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Evaluation of flow parameters in transformation plasticity of an Sn–5 wt% Cd alloy

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The transformation plasticity in an Sn–5 wt % Cd eutectoid alloy was studied using isothermal tensile creep and Instron tests on specimens up-quenched to above the eutectoid temperature. As a result of concurrent transformation, an enhanced creep rate was noticed for several minutes from the start of such tests. By investigating the stress and temperature dependence of the enhanced creep region, a stress exponent of 1.1 and an activation energy of 56 kJ mol^{-1} were obtained. The role of enhanced diffusion in diffusional transformation plasticity is confirmed. With concurrent transformation, the flow stress decreased rapidly with strain before it increased to the level of steady state flow stress characteristic of fully transformed material.

1. Introduction

Materials undergoing phase transformation under a nominal applied stress exhibit excess plastic deformation known as transformation plasticity. This phenomenon can be exploited to obtain superplastic strains through thermal cycling under a small load around the phase transition temperature. According to Greenwood and Johnson [1], the transformation plasticity results from flow in the weaker phase due to the combined action of applied stress and internal stress arising from volume dilatation of the phase transformation. While the Greenwood–Johnson model leads to a phenomenological understanding of the transformation plasticity, further considerations of the operative micro-mechanisms are essential. The suggested qualitative mechanisms for diffusional transformation plasticity are (a) loss of cohesion of the bonding between atoms [2–4], (b) enhanced diffusion and dislocation climb [5], (c) dislocation motion [6], and (d) transient ultrafine grain superplasticity [7]. One drawback in identifying the operative mechanisms is the lack of experimental data on constitutive equations representing the transient flow behaviour of transformation plasticity, although several test procedures such as constant load creep tests with temperature cycling [1, 2, 8–10], isothermal constant load creep tests at transformation temperature [6], differential strain rate test while the transformation is taking place [7, 11], and hot hardness test during transformation [1], have been employed in this regard. We have studied the transformation plasticity in an Sn–5 wt % Cd eutectoid alloy by up-quenching from room temperature to above the eutectoid temperature (133° C) and then carrying out isothermal tensile and creep tests. The intent of this paper is to report our experimental results relating to the flow parameters for diffusional transformation plasticity.

2. Experimental procedure

Sn–5 wt % Cd eutectoid alloy was prepared by melting tin and cadmium, both of 99.99% purity, in proper proportions in a vacuum sealed pyrex tube of 50 mm diameter. The solidified alloy was homogenized at 160° C for 24 h and then extruded at room temperature to a rod of 12.5 mm diameter. These were further swaged to 9 mm diameter rods, which were annealed at 170° C for 2 h, quenched into an oil bath at 120° C, held for 2 h at 120° C and then air-cooled to room temperature.

Tensile specimens of 4 mm diameter and 25 mm gauge length were obtained from the above heat-treated material. Isothermal tensile creep and Instron tests were carried out at test temperatures above the eutectoid temperature up to 170° C employing an electrically heated silicone oil bath.

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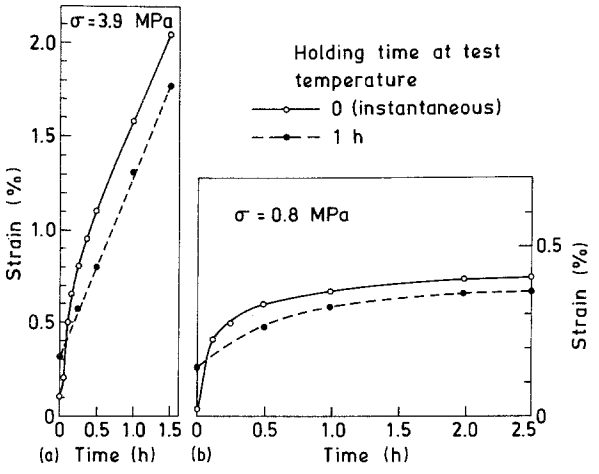

Figure 1 Creep curves of specimens up-quenched to 150° C and tested immediately or after a holding time of 1 h.

