JOURNAL OF MATERIALS SCIENCE 13 (1978) 1849–1854

Anelasticity and creep transients in an austenitic steel

D. G. MORRIS*
Metallurgy Division, AERE, Harwell, Didcot, Oxon, UK

The extent of the anelastic strain imposed during the creep of a Type 316 stainless steel has been examined in detail and related to the structure of the dislocations within the material. The stress dependence of the anelastic strain recovery has been used to show that the intergranular carbide particles present gave rise to a small frictional stress. The accummulation of creep strain on reloading has been examined and shown to take place only after the anelastic strain has been re-imposed.

1. Introduction

Anelasticity during creep has been investigated fairly extensively and the principal mechanism of anelastic strain has been related to dislocation bowing under the influence of the effective stress operating on the dislocation links [1–4]. Since the anelastic strain is dependent on the effective stress, the applied stress dependence of the anelastic strain is a useful technique for the determination of the internal stress [5, 6]; this is particularly relevant in view of the recent criticisms [5, 7] of the more commonly used techniques for internal stress measurement [8–10].

An examination of anelasticity in a precipitation hardened Nimonic alloy [6] has demonstrated the existence of an internal stress which has to be exceeded to produce dislocation movement; furthermore, the creep rate observed depends on the effective stress which is less than the applied stress. However, a recent study of the creep in Type 316 stainless steel at 625° C [11] has shown that the creep rate is related principally to the distribution of dislocation link lengths, with little evidence of any internal stress effect.

The present paper describes a more detailed investigation of anelastic strain recovery on unloading Type 316 steel than was carried out previously [11] and also examines the creep transient on reloading. The analysis of such transients is of importance in interpreting cyclic creep behaviour which will be the subject of a later report.

2. Experimental

The composition of the Type 316 steel and the design and preparation of specimens have been detailed previously [12]. The specimens, with a 12 mm gauge length, were tested under constant stresses of 200 and 235 MPa at 625° C in the 1050° C solution treated state using a 50 kN capacity servotest servo-hydraulic machine. The primary and secondary stages of creep only were examined, and specimens were not taken to rupture. The load was controlled to within ±0.02kN, equivalent to ± 1 MPa on the specimen, and a strain resolution of 0.1 μm (less 10^{-5} strain) was possible. Constant stress creep testing was carried out after ramping the load to its maximum value over a 10 second period and then holding the load steady, making frequent slight reductions in the load level to maintain constant stress on the creeping specimens. Anelastic recovery was examined at various times during the primary and secondary stages of creep by instantaneously reducing the load to zero. Unload periods up to 100 sec were normally used, with occasional longer periods up to 2000 sec. Between unload periods the load was always imposed for times in excess of 30 min. In addition, the stress dependence of the anelasticity was determined by examining the effect of a range of stress increments. In all cases the stress was reduced from the creep stress (200 or 235 MPa) to a lower value.

Transmission electron microscopy was carried

*Now at: Institut CERAC, Chemin des Larges Pieces, Ecublens 1024, Switzerland.

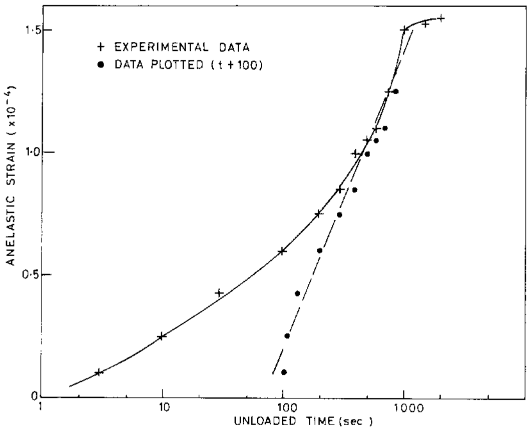

Figure 1 Time dependency of anelastic strain recovery on unloading from 235 MPa.

out on specimens taken to various stages throughout the creep life and precipitate and dislocation structures determined using the standard techniques described previously [11].

3. Experimental results

The typical time dependence of the anelastic recovered strain observed on completely unloading from 235 MPa is shown in Fig. 1; similar results were obtained on creeping at 200 MPa. The anelastic strain, ϵ_a, could be described by:

$$\epsilon_a = A \log (B + t) \qquad (1)$$

for unload times up to 1000 sec, where t is the duration of the unload period (sec) and B is a constant independent of the stage during the creep tests of the order of 100 sec. Saturation of the anelastic strain recovery occurred after about 1000 sec, with only a very slight increase in the anelastic strain recovery after this time.

