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by

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Temperature Controlled 'Void' Formation and Associated Mechanical Property Changes

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Abstract

'Voids' are frequently observed in structural materials during high temperature service, and also in neutron irradiation environments. Present paper briefly reviews the nucleation and growth of these voids and their influence on some mechanical properties.

Introduction

Both macroscopic and sub-microscopic voids are known to nucleate and grow at weld interfaces<sup>(1)</sup>, grain boundaries<sup>(2,3)</sup> and also in the matrix of structural materials during high temperature service under varying conditions like stress or irradiation environments<sup>(4,5)</sup>. Appropriate lattice defect concentrations and atomic migration rates<sup>(3,6)</sup>, besides presence of an uniaxial stress<sup>(7)</sup> are some of the important parameters in the nucleation and growth of such voids. All equilibrium defect processes are temperature dependent, and as such, this factor plays an important role in void formation. Present paper briefly reviews the theoretical aspects involved in void growth, and their influence on some mechanical properties of structural materials.

Void Nucleation and Diffusion Growth Models

The basis of void nucleation and their growth by diffusion processes is primarily dependent upon two parameters - (i) vacancy concentrations at internal 'sources' and 'sinks' and (ii) the chemical potential of atoms existing at different sites. Normally a matrix maintained at a constant temperature is expected to have an equilibrium vacancy concentration.

However, fluctuations in defect density can occur at impurity atoms, dislocations, grain boundaries, and second phase interfaces, and appropriate supersaturation conditions can develop to nucleate small voids. These structures, initially consisting of  $10^3$ - $10^4$  vacancies, once nucleated can grow by absorption of vacancies generated thermodynamically at normal vacancy 'sources'. Localised vacancy currents between internal 'sources' and 'sinks' have already been used conceptually in Herring's model to explain diffusion creep. <sup>Ample</sup> ~~Less~~ experimental evidence concerning this method of 'void' nucleation is also available now (9).

The growth of voids whether located at the grain boundaries or in matrix, has been analysed on the basis of chemical potential ( $\mu$ ) of an atom existing at these sites. Chemical potential gradient being the driving force for vacancy migration,  $\mu$  for an atom at void interface ought to be higher than at the vacancy source, for void growth to occur. To account for voids observed in high temperature deformation, Balluffi and Seigle (5) have also examined the involvement of an uniaxial stress in chemical potential formulation. Using an elastic model, these workers have concluded that chemical potential difference of an atom existing at the grain boundary, and at void surface over that in free state is given by eqns. (1) and (2).

$$(\mu - \mu_0) = \frac{-n \sigma^2}{2E} - n \sigma \left(1 + \frac{\sigma}{E}\right) c_0 s^2 \Theta \dots (i)$$

$$(\mu - \mu_0)_{\text{v.c.d}}^{gb} = \frac{3(1-\nu) n \sigma}{4E} \left(3\sigma - \frac{8\gamma}{r}\right) - \frac{2\sigma\gamma}{r} \dots (ii)$$

$n, \sigma, E, \gamma, \nu, r$  and  $\Theta$  are atomic volume, uniaxial stress, elastic modulus, surface energy, Poisson's ratio, void radius and interface inclination to stress axis respectively. For void grow to occur  $(\mu - \mu_0)_{\text{void}} > (\mu - \mu_0)_{\text{gb}}$ . Simplification of this condition leads to critical stress  $\sigma^* \approx \frac{2\gamma}{\cos^3 \Theta}$ .

Calculations show that for growth void of  $r = 10^{-4}$  cm, requires

$$\sigma^* > 10^7 \text{ dynes/cm}^2.$$

### Void Formation at Diffusion Zones, and under Creep

The presence of voids in diffusion zones and weld interfaces can be attributed to Kirkendall type vacancy supersaturation due to different diffusion mobilities of the constituent atoms. The phenomenon is observed even in absence of an externally applied stress. The condition required by Seigh and Balluffi<sup>(6)</sup> is, however, fulfilled by an intrinsic stress which develops due to shrinkage of diffusion zone in the direction of diffusion. Vacancy concentrations at edge dislocations in diffusion zone <sup>have</sup> a role in inducing this internal stress field. Grain boundary cavitation is frequently seen in high temperature deformation. Absence of pores in the matrix in this case, is ascribed to either low dislocation density or weak vacancy currents. On the other hand, once nucleated, a void at grain boundary can grow ~~substantially~~ much faster due to higher atomic diffusion rates along the grain boundary itself. Michelin<sup>(10)</sup> has proposed that vacancy flux for void growth may be provided by plastic deformation as well. Inhibition of grain boundary cavitation under hydrostatic pressure is a point in favour of diffusion models.

Grain boundary sliding during high temperature creep has been suggested<sup>(11,12)</sup> as an alternate explanation for grain boundary cavitation. Voids are supposed to nucleate at triple points or boundary ledges due to 'sliding' provided surface energy considerations are satisfied. Mathematical analysis<sup>(11)</sup> of the problem, shows that growth of voids should occur in two stages. In the initial stage, growth rate is independent of applied stress. Subsequently, when the cavity becomes large, normal stress makes a significant contribution.

