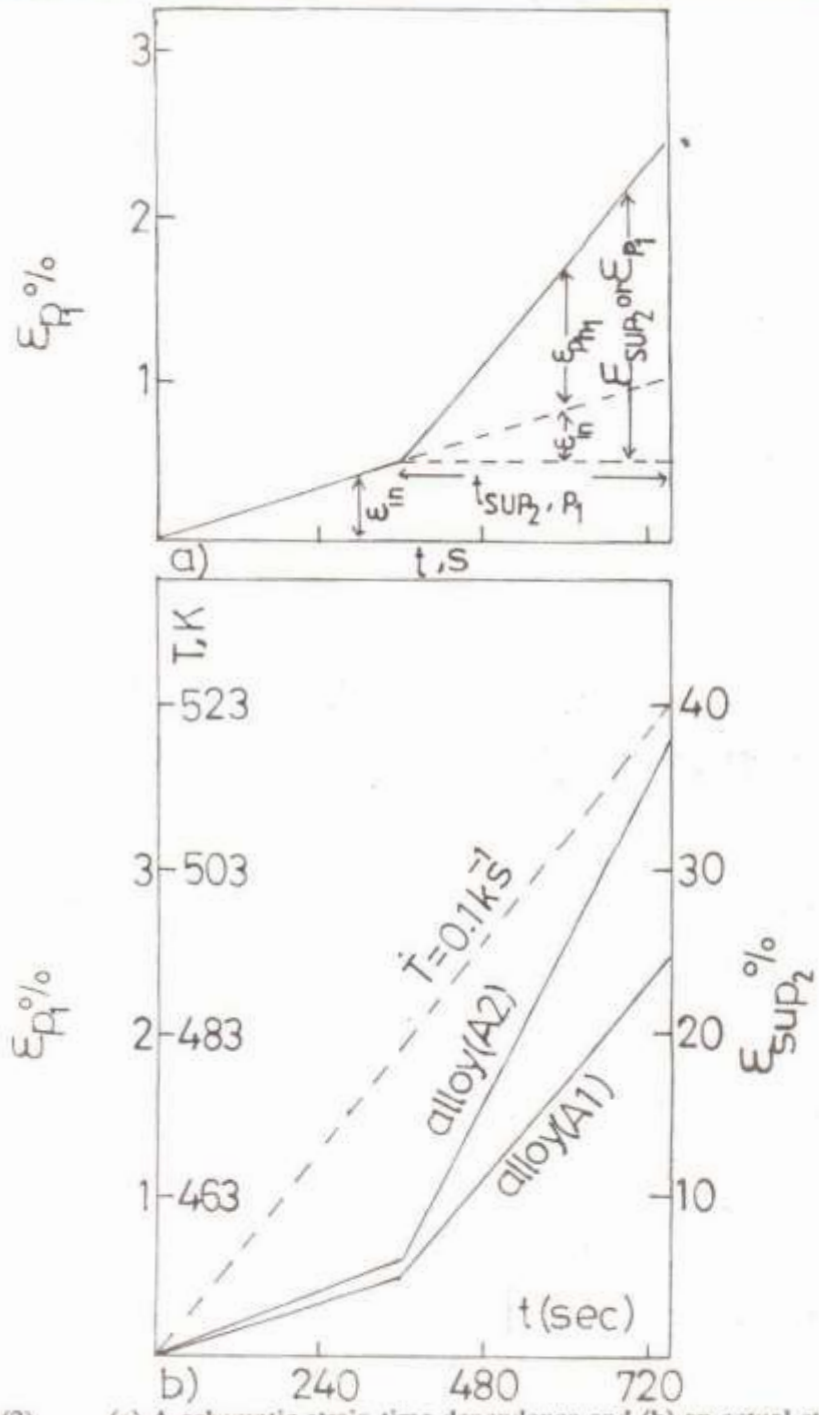

Figure (2) (a) A schematic strain-time dependence and (b) an actual strain-time dependence (σ = 2.55 MPa; T=0.1 Ks^{-1}) across forward transformation for both alloys (E_{pl} for alloy (a1,), E_{sup2} for alloy (A2).

