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The Kinetics of Cavity Growth in 20 Cr/25 Ni Stainless Steel

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The development of cavitation during creep has been examined in an unstabilised 20 Cr/25 Ni austenitic stainless steel, using precise density measurements as an indicator of the level of cavitation. The change in density was proportional to the duration of creep to the power of 3.0 increasing to 4 to 5 just before fracture, and this time exponent was not affected by either grain size or irradiation. Metallographic examination showed that wedge-cracks were the predominant mode of cavitation and helium bubble growth in irradiated specimens increased the rate of crack propagation, thus reducing the ductility. A simple model for the growth of triple-point cracks is used to explain the experimentally observed changes in density.

1. Introduction

Irradiation reduces the ductility of austenitic steels at high temperatures and generally causes intergranular failure. This embrittlement has been associated with the presence of helium formed by $^{10}B\ (n, \alpha)^7\ Li$ reactions with thermal neutrons. Because helium is very insoluble in austenite, the helium precipitates as bubbles when irradiated material is heated above 0.5 T_m and this often occurs preferentially at grain boundaries, partly since this is energetically favourable and partly because the boron tends to concentrate near grain boundaries.

In an unstressed material at equilibrium, the pressure P_0 inside a bubble of radius r_0 balances the surface tension forces $2\gamma/r_0$, where γ is the surface energy. Hyam and Sumner [1] showed that when a stress σ is applied, the bubble will grow to a radius r by vacancy condensation according to the equation:

$$\sigma = \frac{2\gamma}{r}\left(1 - \frac{r_0^2}{r^2}\right) .$$

This equation becomes a maximum when $r = \sqrt{3}r_0$ and consequently

$$\sigma_m = \frac{2}{3\sqrt{3}}\left(\frac{2\gamma}{r_0}\right) .$$

The bubble can then grow spontaneously and the stress to cause further growth will decrease. A number of investigations [2–5] have shown that growth can occur by this mechanism in slow tensile testing at temperature, although in some cases coalescence of bubbles may be required for the bubble to reach a critical size. Waddington [3] and Waddington and Lofthouse [6] showed that wedge-cracking was more common than rounded cavities although the growth of helium bubbles in irradiated specimens accelerated the rate of propagation of the crack.

A creep crack will grow in a stable manner provided $\sigma\, na < 2\gamma$ where na is the wedge displacement, and an analysis by Williams [7], applicable to large-grained materials, assumes that failure occurs when the crack length is equal to the grain-edge length. Waddington [3] has shown that the model for stable crack growth is applicable to a large-grained 20 Cr/25 Ni stainless steel. However, in fine-grained materials crack lengths are frequently several facets long well before fracture [8] and they appear to propagate rapidly across a facet until held up temporarily at the next triple point. Lindborg [9] suggests that these cracks grow in an unstable manner (i.e. $\sigma\, na = 2\gamma$) and the time to fracture will depend on the largest crack reaching an infinite size. Clearly, both the mode of crack growth and the criterion for fracture are uncertain and may conceivably vary with grain size for a given material.

The 750° C tensile ductility of 18 Cr/10 Ni/Nb stainless steel after irradiation at 50° C is

*Now at Broken Hill Proprietary Co Ltd, Melbourne Research Laboratories, Clayton, Victoria, Australia independent of the boron content [10] although after irradiation at 650° C the tensile ductility at 750° C decreases with increasing boron [5]. In the former case cracking was generally fairly uniform whereas after high-temperature irradiation the cracks were more localised to regions adjacent to the final fracture. In both cases the growth of helium bubbles during straining probably affects the rate of crack propagation. Helium atoms produced by irradiation at high temperature will diffuse rapidly until absorbed by a bubble. The concentration of helium in solution at any time will therefore be relatively small so that helium may diffuse over relatively large distances before entering a bubble and this may enable more helium to reach the grain boundary during irradiation. However, when a material is heated after irradiation at low temperatures, the degree of supersaturation is high, so the distance helium diffuses before coalescing with other helium atoms will be less. A higher concentration of helium bubbles on the grain boundary may be critical and lead to rapid propagation of an unstable crack.

In this investigation, experiments were carried out on an unstabilised 20 Cr/25 Ni stainless steel similar to that used by Waddington [3] to examine the effect of low-temperature irradiation, grain size and stress on the kinetics of cavity growth during creep at 750° C.

2. Materials and Methods

Table I shows the composition of the steel used in this investigation together with the Nb:(C + N) ratio. The ratio is less than stoichiometric for either NbC (7.7:1) or Nb_4C_3 (10.3:1) so that chromium carbides will be present.

Cylindrical specimens with a gauge length of 30 mm and a diameter of 5.7 mm were machined from a cold-worked bar and then heat-treated in a vacuum at either 1000° C for 1 h or 1250° C for 2 h followed by air-cooling, to give grain sizes of 30 and 250 μm respectively. The specimens were then aged at 820° C for 20 h before creep-testing. Several specimens were irradiated at 40° C to a neutron dose of 6.5×10^{19} n cm^{-2} (thermal) and 2.9×10^{19} n cm^{-2} (fission) and then annealed at 750° C for 25 h to precipitate helium.

