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G. K. Williamson & G. M. Jenkins

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SUMMARY

The steady state creep of polycrystalline graphite under irradiation is adequately explained above 250°C by the rearrangements in internal stresses caused by the anisotropic expansion of component crystals. At lower temperatures the creep strains predicted are greater than those observed. This is attributed to the more effective pinning of dislocations at these lower temperatures by irradiation damage. This theory does not predict any increase in total plasticity due to steady state irradiation induced creep.

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Barry Steele
Authorizing Official

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1. INTRODUCTION

Creep in graphite (see Losty (1963), Blake-lock (1963) and Simmons & Perks (1963)), can conveniently be considered as a transient component and a steady state creep. It is proposed to discuss these separately.

2. STEADY STATE CREEP

Internal stresses in graphite caused by anisotropic crystal growth have been observed to produce a creep-like relaxation under stress. The effects of thermal cycling (Jenkins & Williamson 1963) and of bromination (Jenkins to be published) have been analysed in detail and both obey the predictions based on the following simple mechanism (Jenkins & Williamson, 1963). On initial loading the strain ϵ is given by (Jenkins (1962))

$$\epsilon = A\sigma + B\sigma^2 \quad (1)$$

whereas on unloading from a stress σ_m it follows a curve

$$\epsilon_m - \epsilon = A(\sigma_m - \sigma) + \frac{B}{2}(\sigma_m - \sigma)^2$$

This behaviour is shown schematically in Figure 1.

Figure 1: Stress-strain curve for graphite demonstrating the irradiation induced creep OO' due to generation of internal stresses ($\pm\Delta\sigma$) in a constant stress (σ_m) creep test. When the graphite is self stressing all parts of the graphite are proceeding up the curve $CAOF$ and the increment of strain is OG , much smaller than OO' .

It can be shown that the strain increment $\Delta\epsilon$ due to a change in the standard deviation of internal stress $\Delta\sigma$ is given by

$$\Delta\epsilon = B \sigma \Delta\sigma + B (\Delta\sigma)^2 / 4$$

The last term is small and will be neglected, (i.e. $\Delta\epsilon \sim B\sigma\Delta\sigma$). In the case of thermal cycling $\Delta\sigma$ can be expressed as a function of the differential growth of crystals,

$$\text{i.e., } \Delta\sigma = F\Delta T (\alpha_c - \alpha_a) / A$$

where F is a geometrical averaging factor and $\frac{1}{A}$ the shear modulus, $(\alpha_c - \alpha_a)$ the differential expansion coefficient of the crystals and ΔT the thermal cycle. The thermal cycling data showed F to be $\frac{1}{20}$ agreeing with approximate estimates which included the effect of Mrowzowski cracks. The value of F will be larger in the absence of Mrowzowski cracks and may be as large as $\frac{1}{7}$. This is not a continuous creep process, the increment of strain appearing on the first cycle.

This mechanism must necessarily apply to the case of irradiation creep (Williamson 1962) where anisotropic growth occurs. Substituting $(G_c - G_a)$ for the differential strain term $(\alpha_c - \alpha_a) \cdot \Delta T$ gives

$$\Delta\sigma = F (G_c - G_a) / A \simeq (G_c - G_a) / 20A$$

$$\text{and } \Delta\epsilon = B \sigma (G_c - G_a) / 20A \quad (2)$$

This form of calculation has been confirmed by Blackburn (1962) though he derived a different value for F.

Blackburn (1962) has considered the thermal cycling creep that would have been obtained if graphite yielded plastically at each cycle (Cottrell creep). He finds a strain increment per cycle of

$$\Delta\epsilon = 7 \frac{\sigma B}{A} (\alpha_c - \alpha_a) \Delta T$$

The scatter in the observed data would permit a continuous creep increment of at the most $\frac{1}{1000}$ of this and hence we believe that this mechanism makes no significant contribution.