Some of the tests, referred to as those without concurrent transformation, were performed by holding the specimens at the test temperature for 1 h. Others referred to as concurrent transformation tests were conducted within 1 min after immersing the specimens in the hot oil bath set to the desired test temperature. In the case of Instron tests, both constant corsshead speed and differential strain rate [12] tests were carried out.

3. Results

The microstructure of the starting material at room temperature consisted of grains of 94 μm size, as measured by the linear intercept method, with lamellar structure of the two phases obtained by eutectoid decomposition. Typical creep curves for specimens with and without concurrent transformation are shown in Fig. 1 for a test temperature of 150° C. In specimens without concurrent transformation, primary creep was evident only at the lowest stress employed. For higher stresses, steady state creep was dominant right from the start. On the other hand, specimens with concurrent transformation exhibited initially a linear region of high creep rate followed by a steady state creep region with the same creep rate as that of fully transformed material (Fig. 1). This initial linear region of high creep rate is a consequence of concurrent transformation and it is considered as a pseudo-steady state creep region representing the transformation effect. The steady state creep rate data of specimens without concurrent transformation and the pseudo-steady state creep rate data of specimens with concurrent transformation as a function of applied stress are presented in Fig. 2 for various test temperatures. The observed values of the stress exponent n for specimens with and without concurrent transformation are 1.1 and 3.1, respectively, whereas the corresponding activation energies for creep in the above two cases as obtained from Arrhenius plots (Fig. 3) are 56 and 102.4 kJ mol^{-1}.

On comparing the creep curves (Fig. 1) of specimens with and without concurrent transformation, it is evident that the total strain accumulated in the initial period of the creep test is more in the former case. The difference in creep strains remains the same at later times when both the types of specimens attain the same steady state creep rate. The excess creep strain due to transformation has been obtained by subtracting the total strain accumulated for a fixed time in a specimen without transformation from that with transformation. This excess creep strain is linearly dependent on stress up to the highest stress employed in this study (Fig. 4). Furthermore, Fig. 4 indicates that the higher the test temperature, the greater the excess creep strain due to transformation at a given stress.

Fig. 5 shows the tensile stress–strain curves of specimens tested at 170° C with and without concurrent transformation. It is seen that the flow stress remains steady in the case of specimens which do not undergo concurrent transformation. In the other case of concurrent transformation, the stress–strain curve exhibits a dip and its minimum flow stress is significantly lower (~ 4 MPa) than the steady state flow stress of specimens without concurrent transformation. About 10 to 25% more tensile elongation was observed in specimens undergoing concurrent transformation depending on the strain rate employed. In both