TABLE I Chemical Composition (wt %) of steel

Chromium	20.0
Nickel	24.7
Carbon	0.015
Niobium	0.05
Manganese	0.01
Silicon	0.05
Boron	0.001
Nitrogen	0.002
Sulphur	0.009
Phosphorus	0.007
Titanium	0.005
Aluminium	0.011
Nb/(C + N) ratio	3:1

Creep tests at constant load were carried out at 750° C in a vacuum of 2×10^{-5} torr and during testing the approximate strain was measured on a dial gauge mounted on the beam. The strain was later measured accurately after removal of the specimen from the apparatus so that the true creep curve could be interpolated from the dial gauge readings.

The specimen was cooled from the creep temperature under load after regular increments of strain, and the fractional change in density $\Delta D/D$ between the specimen and an identical control determined, using the technique described previously [11]. Density measurements provide an estimate of the total volume of cavities formed during creep and although cavitation is confined to the gauge length, the true level of cavitation can be estimated by multiplying the observed $\Delta D/D$ by the ratio of the total specimen weight to the weight of the gauge length. The sensitivity in measurement is reduced to 3×10^{-5} by not cutting out the gauge length before determining the density but the same specimen can be tested to fracture, thus eliminating sampling errors.

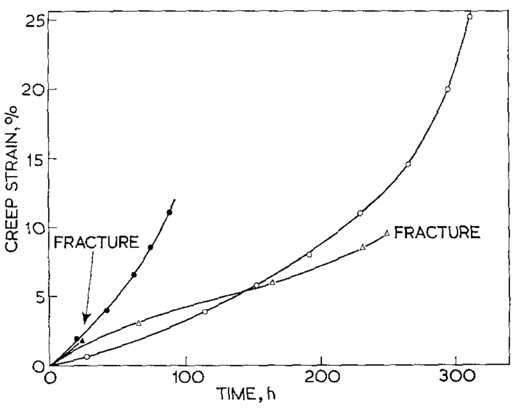
To ensure that $\Delta D/D$ was not affected by ageing at 750° C, the control-specimen was mounted alongside the specimen during creep and frequently the density of this furnace control-specimen was compared with a room temperature control-specimen after different creep times. It was found that $\Delta D/D$ between these controls was invariably $< 2 \times 10^{-5}$ so that the observed values of $\Delta D/D$ indicate only the total volume of cavities formed during creep.

Surface measurements of the vertical component of grain boundary sliding, d, were made on some of the specimens using interferometry at low strains and differential focusing on either side of the boundary at higher strains. Approximately 100 measurements of d were made so that the mean value was significant to $\pm$ 10% at a 90% level of confidence.

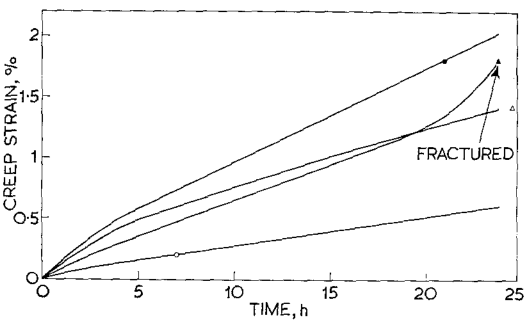
3. Results

3.1. Effect of Grain Size and Irradiation on Creep Rate

Figs. 1a and b show the creep curves for both fine-grained (30 μm) and coarse-grained (250 μm) 20 Cr/25 Ni stainless steels, tested at 750° C and 41.5 MN m^{-2}. There was no appreciable primary creep in the fine-grained material as the creep rate increased continuously with time. Increase in solution-treatment temperature to increase the grain size resulted in a conventional three-stage creep curve although there was no significant difference in the creep rate.



(a)



(b)

Figure 1 (a) Creep curves for 20 Cr/ 25 Ni unstabilised stainless steel at 750° C 41.5 MN m^{-2}. (b) Initial part of creep curves for 20 Cr/25 Ni unstabilised stainless steel at 750° C 41.5 MN m^{-2}. ○ unirradiated fine-grain size; △ unirradiated large-grain size; ● irradiated fine-grain size; ▲ irradiated large-grain size.

Creep curves for large- and fine-grained irradiated specimens are also shown in fig. 1. Irradiation reduced the creep ductility from > 25% to 11.3% in the fine-grained specimens and from 9.6% to 1.8% in the coarse-grained specimens. Although irradiation had no effect on the creep rate in the coarse-grained specimens, the creep resistance of fine-grained specimens was reduced by irradiation. The effect was confirmed in creep tests repeated under identical conditions.

3.2. Effect of Grain Size and Irradiation on Density During Creep

The points on the creep curves show the strains at which the creep test was interrupted and the fractional change in density determined. The irradiated specimens were annealed at 750°C before creep to precipitate helium from solution. To ensure that precipitation was complete, one of a pair of identical specimens was annealed in vacuum at 750° C; then $-\Delta D/D$ was determined. The results shown in table II indicate that all the helium is precipitated by a 1 h anneal at 750° C and that both specimens contain the same amount of helium. Specimen 2 was subsequently used as the control for specimen 1 so that the contribution of helium bubbles to the density change is not shown in the following graphs. However, this contribution is relatively small compared with the total density change.