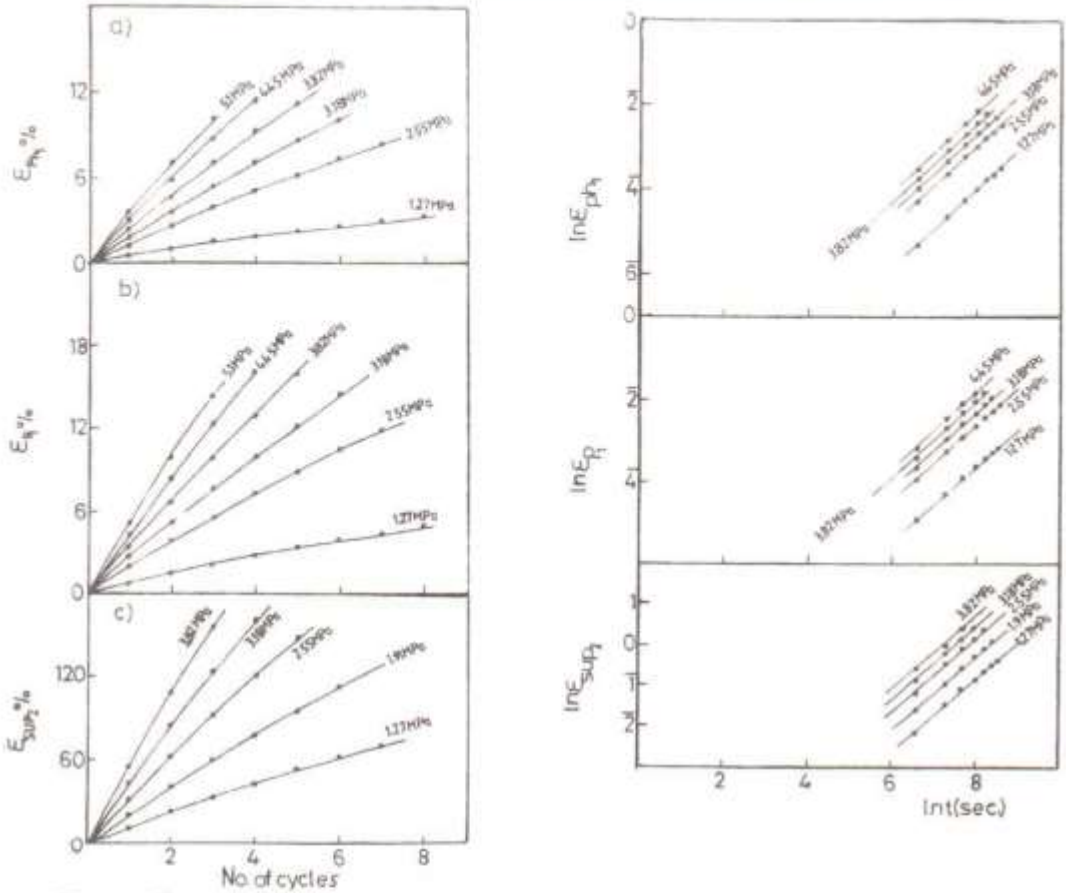

Figure (3) (a) The transformation strain E_{ph1} and (b) the plastic strain E_{pl} for alloy (A1); (c) the superplastic E_{sup2} of alloy (A2) as a function of the number of cycls under different applied stresses.
(b) Relation between ln E_{ph1} and ln E_{pl} of alloy (A1); ln E_{sup2} of alloy (A2) versus ln t at different applied stresses.

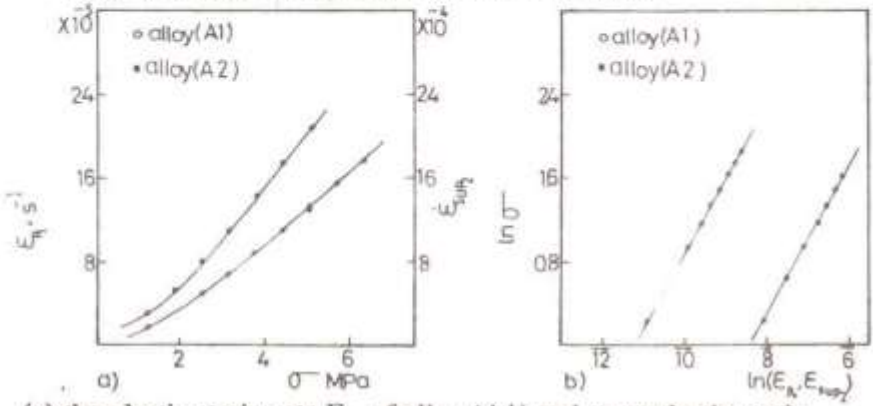

Figure (4) (a) the plastic strain rate E_{pl} of alloy (A1) and superplastic strain rate E_{sup2} of alloy (A2). as a function of applied stress σ (T = 0.1 K s^{-1}).