Submicroscopic Voids during Irradiation

'Void' formation in fast reactor structural materials is reported to be a major problem. It causes swelling which can be expressed by a relationship,  $\Delta V/V = K (\phi.t)^n$ . (Temp. func.). Here,  $(\phi.t)$  is an integrated irradiation dose, and  $n$  is the flux exponent. Examination<sup>(13)</sup> of stainless steel foils irradiated to fast neutrons in 300-600°C range has revealed (i) the voids to be polyhedral (ii) exhibiting matrix dispersion, and (iii) of fairly small size ( $\sim 100-400\text{\AA}$ ). The void density is seen to be highly temperature dependent.

The origin of irradiation induced voids is different from those discussed earlier, though similar in logic. Fast particle irradiation is supposed to produce simultaneously both interstitial and vacancy type point defects. The favourable conditions for void nucleation depend upon the migration rates of the two types of defects and their mutual annihilation. These parameters are thermally activated and ~~then~~<sup>these</sup> temperature controlled. The governing equations for interstitial and vacancy accumulation in matrix are

$$\begin{aligned} D_i \nabla^2 c_i + K - \alpha c_i c_v - D_i Z_p P_a c_i \\ D_v \nabla^2 c_v + K - \alpha c_i c_v - D_v Z_v P_a c_v \end{aligned} \quad \dots (3)$$

Here,  $c_i$  and  $c_v$  are interstitial and vacancy concentrations,  $K$  is defect production rate. The third and fourth terms represent the loss by recombination and at dislocation sinks respectively. The growth of the void occurs, provided

$$\frac{\partial r_v}{\partial t} = \left[ D_v \frac{\partial c_v}{\partial r} - D_i \frac{\partial c_i}{\partial r} \right]_{r=r_v} \quad \dots (4)$$

Without going into details it can be said, that void growth depends upon favourable vacancy supersaturation and absence of interstitials. At

temperatures above  $T \sim 0.3 T_m$ ,  $D_1 \gg D_v$ , and the required condition is obtained. Above  $T = 0.5 T_m$ , thermal activation of vacancies is too fast, leading to dissolution of voids. Temperature is, therefore, truly important here.

Mechanical Properties

The influence of voids on mechanical properties of materials varies with the nature and distribution of voids. In irradiated metals, these enhance the yield stress and decrease ductility. Grain boundary voids in high temperature creep effect the creep rupture properties. The mechanisms accounting for these <sup>are</sup> discussed below.

Hardening due to Small 'Voids'

The strengthening involving long or short range interaction terms, due to dislocation void interaction has been treated as an elastic modulus interaction by Weeks et al. (14) and Fliischer (15). Hardening is due to a force on dislocations moving through a field of voids. Force is equal to interaction energy gradient which is greatest near the void interface. Following a weak interaction model, Weeks et al. (14) find the yield stress to be  $\sigma_1 = -E_{max}/b.l.d.$ ; 'a' and 'l' being the void radius and separation respectively. The flow stress corresponding to a strong interaction model is given by  $\sigma_2 = \left( Gb/2\pi b \right) \ln \left( a^*/r_c \right)$  ; where G, b, a\* and  $r_c$  are rigidity modulus, Burger's vector, void radius and dislocation core respectively. The short range interaction due to void is usually -ve and leads to strong dislocation-void binding. Presence of voids, therefore, contributes to resistance to dislocation climb and enhances creep resistance. A positive interaction term causes creep resistance at low applied stresses. Either way the void causes hardening. Monteff et al. (17) have found excellent agreement between predicted values and experimental data.

Creep Rupture Characteristics due to Grain Boundary Voids

Large grain boundary voids do not cause hardening. These influence the creep properties drastically. Models for creep rupture are based on diffusion controlled growth, or stress induced cavitation and mechanical shearing. Both models tend to estimate the time for creep-rupture. Dobes and Cadek<sup>(16)</sup> have given an expression

$$t \approx 1.7 \frac{L^6}{\beta^3} \quad \dots (5)$$

where  $p$ , and  $\beta$  are two stress and temperature dependent material constant and  $B$  is a diffusion mobility term. Chemical potential and diffusion coefficient values favour grain boundary void growth, but a good agreement with experimental data is not found. Assuming a time-dependent stress relaxation process and g.b. sliding mechanism, Evans<sup>(11)</sup> has given the correlation for time to creep fracture as

$$t_f = \frac{L}{A \cdot K \cdot \dot{\epsilon}_m} \quad \dots (6)$$

where  $\dot{\epsilon}_m$  is the mean strain rate and  $L$ ,  $K$ ,  $A$  are proportionality constant. Even in this case, ample variation in relation to actual results are seen. There is scope to improve the theoretical aspects here.

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