The dependence of the anelastic strain recovery on the creep time is shown in Fig. 2. There was an initial rapid increase in the anelastic strain during the primary stage of creep, similar to that noted by Evans and Harrison [6], and a subsequent decrease to a minimum after about 100 h of creep followed by a gradual increase at longer times. The ratio of anelastic strain after 1000 sec unload time to elastic strain (noted during the instantaneous load removal) achieved a value of 0.13 after 25 h of creep and decreased to 0.09 after 100 h. The value of 0.13 is close to that predicted by Hesketh [13] for situations where particles have little effect on restricting dislocation movement.

The stress dependency of the anelastic recovered strain was examined after about 100 h of creep and the results are shown in Fig. 3. Whilst the slope of the stress increment–anelastic strain relationship decreased with increasing unload time, the data extrapolate to give an intercept of 25 ± 5 MPa such that the anelastic strain recovery was proportional to ($\Delta\sigma - 25$ MPa).

Equation 1 may thus be re-expressed as:

$$\epsilon_a = C(\Delta\sigma - 25) \log (B + t) \qquad (2)$$

where the stress increment, $\Delta\sigma$, is expressed in MPa, B is a constant independent of stress and stage during creep, and C is a constant dependent only on the temperature and the creep structure.

Dislocations and carbide densities at various stages throughout creep at a constant applied stress of 235 MPa are plotted in Fig. 4. There was

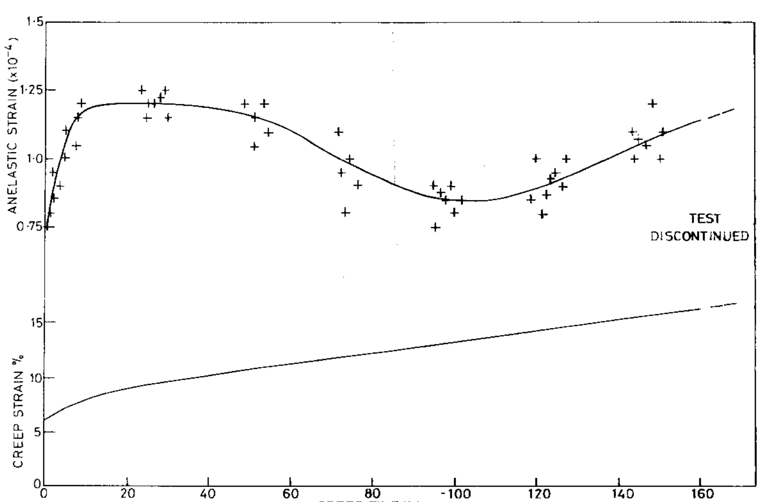

Figure 2 Anelastic recovered strain (after 300 sec load-off) and creep strain as a function of time for creep at 235 MPa and 625° C.

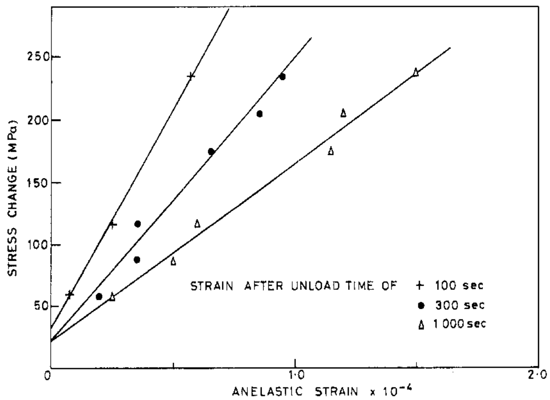

Figure 3 Stress dependence of anelastic strain observed after 100, 300, and 1000 sec unload time. Crept to 100 h at 625° C.

no evidence of intragranular carbide precipitation in the as-loaded state, but precipitation occurred during the creep test and the intragranular carbide density was a maximum after about 75 h on test; subsequently particle coarsening appeared to lead to a decrease in the intragranular precipitate density. The dislocation density appeared to increase from the as-loaded value to a maximum within 25 h. The fraction of mobile dislocations, calculated from an analysis of the dislocation link length distributions, as described in a previous paper [11] decreased rapidly from an as-loaded value of 0.44 to 0.2 after 25 h and remained constant at longer times. Similar increases in the total dislocation density and decreases in the mobile fraction during primary creep have previously been noted in an austenitic steel by Oden *et al.* [14].