TABLE II Density measurements on irradiated specimens before creep

Heat-treatment after irradiation: Specimen 1	Specimen 2	$\frac{D_1 - D_2}{D}$
None	none	$+ 0.39 \times 10^{-4}$
1 h 750° C, AC	none	$- 4.12 \times 10^{-4}$
25 h 750° C, AC	none	$- 4.24 \times 10^{-4}$
25 h 750° C, AC	25 h 750° C, AC	$+ 0.02 \times 10^{-4}$

Fig. 2 shows the relationship between $-\Delta D/D$ and creep time for the four specimens. The gradient $\partial(\ln - \Delta D/D)/\partial(\ln t) = 3$ for both unirradiated and irradiated samples and this relationship is maintained for three specimens until there is an appreciable acceleration of the creep rate in the tertiary stage. The value of $-\Delta D/D$ was independent of grain size in the unirradiated material, and this appears to be true also for the irradiated material, although $-\Delta D/D$ was only measured on the large-grained specimen at fracture. This point lies slightly above the line for the fine-grained irradiated specimen, which can be attributed to the acceleration of cavity growth during tertiary creep.

Fig. 3 shows that the strain exponent $\partial(\ln - \Delta D/D)/\partial(\ln \epsilon)$ for a particular specimen is

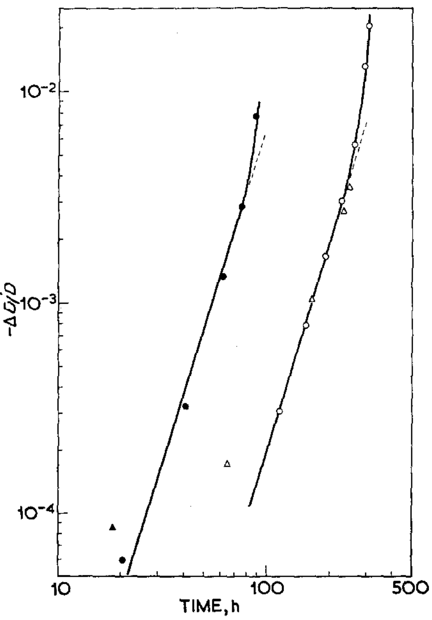


Figure 2 $\Delta D/D$ vs. time for 20 Cr/25 Ni unstabilised stainless steel at 750° C 41.5 MN m^{-2}. ○ unirradiated fine-grain size; △ unirradiated large-grain size; ● irradiated fine-grain size; ▲ irradiated large-grain size.

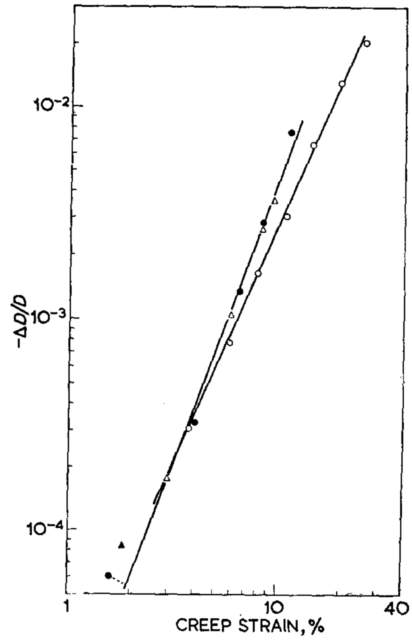


Figure 3 $\Delta D/D$ vs. strain for 20 Cr/25 Ni unstabilised stainless steel at 750° C 41.5 MN m^{-2}. ○ unirradiated fine-grain size; △ unirradiated large-grain size; ● irradiated fine-grain size; ▲ irradiated large-grain size.

constant up to fracture but increases with grain size and irradiation. Extrapolation to low values of $-\Delta D/D$ indicates that cavities are nucleated at strains of $< 1\%$, and fig. 4 shows how this damage is accumulated progressively during creep.

3.3. Effect of Stress on Cavity Growth

Fig. 5 shows that $\partial(\ln - \Delta D/D)/\partial(\ln t) = 3$ over the stress range 41.5 to 62 MN m^{-2}, provided the creep rate was approximately constant, although the strain exponent $\partial(\ln - \Delta D/D)/\partial(\ln \epsilon)$ increases from 2.24 to 2.54 with increase in stress (fig. 6). The level of cavitation at a given strain decreases with increase in stress with the result that the ductility is enhanced.

3.4. Effect of Heat-treatment after Irradiation on Creep Properties.

A second large-grained irradiated specimen was given the same heat-treatment after irradiation (2 h, 1250° C, 20 h 820° C, AC) and then tested under similar conditions to the first large-grained specimen. The results from these two specimens are compared in table III.