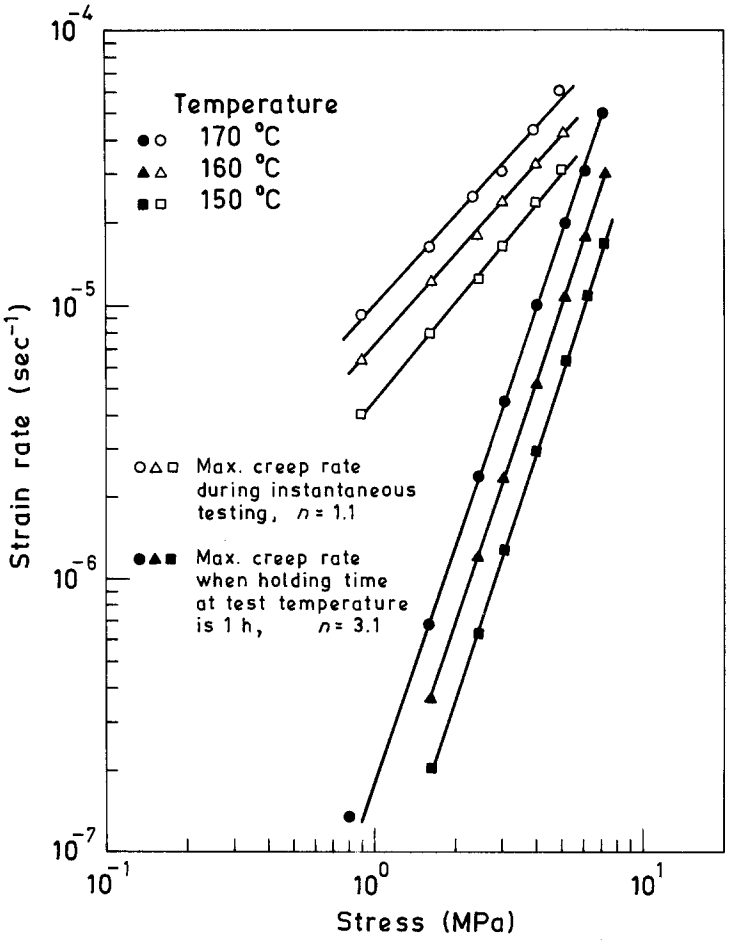

Figure 2 Stress dependence of creep rate of specimens up-quenched to different temperatures and tested immediately or after a holding time of 1 h.

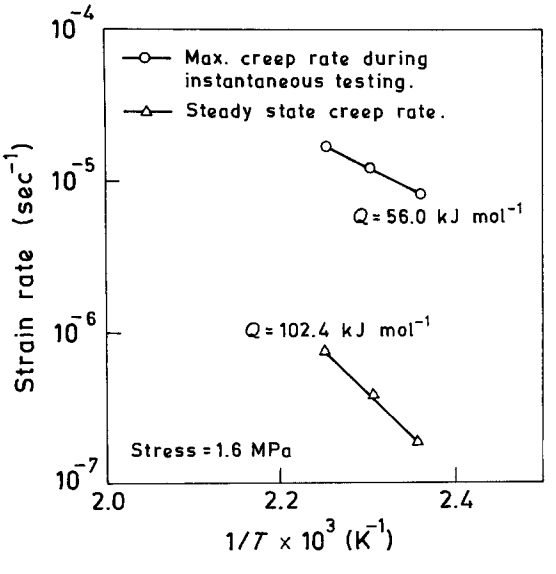

Figure 3 Arrhenius plots for the determination of activation energy for creep.

the types of specimens, intergranular fracture occurred with little necking.

An attempt was made to evaluate the strain rate sensitivity index ($m = \partial \ln \sigma / \partial \ln \dot{\epsilon}$) by means of the differential strain rate test. Increasing and decreasing orders of crosshead speeds were employed for this purpose. m decreased from 0.38 (at $\dot{\epsilon} = 3.2 \times 10^{-5}\,\text{sec}^{-1}$) to lower values with increasing strain rate in specimens tested at 170° C without concurrent transformation. In the case of specimens with concurrent transformation, the maximum m was only 0.22 under increasing crosshead speed conditions. On the other hand, anomalously high m (>1) values were obtained on following decreasing crosshead speed sequence.

4. Discussion

A holding time of 1 h above the eutectoid temperature of the Sn–Cd alloy would ensure completion of the reaction between the two phases leading

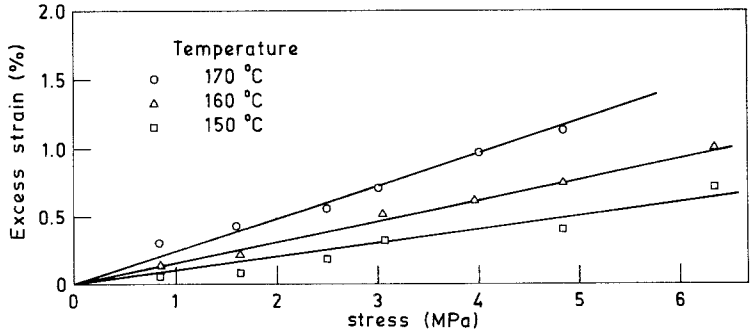

Figure 4 Stress dependence of excess creep strain due to transformation at different temperatures.