Figure (5) shows the dependence of the plastic and superplastic strain rates E_{pl} and $E_{sup\ 2}$ on the cyclic temperature rate T during heating half cycle which indicate that the plastic and superplastic strain rates increase with increasing the cyclic temperature rate.

An initial deformation E_{in} was observed immediately after loading (see Fig. 2 a), the superplastic deformation $E_{sup\ 2}$ for eutectic alloy (A 2) was very pronounced after the onset of 0.8 T_e) T_e = eutectic temperature, 483 K). The difference between E_{pl} and E_{in} is called E_{phl} and characterized the part of creep which is solely dependent on the transformation reaction.

The relation between transformation deformation E_{phl}, superplastic deformation E_{sup2} and the corresponding initial deformation E_{in} within constant intervals of time are shown in Fig. (6). It can be seen that E_{in} / E_{sup2} and E_{in} / E_{sup2} are nearly constants for different stresses and equal to about 37% and 21% respectively.

The average grain diameter (d) of the first alloy (A 1) and the integral intensity (I), the lattice parameter (a_o), the ratio (c_o / a_o) of the hcp α-phase (Cd-rich phase) and the half line width ($\Delta\ 2\theta$) for both alloys as a function of the total deformation rate E_{tot} are shown in Fig. (7). In case of the first alloy (A 1) d, I, a_o and c_o / a_o were found to increase while ($\Delta\ 2\theta$) decrease with the increase the rate of total strain. Regarding the second alloy (A 2) the parameters I, c_o / a_o and $\Delta\ 2\theta$ were found to decrease while the lattice parameter a_o was constant with increasing the rate of total strain.

Discussion :

The dependence of the tensile deformation of the thermal cycling conditions through the transformation and 0.8 eutectic temperature (483 K) [19] of the first and second alloys, respectively, is shown in figure (1 a,b).

At constant applied stress, it could be seen that the total strain is increased with increasing the number of cycles. This behaviour might

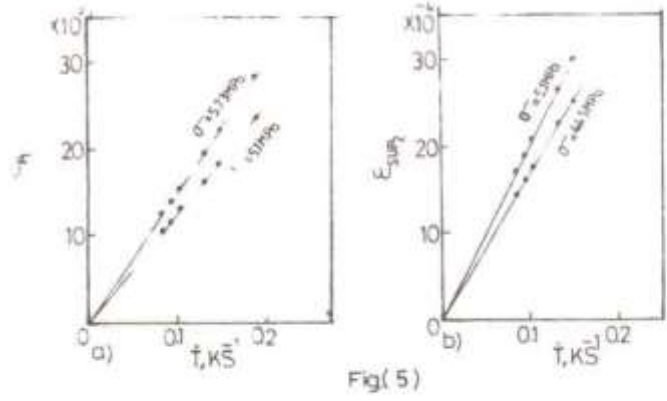

Figure (5) The influence of cyclic temperature rate T on (a) the plastic strain rate E_{pl} for alloy (A1) and (b) the superplastic strain rate E_{sup2} for alloy (A2).

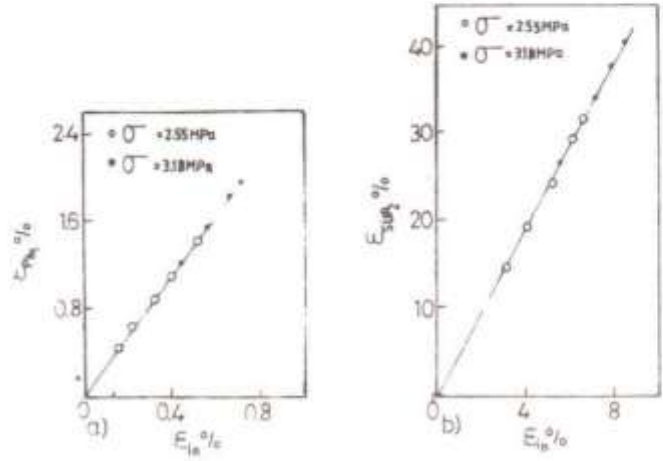

Figure (6) (a) The relation between transformation strain E_{pl1} and initial strain E_{in} for alloy (A1) (T=0.1K s^{-1}).
(b) The relation between suprplastic strain E_{sup2} and initial strain E_{in} for alloy (A1) (T=0.1K s^{-1}).