TABLE III Effect of post-irradiation annealing on creep properties of unstabilised 20Cr/25Ni stainless steel at 750° C, 42 MN m^{-2}

Initial condition	Annealing after irradiation, before creep	$\Delta D/D$ during anneal	Ductility	Time to fracture, h	$\Delta D/D$ during creep
2 h 1250° C, 20 h 820° C, irradiated	24 h 750° C	-4.0×10^{-4}	1.83%	19.5	-4×10^{-4}
As above	2 h 1250° C 20 h 820° C	-13.4×10^{-4}	1.1%	3.5	not significant

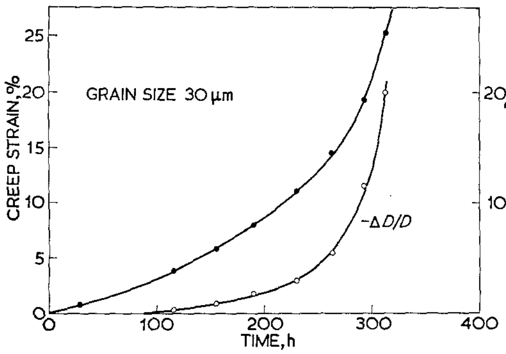


Figure 4 Variation of creep strain and $\Delta D/D$ with time for unstabilised 20 Cr/25 Ni stainless steel at 750° C 41.5 MN m^{-2}.

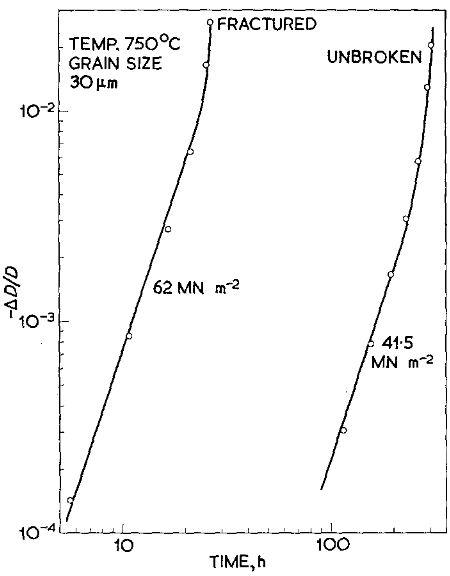


Figure 5 Effect of stress on $-\Delta D/D$ at 750° C for unstabilised 20/25 stainless steel.

Clearly, both the ductility and the time to fracture are significantly reduced by annealing at high temperature after irradiation, and intergranular fracture can occur without a significant change in density.

3.5. Metallographic Observations

Because density measurements were made on the same specimen up to fracture, metallographic observation was only possible at the end of the test. Figs. 7 and 8 show that in both irradiated and unirradiated fine-grained specimens, wedge-cracks were more common than rounded cavities, and this was also true of the large-grained specimens. Figs. 9a to e shows a series of

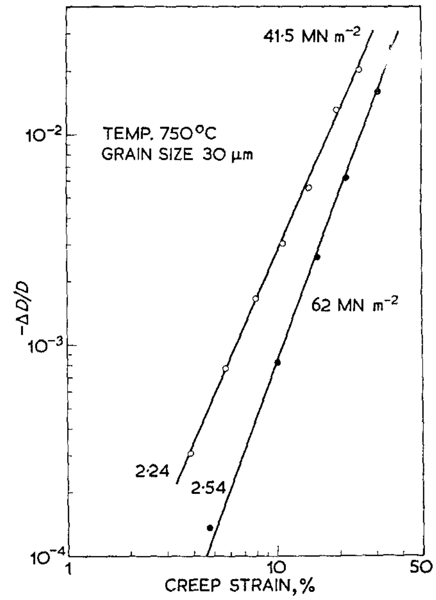


Figure 6 $\Delta D/D$ vs. creep strain for 20/25 unstabilised stainless steel.

SEM pictures from the fracture surface of unirradiated and irradiated large-grained specimens. The amount of ductile tearing associated with fracture decreases as the specimen ductility decreases. In the irradiated specimens (figs. 9b to e) there is an almost continuous network of grain-boundary cavities 1 to 5 μm in diameter that have nucleated on helium bubbles. In the irradiated specimen given the second heat-treatment at 1250° C, the bubble density on the boundary was much higher although the amount

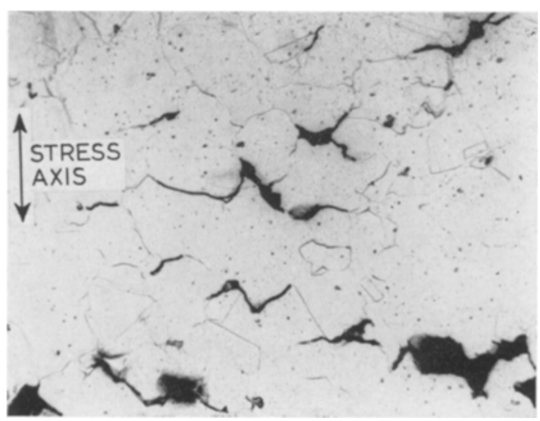


Figure 7 Longitudinal section from unirradiated 20 Cr/25 Ni unstabilised stainless steel strained 25% at 750° C 42 MN m^{-2} (× 258).