to the equilibrium phase. On the other hand, when the tests are started immediately after attaining the test temperature, the transformation from the two phases to the equilibrium phase is likely to continue up to several minutes during straining. The observed flow behaviour of the two types of specimens is in conformity with these expectations. The isothermal tests of the present study clearly indicate the softening effect of the concurrent transformation during plastic flow. The softening effect manifests as an increase in creep rate in creep tests and as a decrease in flow stress in constant crosshead speed tests. On completion of the transformation, the flow behaviour becomes similar to that of specimens held for 1 h at the test temperature. Since the flow stress under concurrent transformation dips below the steady state flow stress of the non-transformation case, the extent of softening due to transformation can, in general, be measured directly by isothermal constant crosshead speed tests of this type.

The strain rate sensitivity index (m) characteristic of transformation plasticity has been mostly determined in previous studies by thermal cycling around the phase transition temperature under a small load [10, 13]. In this procedure, $\dot{\epsilon}$ is obtained by dividing the strain per cycle by the cycle time and the temperature dependence of the constant K in the flow equation $\sigma = K\dot{\epsilon}^m$ is neglected. The maximum slope of the strip chart recording of deformation plotted against time obtained during thermal cycling at a constant load was also used for evaluating m [14]. While $m \approx 1$ is obtained usually in this manner, to what extent this is characteristic of transformation plasticity is not clear in view of the thermal cycling involved in this procedure. In our study, widely different values of m were obtained depending on strain, strain rate, temperature and the direction of change in crosshead speed. The low m values obtained by increasing the crosshead speed are due to drop in flow stress as a result of transformation, whereas overestimation of the load drops due to transformation in decreasing crosshead speed tests leads to unusually high m values. In view of the rapid change in flow stress with

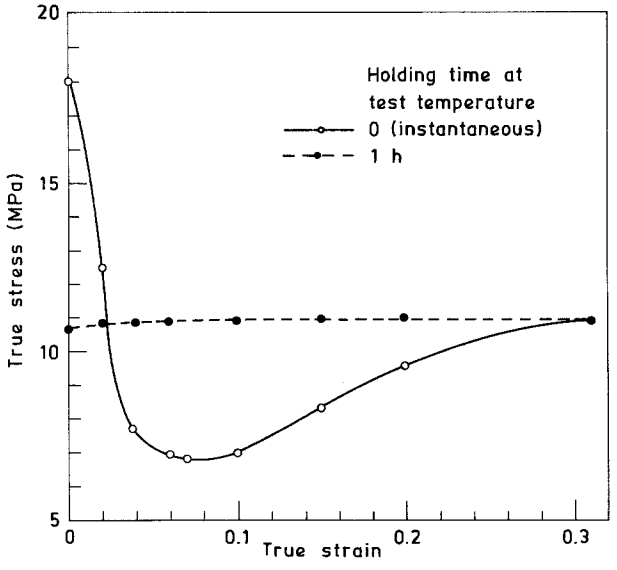

Figure 5 Tensile stress–strain curves of specimens up-quenched at 170° C and tested immediately or after a holding time of 1 h ($\dot{\epsilon} = 1.3 \times 10^{-4}\,\mathrm{sec}^{-1}$).

strain in specimens undergoing concurrent transformation, the evaluation of m by the differential strain rate test is of little significance in transformation plasticity studies.

The isothermal creep tests on specimens with and without concurrent transformation bring out the transformation plasticity effects in a convenient manner to assess the flow parameters of transformation plasticity. Our results indicate that the transformation effect manifests as a pseudo-steady state creep region in isothermal creep tests in which concurrent transformation takes place. From a study of this region as a function of stress and temperature, the stress exponent and activation energy for creep deformation uniquely representing the transformation effect can be evaluated and thereby operative micromechanisms of transformation plasticity can be assessed.