be attributed to the increase in the rate of mobile lattice defects which are generated during transformation in the first alloy (A 1) and due to grain boundary sliding of the ultra-fine grain of the second alloy (A 2). Moreover, the total strain per cycle is found to increase with increasing the applied stress. This observation might be attributed to the superposition of the applied stress and the induced internal stress during the transformation reaction in the first alloy (A 1) and grain boundary sliding in the second alloy (A 2). It can be seen that within each thermal cycle the deformation rate changes as the sample passed through the transformation temperature and 0.8 T_e of the two alloys in the heating direction. The values of the strain and strain rate of the forward half cycle (heating direction) are greater than those of the reverse direction (Fig. 1). For the same applied stress, the same rate of temperature cycle (T) and the number of cycles, the parameter E_{sup2} of the eutectic alloy (A 2) is larger than E_{p1} and E_{ph1} of the first alloy (A1), (see Fig. 3). This observation might be as a result of the ultrafine grain size of the eutectic alloy relative to the first composition which leads to an intensive grain boundary diffusion mechanism The derived semi-empirical relation (1) verified that the dynamic creep during successive cycles is similar to the transient creep equation. The values of strain rate sensitivity (SRS) parameter m during heating have been found to be 0.7 for both alloys, (Fig. 4 b). This value indicates that the tensile deformation in the heating direction is controlled mainly by the diffusion mechanisms [20]. The increase of plastic strain rate of the first alloy (A 1) E_{p1} and the superplastic strain rate of the secondly alloy (A 2) E_{sup2} in heating direction with increasing the rate of temperature cycling (T), (Fig. 5) implies that the diffusion processes have been enhanced by increasing cyclic rate. From figure (6), it seen that $(E_{in} / E_{ph1}) = 37\%$ and $(E_{in} / E_{sup2}) = 21\%$. These results were found to be the same for all successive cycles. Since the initial deformation (E_{in}) is controlled mainly by dislocation mechanisms [21], it could be concluded that the role of dislocation mechanism presents about 37% of the observed transformation strain of the first alloy (A 1) and 21% of the observed superplastic strain of the second alloy (A 2). For both alloys, the remaining part for the

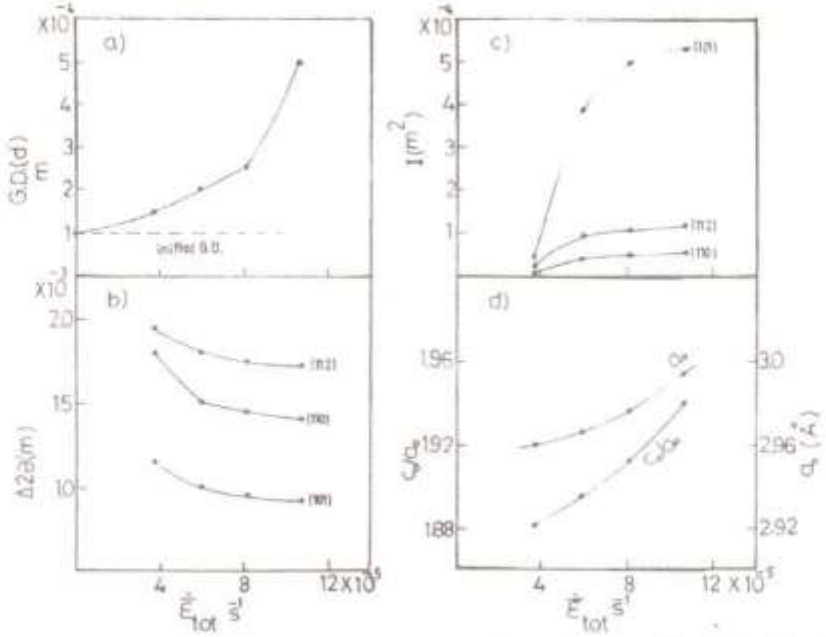

Figure (7 A) The relation between the total strain rate E_{tot} and (a) the grain diameter (b) the X-ray linewidths of α-phase (c) the integral X-ray diffraction intensities for α-phase and (d) the lattice parameters of α-phase, for alloy (A1).