The transient creep behaviour observed on reloading after unloading for 1000 sec is shown in Fig. 5, the strain and time origins being defined as the moment at which the applied stress reached 235 MPa. The rate of anelastic strain accumulation on loading was examined as described below: specimens were first allowed to recover all the anelastic strain by holding unloaded for 1000 to 2000 sec; the load was imposed for a short time (10 to 1000 sec), subsequently removed, and the total anelastic strain recovered over a further 1000 sec period. In this way it was shown that the rate of anelastic strain accumulation on loading was the same as the rate of anelastic strain recovery on unloading. Consequently, virtually all the strain observed in Fig. 5 during the first 1000 sec was recoverable anelastic strain, and there was very little unrecoverable creep deformation (shown dotted in Fig. 5) until almost all the anelastic strain had been imposed.

4. Discussion

4.1. Creep time dependence of anelasticity

The extent of anelastic strain that may be imposed during creep depends on several factors: (i) the total dislocation density and the fraction of mobile dislocations are both important since only the immobile dislocations can be considered to bow about the link ends; (ii) for a given dislocation density, the extent of bowing is a very sensitive

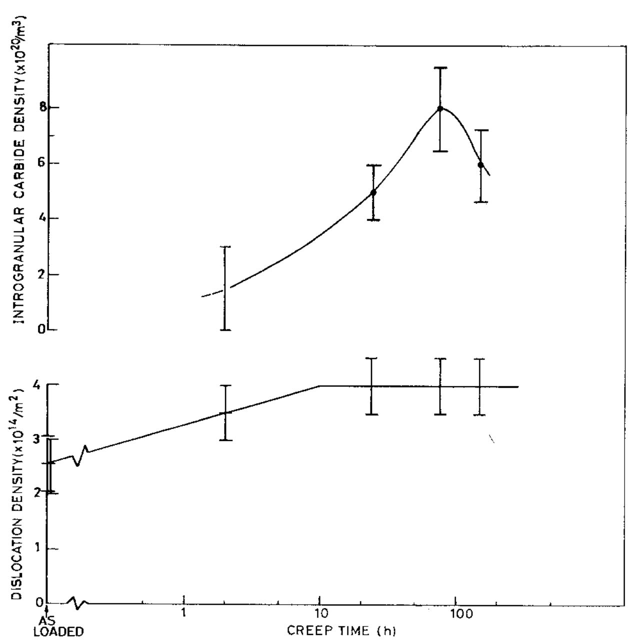

Figure 4 Dislocation and intragranular carbide density during creep at 235 MPa and 625° C.

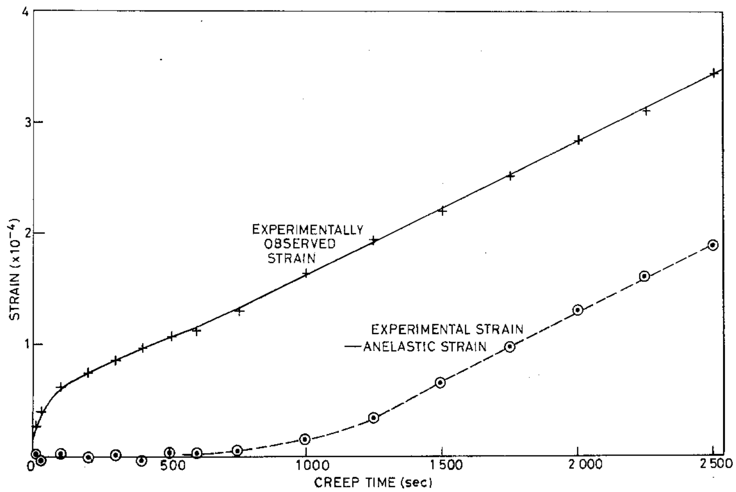

Figure 5 Strain observed on reloading after holding specimen unloading for 1000 sec, and observed strain with anelastic strain subtracted.

function of the free link length; extensive precipitation, leading to a reduction in the free link length, is thus very important; (iii) changes in the effective stress on the dislocation links lead to changes in the extent of anelastic strain and therefore, internal or friction stresses affect the anelastic strain. These effects may be illustrated by modifying Equation 2 as:

$$\epsilon_a = D\rho_i L^2(\Delta\sigma - \sigma_0)\log(B + t) \qquad (3)$$

showing the dependency of the anelastic strain on the immobile dislocation density, ρ_i, the dislocation link length, L, and the effective stress change. σ_0 is the internal stress and D a constant only on geometrical and elastic terms.