The observed activation energy for flow with concurrent transformation is significantly lower than that for steady state creep without concurrent transformation. It is, in fact, about one-half of the self diffusion activation energy of pure tin. This observation confirms the direct role of transformation-enhanced diffusion in transformation plasticity. Evidence for the generation of point defects by volume changes of transformation has been demonstrated by several investigators [15–18]. It has also been observed that transformation leads to an increase in the value of the diffusion coefficient by several orders of magnitude at the transformation front [19].

Enhanced diffusion by generation of excess point defects through transformation would be expected to have an activiation energy equal to the migration energy of point defects. The present value of stress exponent $n = 1.1$ for transformation plasticity is similar to that obtained by thermal cycling experiments in various materials. Furthermore, the stress exponent and activation energy values of the present study suggest that the operative mechanism for transformation plasticity is not likely to be dislocation climb with enhanced diffusion as proposed by Clinard and Sherby [5]. Further considerations of operative mechanisms are presented elsewhere [20].

5. Conclusions

1. The flow parameters of transformation plasticity in an Sn–Cd eutectoid alloy have been uniquely assessed through isothermal creep tests by up-quenching the alloy to above the eutectoid temperature and beginning the test immediately thereafter. A linear region of high creep rate was observed in the initial period of these tests with a stress exponent of 1.1 and an activation energy of 56 kJ mol^{-1}. The direct role of enhanced diffusion in transformation plasticity has been confirmed.

2. The lowering of flow stress as a consequence of phase transition has been assessed from isothermal constant crosshead speed Instron tests of specimens with concurrent transformation. The differential strain rate test is, however, of little interest in assessing the strain rate sensitivity in transformation plasticity.

References

1. G. W. GREENWOOD and R. H. JOHNSON, *Proc. Roy. Soc.* **A283** (1965) 403.
2. M. G. LOZINSKY and I. S. SIMEONORA, *Acta Metall.* 7 (1959) 709.
3. M. G. LOZINSKY, *ibid.* 9 (1961) 689.
4. I. A. ARBUZOVA and L. G. KHANDROS, *Phys. Met. Metallogr.* 17 (3) (1964) 68.
5. F. W. CLINARD and O. D. SHERBY, *Acta Metall.* **12** (1964) 911.
6. L. F. PORTER and P. C. ROSENTHAL, *ibid.* 7 (1959) 504.
7. S. B. AGARWAL and M. L. VAIDYA, *Scripta Metall.* 9 (1975) 447.
8. M. de JONG and G. W. RATHENAU, *Acta Metall.* 7 (1959) 246.
9. *Idem, ibid.* 9 (1961) 889.
10. D. OELSCHLAGOL and V. WEISS, *Trans. Amer. Soc. Met.* 59 (1966) 143.
11. G. R. YODER and V. WEISS, *Met Trans.* **3** (1972) 675.
12. W. A. BACKOFEN, I. R. TURNER and D. H. AVERY, *Trans. Amer. Soc. Met.* 57 (1964) 980.
13. R. A. KOT and V. WEISS, *Met. Trans.* **1** (1970) 2685.
14. J. R. SMYTH, R. C. BRADT and J. H. HOKE, *J. Amer. Ceram. Soc.* 58 (1975) 381.
15. F. W. CLINARD and O. D. SHERBY, *Trans. AIME* **233** (1965) 1975.
16. W. B. GRUPEN and G. SINES, *Acta Metall.* **13** (1965) 527.
17. F. H. FROES, R. W. K. HONEYCOMBE and D. H. WARRINGTON, *ibid.* **15** (1967) 157.
18. F. H. FROES and D. H. WARRINGTON, *Trans. AIME* **245** (1969) 2009.
19. K. SMIDODA, W. GOTTSCHALK and H. GLEITER, *Acta Metall.* **26** (1978) 1833.
20. M. M. SODHI, G. S. MURTY and M. L. VAIDYA, *Trans. Ind. Inst. Met.* **36** (1983) 321.

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