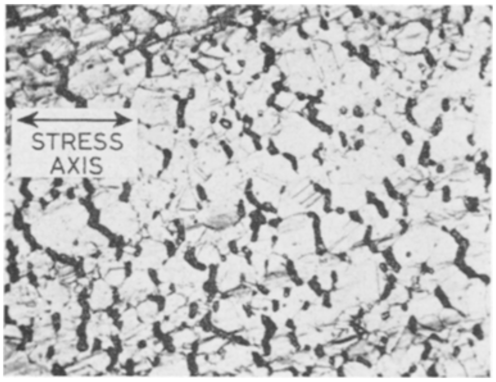


Figure 8 Longitudinal section from irradiated 20 Cr/25 Ni unstabilised stainless steel strained 11.3% at 750° C 42 MN m^{-2} (× 103).

of bubble growth during creep was apparently lower (compare figs. 9c and e). In this specimen ductile tearing was limited to the corners of the grain boundary facets (fig. 9d). If the original bubble size after irradiation was 150 to 250 Å [2], then on the fracture surface at least considerable growth must have occurred during creep.

3.6. Stability of Cavities in Unstabilised 20 Cr/25 Ni Stainless Steel

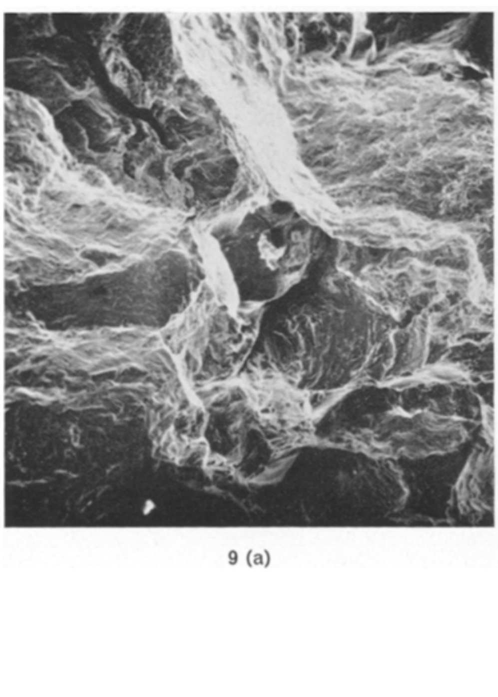
Using a 20 Cr/25 Ni/Nb stabilised stainless steel containing a large amount of nitrogen, Waddington and Evans [12] have found that intermittent annealing at the creep temperature both reduced the cavity volume and increased the ultimate ductility. In the present investigation, a large-grained unirradiated specimen strained 3% at 750° C and then annealed for 25 h at 750° C showed no change in density during the anneal. A similar result was obtained after 6% strain, showing that the cavities formed are stable during annealing and consequently the elongation at fracture, 9.8% was approximately the same as for a specimen strained continuously to fracture.

3.7. Observations of Grain Boundary Sliding

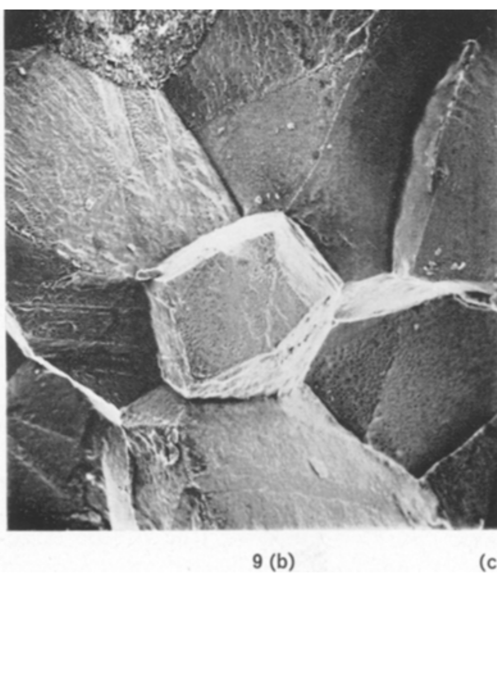
Measurements of the vertical component of sliding are shown as a function of both time and creep strain in fig. 10. Sliding increases linearly with time for most of the life when the creep rate is approximately constant. Extrapolation to zero time or strain shows that there is a period of either primary or instantaneous sliding although no measurements were made at low strains.