Values for G_c and G_a have been given by Simmons and Reynolds (1962) and Martin and

Kelly (1963). Equation 2 can be rewritten in terms of Losty's creep constant K as

$$K = \frac{\Delta\epsilon}{\sigma D} = \frac{B}{20A} \frac{G_c - G_a}{D}$$

Values are listed in Table 1 and are temperature dependent. Values for the ratio B/A are not well defined but are within the scatter band of currently available data.

TABLE 1

Temperature °C	150	200	250	300	350	450
K (calculated)* $\parallel$	12.0 6.0	4.0 2.0	1.0 0.5	0.5 0.25	0.4 0.2	0.3 0.15
K (Losty) $\parallel$	— —	— —	0.4 0.15	— —	— —	— —
K (Simmons) $\parallel$	— —	— —	— —	0.5 0.2	— —	— —

K in (dynes/cm²)⁻¹ (thousand MWD)⁻¹ 10⁻¹¹

*calculated from B/A $\perp \approx 4.10^{-9}$ (dynes/cm²)⁻¹
 $\parallel \approx 2.10^{-9}$ (dynes/cm²)⁻¹

Blakelock (1963) has shown that the temperature dependence of the creep in the range 200 – 300°C is very much smaller than that shown in Table 1. The two possible explanations for the lack of temperature dependence are:

- (1) Another temperature dependent mechanism is present which becomes significant above 200°C and increases in importance with increasing temperature. This seems unlikely since the calculated effect of internal strains is larger than the observed creep at 250°C and an additional mechanism would add to the discrepancy.
- (2) The factor B appropriate for use in equation 3 is temperature dependent, the product $[B (G_c - G_a)]$ being approximately independent of temperature in the temperature range where results are available.

The last possibility is the more attractive since it is known that, under short term testing the factor B is decreased by irradiation, particularly low temperature irradiation (Jenkins,

Williamson & Barnett (1963)). On the other hand there is increasing evidence that the interstitial clusters can migrate slowly, particularly at higher temperatures. In a creep test of long duration such effects allow the value of B appropriate to equation 3 to approach the pre-irradiation value. The use of the pre-irradiation B value in Table 1 gives the maximum possible creep rate due to this mechanism. On this basis it would appear that the interstitial drift rate at 300°C allows the dislocations to move without significant impediment, hence the agreement between calculated and observed creep rates. To get agreement at 250°C a B value half the pre-irradiation value is needed implying that the interstitial drift rate is too slow. It is not yet clear whether irradiation damage plays some part in controlling the drift rate or whether it is purely a function of cluster size and dislocation arrangement, but experiments are being planned to study these effects.

The mechanism which we have proposed above involves no change in the basic mode of deformation of graphite and hence no significant increase in ductility over the unirradiated value can be expected. Steady state irradiation creep could only increase the ductility if irradiation introduced additional plastic deformation mechanisms. It would have to introduce three additional independent systems if it were to provide unlimited ductility. These conclusions follow from the Taylor-Von Mises criterion, i.e., that a solid polycrystalline body requires 5 independent slip systems if it is to suffer an arbitrary shape deformation without breaking up. In the case of graphite the presence of pores and cracks allows some relaxation of the requirement of strain continuity. Hence it is possible to explain the low Poisson's ratio of graphite (about 0.1) instead of the value of 0.5 for plastic (or elastic) shear deformation in solid bodies.

The only additional deformation modes which can be induced by irradiation are those involving climb of the basal plane edge dislocations (see Hesketh 1963 and Jenkins et al 1962). However, the "climb" deformation is identical with that produced by kinking into Mrozowski cracks already available before irradiation so that this process will involve little if any change in ductility. Hence irradiation can cause a reduction in ductility (in the steady state creep) by filling Mrozowski cracks though, if climb mechanisms do operate they may compensate for this.

In both unirradiated and irradiated graphite the Taylor-Von Mises criterion is only satisfied to small strains which are limited because the free volume can only accommodate a small amount of distortion and hence can only substitute for the missing deformation systems to a limited extent.