The initial increase in anelasticity observed during primary creep (Fig. 2) is caused by the decrease in the fraction of mobile dislocations; ρ_m/ρ, and the increase in the total dislocation density, ρ. The influence of precipitation on dislocation link length and effective stress change can be noted after 30 h creep straining and leads to the minimum in anelastic strain at about 100 h creep. At this stage the stress dependency of anelasticity indicates that a small friction stress is operating ($\sim$25 MPa). The change in effective stress, however, (235 to $\sim$210 MPa) is relatively small and it is likely that the decrease in dislocation link length by precipitation is responsible for the large decrease in anelastic strain observed. Creep for times in excess of 100 h leads to a reduction in intragranular particle density so that the influence of the particles on anelasticity diminishes.

4.2. The internal stress

The present, detailed studies of anelastic behaviour have shown that there is a small frictional stress affecting the dislocation links and creep behaviour. The influences of internal stress and distribution of dislocation link lengths in the analysis of creep rates have been considered in detail elsewhere [11] and the relative importance of the two factors discussed.

The particle-produced frictional stress calculated using the Orowan expression ($\sigma_{f\,max} = 2\mu b/\lambda$, where $\sigma_{f\,max}$ is the stress required to bow around hard, uncutable particles, μ is the shear modulus, b the Burgers vector, and λ the particle spacing) is approximately 150 to 200 MPa under the conditions relevant to the present tests. However, for softer particles, the friction stress is reduced to $\sigma_f = \sigma_{f\,max}\cos\phi/2$, where ϕ is the angle between the dislocation segments as the dislocation breaks through the particle [15]. It was difficult in the present study to measure ϕ because dislocation links often joined at, or close to, particles. However, there was no evidence to suggest that the intragranular carbide particles present in the steel were "hard", requiring extensive dislocation bending around the particles (low ϕ) before movement through or around the particles occurred. It does not seem unreasonable to propose that dislocation cutting through the carbide particles occurred relatively easily, corresponding to only slight dislocation bending, high breaking angles (e.g. $\phi \sim 150°$) and consequently much reduced friction stresses.

4.3. Loading transients

Studies of dislocation recovery (to be reported) have shown that annealing the creep structure without load for about 1000 sec at 625° C leads to dislocation link straightening (anelastic recovery) with little other changes in dislocation structure.

It is likely, therefore, that the very low creep rate observed initially on reloading (Fig. 5) is not caused by dislocation structure changes but is due to the unbent nature of the dislocation links. In terms of the theory of creep dislocation link breakage, it is apparent that the dislocation links cannot move until a high stress has been imposed on the link ends by extensive bowing; high stress is imposed on the dislocation nodes by bowing of the associated dislocation links and should not be considered to apply when the dislocation links are straight. Accordingly, the stress imposed on the link ends may be described by $\mu b/R$, where R is the radius of curvature of the associated dislocation segments, μ the shear modulus and b the Burgers vector. The creep rate may be determined using this model by (i) relating the amount of anelastic strain (ϵ_a) to the extent of bowing as $\epsilon_a/\epsilon_{a\,max} = R_{max}/R$, where the subscript max refers to the maximum achievable for a given applied stress and creep structure; (ii) deducing the stress applied to the link ends, $\mu b/R$; and (iii) calculating the rate of link activation under the operating stress and consequently determining the creep rate using the details of the dislocation link length distribution (as in [11]). It should be noted that this analysis ignores the small frictional stress term, which being relatively small will have a significant effect only at very low applied stresses.

The predicted and observed transient rates are compared in Table I and Fig. 6, and, bearing in mind the errors in the determination of the creep rate immediately on reloading, the agreement is reasonably good.

5. Conclusions

(1) The extent of recoverable anelastic strain during creep of a Type 316 steel at 625° C depends on the density of mobile and immobile dislocations, the effective stress, and the free dislocation link length. During primary creep, there is an increase in the anelastic strain due to an increase in the immobile dislocation density. Extensive carbide precipitation decreases the anelastic strain during secondary creep, principally by reducing the free dislocation link length since the effective stress is only slightly modified.

(2) Measurements of the stress dependency of the anelastic strain show that a small frictional stress (~25 MPa) operates during secondary creep. It is argued that this small frictional stress plays a relatively insignificant part in controlling the creep rate, the dislocation link length distribution being the primary factor involved.