4. Discussion

For both large- and fine-grained specimens the rate of cavity growth $\partial\,(\ln - \Delta D/D)/\partial(\ln t) = 3$ until the creep rate accelerates appreciably into tertiary. Fig. 2 shows that the time exponent was not changed by irradiation and it follows that, at constant stress and temperature, the total cavity volume $V_{\sigma,T} = Kt^3$ where K is a constant, independent of grain size but increasing with irradiation. After a given creep time the value of K is 27 times greater in the specimen irradiated



9 (a)



9 (b)

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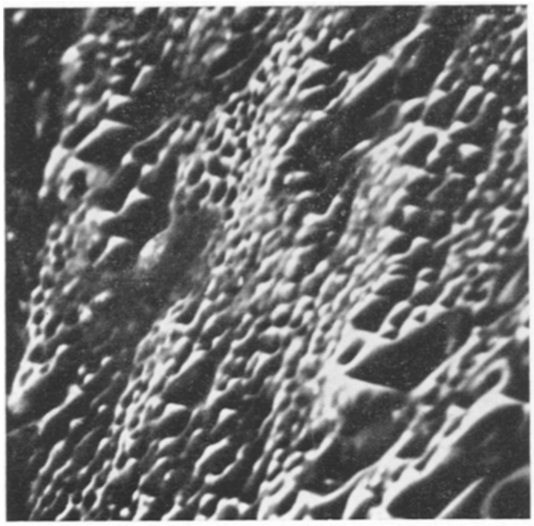
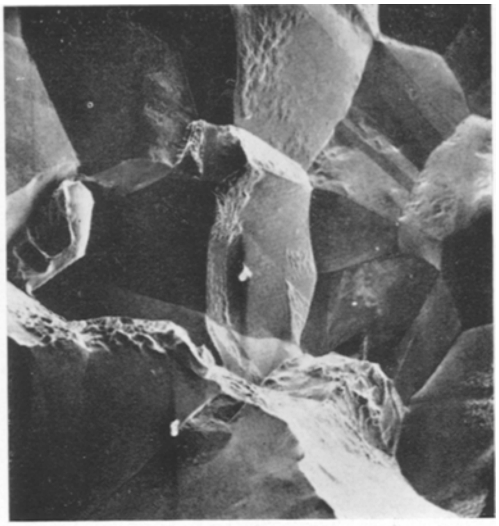
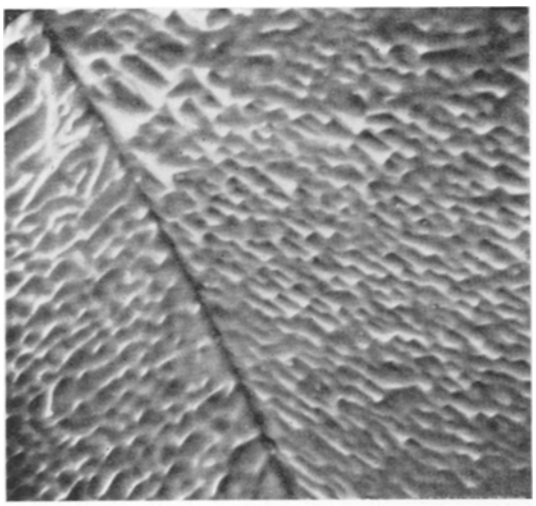


9 (c)



9 (d)



9 (e)

Figure 9 Scanning electron micrographs from the fracture surfaces of large-grained 20 Cr/25 Ni unstabilised stainless steel: (a) Unirradiated. Ductility 9.6% (× 140). (b) Irradiated. Solution-treated before irradiation only. Ductility 1.83% (× 140). (c) As (b) showing cavities that have nucleated on helium bubbles (× 6000). (d) Irradiated. Solution-treatment and ageing repeated after irradiation. Ductility 1.10% (× 140). (e) As (d) showing typical region from smooth grain boundary area (× 2800).

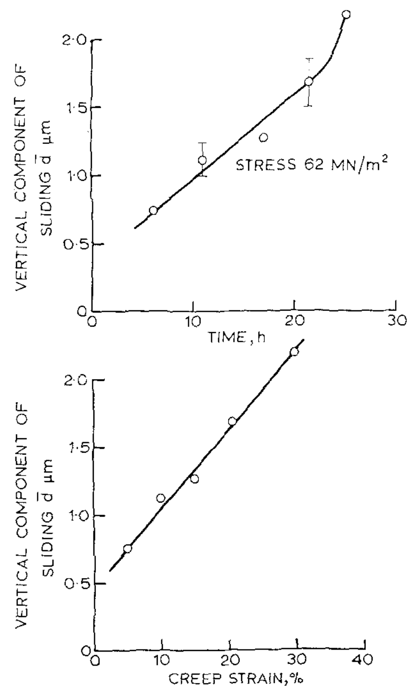


Figure 10 Grain boundary sliding in 20 Cr/25 Ni stainless steel at 750° C Min. creep rate 10^{-2} h^{-1}.

to a thermal neutron dose of 6.5×10^{19} n cm^{-2}.