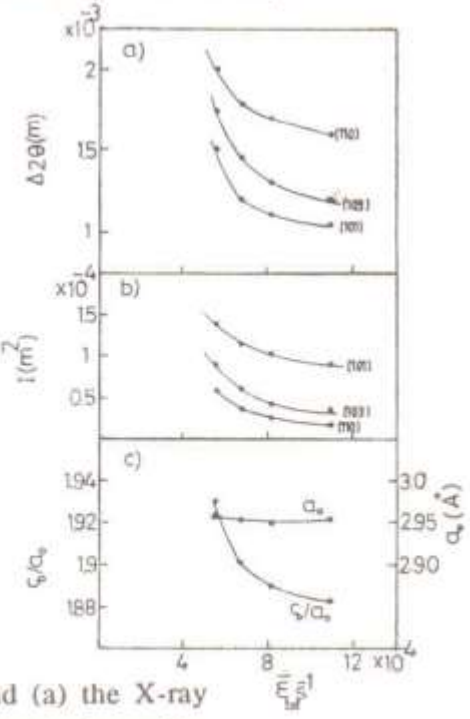

Figure (7 B)

The relation between total strain rate E_{tot} and (a) the X-ray linewidth of α-phase, (b) the integral X-ray diffraction intensities for α-phase and (c) the lattice parameters of α-phase, for alloy (A2).

observed transformation and superplastic deformations might be controlled by diffusion processes.

From microstructure investigations (Fig. 7) the increase of the average grain diameter (d), the integral intensity (I), the lattice parameter a_o and the ratio c_o / a_o of the lattice constants and the decrease in the half line width (Δ 2θ) of Cd-rich phase for the first alloy (A 1) confirmed that the superplasticity did not occur in the first alloy (A 1) due to the growth of the grain size from 100 μm – 500 μm during dynamic deformation through the phase transformation, while for the eutectic alloy (A 2) of the ultrafine grain diameter [19], the decrease in the integral intensity (I), the half line width (Δ 2 θ) and the ratio c_o / a_o of the lattice parameters and the constancy of the lattice parameter a_o with the total strain rate indicate that the relief in the internal stresses is relatively higher in this alloy due to the intensive grain boundary diffusion mechanism.

References

1. G. J. Davies, C. W. Edington, C. P. Culter and K. A. Padmanabhan, J. Mater. Sci. 5 (1970) 1091.
2. D. S. Holt, Trans. met. Soc. AIME, 242 (1968) 740.
3. D. Lee, Acta Met. 17 (1969) 1057.
4. A. Karim and W. A. Backofen, Metallurgical Trans. 3 (1972) 709.
5. O. A. Kaybyshev and I. V. Kazachkov, Fiz. metal metalloved. 34 (1972) 396.
6. D. S. Wilkinson and C. H. Caceres, Acta Metall. 32 (1984) 1335.
7. D. S. Wilkinson and C. H. Caceres, J. Mater. Sci. Lett., 3 (1984) 395.
8. D. S. Wilkinson, C. H. Caceres and X. Wv, in Proc. Conf. °1CSMA 7, 871, 1985, Oxford, Pergamon Press.
9. D. S. Wilkinson, in Proc. Conf. on superplasticity, (ed. B. Baudelet and M. Suery) 6.1 (1985) Grenoble, CNRS.
10. M. A. Calrk and T. H. Alden, Acta Metall. 21 (1973) 1195.
11. C. II Caceres and D.S. Wilinson, Acta Metall. 32 (1984) 415.
12. A. K. ghosh and C. H. Hamilton, Met. Trans. 10A (1979) 699.
13. T. H. Alden in plastic deformation of materials, ed. Arsenault, p. 226, N. Y., Academic Press, (1975).

14. M. F. Ashby and R. A. Verrall, Acta Met. 21 (1973) 149.
15. R. C. Gifkins, Met. Trans. 7A (1976) 1225.
16. M. R. Nagy, M. S. Sakr and R. Kamel, Mater. Sci. and Eng. 52 (1982) 257.
17. M. A. Kenawy, M. R. Nagy and E. M. Sakr, J. Mater. Sci. 21 (1986) 3071.
18. B. N. Watts and M. J. Stowell, J. Mat. Sci. 6 (1971) 228.
19. M. S. Sakr, A. El-Shazly, M. M. Mostafa, H.A. El-Sayed and A. A. Mohamed, Czech, J. Phys. 40 (1990) 1261.
20. E. W. Hart, Acta. Met. 15 (1967) 351.
21. J. Friedel, Dislocation, Pergamon Press, London, P. 304, 317 (1964).