(3) The creep rate is initially very low on reloading after recovering all the anelastic strain and

TABLE I Comparison of experimental creep rates during loading transient with values predicted assuming the stress operating on the dislocation link junctions to be related to the extent of anelastic link bowing

Time (sec)	Anelastic strain ($\times 10^{-4}$)	Forward transient strain ($\times 10^{-4}$)	Unrecoverable creep strain ($\times 10^{-4}$)	Experimental creep rate ($\times 10^{-8}$ sec^{-1})	Operative stress ($\mu b/R$) (MPa)	Creep ratio $\dot{\epsilon}/\dot{\epsilon}_{235}$ ($\dot{\epsilon}_{235} = 1$)	Predicted creep rate ($\times 10^{-8}$ sec^{-1})
10	0.25	0.27	0.02	NS*	40	NS	NS
30	0.43	0.4	−0.03	NS	65	NS	NS
100	0.6	0.62	0.02	NS	91	NS	NS
200	0.75	0.75	0	NS	114	NS	NS
300	0.85	0.87	0.02	NS	129	NS	NS
400	1.0	0.98	−0.02	0.01–0.1	152	0.0005	0.006
500	1.05	1.07	0.02	0.2	159	0.002	0.03
600	1.1	1.12	0.02	0.6	167	0.007	0.09
750	1.25	1.3	0.05	2.0	190	0.04	0.5
1000	1.5	1.65	0.15	5.0	227	0.5	6.5
1500	1.55	2.2	0.65	13	235	1	13
2000	1.55	2.85	1.3	13	235	1	13
2500	1.55	3.45	1.90	13	235	1	13

*NS = negligibly small.

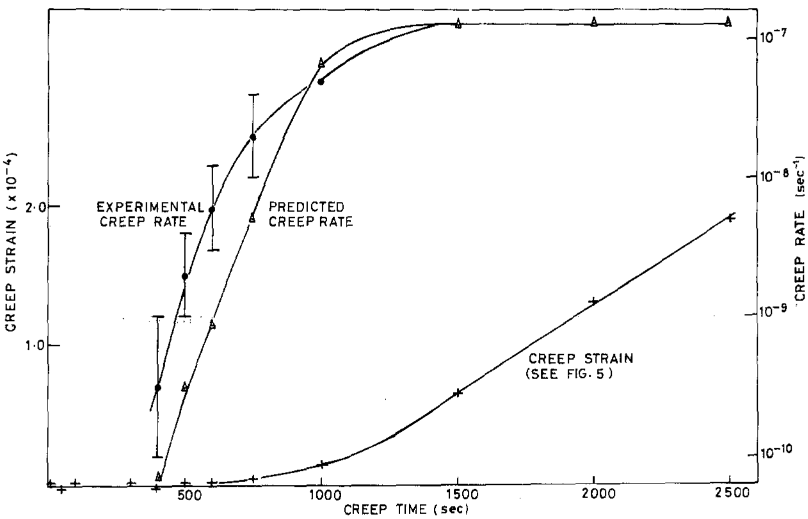

Figure 6 Creep strain and creep rates during the loading transient. Obtained on reloading to 235 MPa after holding the specimen unloaded for 1000 sec.

remains so until the anelastic strain has been reimposed. This behaviour is interpreted on the basis that dislocation links are capable of breaking only after they have become fully bowed.

Acknowledgements
The author gratefully acknowledges critical comments from Dr D. R. Harries.

References

1. C. ZENER, "Elasticity and Anelasticity in Metals" (Chicago University Press, 1948).
2. A. V. GRANATO and K. LUCKE, *J. Appl. Phys.* **27** (1956) 583.
3. J. H. GITTUS, *Scripta Met.* **6** (1972) 247.
4. *Idem, ibid.* **6** (1972) 481.
5. G. J. LLOYD and R. J. McELROY, *Phil. Mag.* **32** (1974) 231.
6. W. J. EVANS and G. F. HARRISON, *Scripta Met.* **9** (1975) 239.
7. G. J. LLOYD and R. J. McELROY, *Phil. Mag.* **32** (1975) 231.
8. J. C. M. LI, *Canad. J. Phys.* **45** (1968) 483.
9. G. B. GIBBS, *Phil. Mag.* **13** (1966) 317.
10. A. A. SOLOMON and W. D. NIX, *Acta. Met.* **18** (1970) 863.
11. D. G. MORRIS, *ibid.* (in press).
12. D. G. MORRIS and D. R. HARRIES, *J. Mater. Sci.* **12** (1977) 587.
13. R. V. HESKETH, *Phil. Mag.* 8 (1963) 1321.
14. A. ODEN, E. LIND and R. MAGNEBORG, Proceedings of the International Conference on Creep Strength in Steel and High Temperature Alloys, Sheffield, 1972.
15. A. J. E. FOREMAN and M. J. MAKIN, *Phil. Mag.* **14** (1966) 91.

Received 18 October and accepted 15 December 1977.