The relationship between V and t up to the end of secondary is not the same for all materials, and time exponents in the range 1.5 to 2.5 have been reported. Woodford [13] and Greenwood [14] have suggested independently that the total cavity volume is determined by both time and strain in an equation of the form:

$$V \propto \epsilon t \quad (1)$$

This equation can be taken to indicate that the number of voids, n, is dependent on strain and the volume of an individual void is proportional to time. For example Gittins [15] found that for copper at 400° C both n and ϵ were $\propto t^{0.5}$ and consequently $n \propto \epsilon$ [14]. Also since $V \propto t^{3/2}$ the volume of an individual void $V_I \propto t$. This is consistent with diffusion-controlled growth of cavities [17]. Neither Hull and Rimmer's analysis [16] nor the subsequent modifications by Speight and Harris [17] predict growth rates $\propto t^3$, even if it is assumed that the nucleation rate $\propto t$. Therefore, over the range of creep rates used, vacancy diffusion would not appear to be important to growth, and this is compatible with metallographic observations which showed that most of the damage was in the form of wedge-cracks. Waddington and Lofthouse [6] also found that wedge-cracking was more common than rounded cavities in constant strain rate tests on the same material, and that the rate of crack growth was increased by irradiation with a consequent lowering of ductility. It is necessary therefore to explain the high time exponent indicated by density measurements by considering the growth of wedge-cracks.

A triple-point wedge-crack is formed when the stress concentration resulting from sliding on one or two boundaries is relieved by cracking on a boundary nearly normal to the stress axis. The crack will be forced to grow by wedging action as more sliding is fed into the triple point, and for a crack nucleated at $t = 0$, Williams [7] has suggested:

$$na = L\dot{\epsilon}_{mcr}t \tag{2}$$

where na is the wedge opening, L the grain size and $\dot{\epsilon}_{mcr}$ the minimum creep rate. Implicit in this relationship is that the rate of sliding $\dot{\epsilon}_{gb}$ is constant, as is observed (fig. 10a). The length of stable crack formed by this wedging action has been derived by Williams using Cottrell's model [18] for a wedge-crack formed by shear on intersecting planes. The rate of crack growth is given by:

$$\frac{dc}{dt} = \frac{\mu L^2 \dot{\epsilon}_{mcr}^2 t}{4\pi(1-\nu)\gamma} \tag{3}$$

where μ is the shear modulus, ν is Poisson's ratio and γ the effective fracture energy. Integration of equation 3 gives $c \propto t^2$, and since the volume of the crack formed depends on the product of c and na, its volume v is given by

$$v \propto \frac{\mu L^3 \dot{\epsilon}_{mcr}^3 t^3}{8\pi(1-\nu)\gamma} \tag{4}$$

i.e.

$$v = KL^3 \dot{\epsilon}_{mcr}^3 t^3 \tag{5}$$

If the fraction of grain-boundaries on which wedge-cracks form is independent of grain size, then the number of cracks per unit volume $\propto 1/L^3$ and therefore the total volume of cracks per unit volume

$$V = K' \dot{\epsilon}_{mcr}^3 t^3 . \tag{6}$$

From this equation, the contribution of helium bubbles to the growth of triple-point cracks may be deduced by comparing both the creep rates and the times to reach a given level of cavitation for irradiated and unirradiated specimens. The value of the constant for irradiated specimens $K_I' = 2.3\, K'_{UI}$ so that although $V_I = 27 V_{UI}$ after a given time (fig. 2) most of this increase is because the creep rate of the irradiated material is higher. However, the irradiated specimen does accumulate crack volume at a significantly higher rate per unit strain (fig. 3), and this, together with the likelihood that the critical crack length is lower in the irradiated specimen, could account for the reduction of ductility by a factor of 2.3 by irradiation.

Irradiation reduces the ductility of large-grained specimens by the larger factor of 6 probably because as the grain size is increased a higher area fraction of grain boundary is occupied by helium bubbles. This follows because in low boron austenitic steels the boron and hence the helium formed by the transmutation of the B^{10} isotope, is concentrated near grain boundaries [10]. Consequently by increasing the grain size the total grain-boundary area is reduced and a similar concentration of helium is confined to this smaller area, enabling the crack to grow more quickly.

By further heat-treating after irradiation an even larger amount of helium accumulates on the grain boundaries, either by migration of the bubbles to the boundary by Brownian movement or more probably by the sweeping action of a migrating grain boundary. This larger concentration of helium reduces both the ductility and the time to fracture appreciably; the boundaries being so densely covered with helium bubbles that fracture can occur without much overall change in density.

There is considerable evidence [2, 3, 5] that helium bubbles on grain boundaries grow during testing and although observations on the fracture surface confirmed this, some of the growth may have been induced by the stress field ahead of the propagating crack [19]. V_I is therefore made up of both the growth of helium bubbles and cracks, but since both occur preferentially on boundaries normal to the stress axis most of the enlarged bubbles are eventually absorbed by a crack, thereby accelerating the rate at which it propagates.