In view of the technological importance of this conclusion it is worth enunciating the general effects required to invalidate the conclusion that there will be no increase in ductility in steady state creep. In terms of Taylor-Von Mises it is that independent deformation in the c axis direction is necessary. This requires either this discovery of a new dislocation deformation system under irradiation (theoretically a very improbable effect), a completely new theory of steady state creep involving stress induced control of c axis growth of crystals or grain boundary sliding.

3. TRANSIENT CREEP

On the above arguments any significant increase in creep ductility during irradiation must stem from the transient creep component. It is now established that the transient creep strain is approximately equal to the pre-irradiation strain under the same load. Hence it is possible that the strain to fracture will be about twice the pre-irradiation value provided that the transient creep mechanism involves some new deformation mechanism (as discussed in section 2). Since total creep strains approaching this magnitude have now been obtained it seems unlikely that theories involving normal deformation modes, will apply. One such theory suggested by Perks and Simmons (1963) and Kelly (1964) assumes that irradiation assists dislocations in overcoming frictional barriers in the same way as in thermal creep. Jenkins (1963) has analysed the total strain (after infinite time) available in thermal creep and shows that the creep strain $\epsilon_c = 2B\sigma_c$ where σ_c is the frictional stress to be overcome. At the fracture stress σ_f since $\sigma_c \ll \sigma_f$ the strain is much less than the fracture stress ($B\sigma_f^2$) and hence too small to agree with the observations.

Hesketh (1963) has examined the effect of an applied stress on the climb of edge dislocations parallel to the c axis and deduces a transient creep behaviour of the correct magnitude and time constant. There are two difficulties in applying his theory. Firstly, most of the strain

arises from normal deformation modes and hence this does not completely avoid the Taylor von Mises restriction. Secondly, the mechanism should necessarily produce an expansion of the crystals in the a axis direction. The maximum value constant with the published data for P.G.A. graphite corresponds to a bulk growth of 0.1% (Williamson and Horner 1963) through no effect is apparent below 250°C. If an applied stress could completely inhibit deposition of interstitials on all boundaries under tension then one could release an additional strain of about 0.1% by this mechanism. This is smaller than the transient strains already observed in P.G.A. graphite.

4. IRRADIATION CREEP AND THE RELIEF OF WIGNER STRESSES

The views outlined so far imply that the strain to fracture under irradiation creep conditions is at the most 0.1% greater than that in the absence of irradiation. If, however, the transient creep gives a larger strain than the 0.1% estimated here it becomes important to know the strain rate and hence the stress levels involved. Care has to be taken in applying the arguments involved in section 2 to the special case of Wigner stresses.

In this case there is no unloading term, all parts of the graphite progressing up the initial loading curve (see CAOF on Fig. 1). Thus when the average stress is σ one population is centred on point A, at a stress of $(\sigma + \Delta\sigma)$ and the other at F, $(\sigma - \Delta\sigma)$. The average strain does not correspond to point O (the strain corresponding to σ) but to the mean of F and A at G. For the case taken this is much less than 00°. It is easy to show that in this case the strain is $B\Delta\sigma^2$, independent of σ . If the stress σ is some multiple $H \times \Delta\sigma$ (and the two will usually be related this way) the creep strain is $\frac{1}{H} (B \sigma \Delta\sigma)$, i.e. $\frac{1}{H}$ of the creep rate determined in a constant stress experiment.

5. CONCLUSIONS

(1) A theory of steady state irradiation creep has been developed which accounts for the observed creep rates at 300°C. The lack of a temperature dependence in creep rate is attributed to a temperature dependence of dislocation pinning by interstitial clusters.

(2) The mechanism involves the same deformation system as in normal tests and hence produces no increase in strain to fracture.

(3) The strain to fracture can not be significantly increased by any steady state creep process unless irradiation induces a new deformation system giving deformation along the c axis.

(4) Transient creep may give rise to some additional ductility of about the same magnitude as the pre-irradiation strain to failure.

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