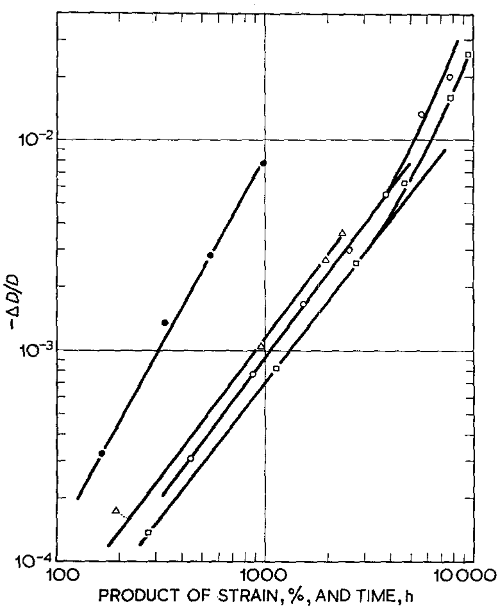


Figure 11 $\Delta D/D$ vs. product of strain and time for 20/25 unstabilised stainless steel at 750° C. ○ unirradiated fine-grain size 41.5 MN m^{-2}; □ unirradiated fine-grain size 62 MN m^{-2}; △ unirradiated large-grain size 41.5 MN m^{-2}; ● irradiated fine-grain size 41.5 MN m^{-2}.

The effect of irradiation on cavitation is also shown in fig. 11 in which the separate effects of strain and time are normalised by plotting ϵt vs $-\Delta D/D$. Fig. 11 also shows that the product ϵt for a given density change increases with decrease in grain size and increase in stress. This is not predicted by the analysis since by integrating equation 3 and assuming fracture occurs when c = grain-edge length, Williams [7] showed that for wedge-crack growth $\dot{\epsilon}_{mcr}\, t_{Fr}$ = constant, independent of stress. Unfortunately, this relationship cannot be verified for 20 Cr/25 Ni steel because there was no true secondary creep rate, although in other materials there is evidence of a slight stress and grain size-dependence [20]. In a recent analysis, Lindborg [9] considered that a creep crack grew the length of one facet and was then temporarily arrested, while sliding initiated a crack on an adjacent facet. The rate of crack growth is given by:

$$\frac{dc}{dt} = \left(\frac{\sigma L}{0.4\gamma}\right)^{0.7} \dot{\epsilon}_{gb} \tag{7}$$

Substituting $\dot{\epsilon}_{gb} = L\dot{\epsilon}_{mcr}$ and integrating, we obtain

$$c = \left(\frac{\sigma}{0.4\gamma}\right)^{0.7} L^{1.7}\dot{\epsilon}_{mcr}t \tag{8}$$

Since v $\propto$ cna and $na = L\dot{\epsilon}_{mcr}t$, it follows that

$$V = K' \frac{\sigma^{0.7}}{L^{0.3}}\, \dot{\epsilon}_{mcr}{}^2 t^2 \tag{9}$$

Although this equation does not give the correct creep rate and time-dependence of V, it predicts that V is slightly stress- and grain size-dependent.

The strain exponent $\partial(\ln - \Delta D/D)/\partial(\ln\epsilon)$ increases with increase in stress (fig. 6) and this is because the stress sensitivity of the sliding rate is different from the stress sensitivity of the creep rate [21].
i.e.

$$\dot{\epsilon}_{gb} = A'\sigma^m ,$$

$$\dot{\epsilon}_{mcr} = A''\sigma^n ,$$

and therefore

$$\dot{\epsilon}_{gb}/\dot{\epsilon}_{mcr} = A\sigma^{m-n} .$$

Since the contribution of sliding to the total creep strain, and hence $\dot{\epsilon}_{gb}/\dot{\epsilon}_{mcr}$ decreases with increase in stress it follows that $m < n$. If the creep rates for the two fine-grained specimens are compared at 10% strain then the stress sensitivity of the creep rate $n = 6.1$. Since $V \propto (\dot{\epsilon}_{gb})^3$ then $V \propto \sigma^{3m}$. At 41.5 MN m^{-2}, $V_{41.5} \propto \epsilon^{2.24}$ and at 62 MN m^{-2}, $V_{62} \propto \epsilon^{2.54}$ (fig. 6). Again if comparison is made at 10% strain then the stress-sensitivity of the sliding rate $m = 5.0$. The difference $n - m = 1.1$ is in agreement with the results obtained from direct measurements of grain boundary sliding by Horton [21], and indicates that when sliding is important to cavity growth, as it certainly is for wedge-cracks, then the strain exponent will vary with stress.

5. Conclusions

(i) In unstabilised 20 Cr/25 Ni stainless steel the rate of density change $\partial(\ln - \Delta D/D)/\partial(\ln t) = 3$ for both unirradiated and irradiated material and this time exponent is independent of both grain size and stress.

(ii) Wedge-cracks were more common than rounded cavities although growth of helium bubbles on the grain boundaries of irradiated specimens accelerated the propagation of the wedge-cracks, reducing the ductility from 25 to 11%.
(iii) The strain exponent of cavity growth $\partial(\ln - \Delta D/D)/\partial(\ln\epsilon)$ increases with increase in stress because the stress exponent of the creep rate $\dot{\epsilon} = A''\sigma^n$ is greater than the stress exponent of the grain boundary sliding rate, $\dot{\epsilon}_{gb} = A'\sigma^m$.
(iv) Analysis of the results using a simple crack growth model indicates that cavities grow in a stable manner for most of the life.

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