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Point Defects and the Creep of Metals

F. A. Nichols

MASTER

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F. A. Nichols

Materials Science Division
Argonne National Laboratory
Argonne, Illinois 60439

ABSTRACT

Basic concepts felt to be important in diffusion-controlled creep of metals are reviewed and it is suggested that such creep is controlled by edge-dislocation climb under a rather wide range of conditions. The effect of a damage-producing flux on such creep processes is explored. It is shown that processes such as Herring-Nabarro creep are unaffected by irradiation. Evidence is presented for a climb-plus-glide mechanism of radiation creep for stresses above unirradiated yield or flow stresses. At lower stresses a preferential dislocation loop nucleation model is suggested.

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Introduction

It is generally considered that the high-temperature creep of metals and alloys is diffusion-controlled over rather wide ranges of testing conditions. This consensus initially arose primarily due to the approximate equivalence of the activation energies for diffusion and high-temperature creep in various materials [1]. The number of materials showing this equivalence has greatly increased over the years. More recently, the approximate equivalence of the activation volumes for diffusion and high-temperature creep of several materials has been cited as additional corroboration of this viewpoint [2,3].

Recognizing that thermal diffusion is controlled by the formation and migration of atomic defects, one naturally wonders whether under irradiation by a damage-producing flux of particles there might not occur processes similar to thermal creep processes but where the rate is higher because thermal formation of atomic defects is not required.

Our purpose here is to summarize some basic concepts of various elemental processes felt to be important for high-temperature creep without irradiation, then to discuss the conditions under which such processes may or may not be altered by a damaging flux, and finally to outline our present understanding of the observed creep rates of various metals and alloys during irradiation.

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Diffusion-Controlled Thermal Creep

If we assume conditions under which dislocation glide is negligible, creep is normally attributed to the net transport of point defects between sources and sinks oriented such that the applied stress does differing reversible work to aid or hinder the emission or absorption of the defects. Perhaps the best-known example obtains when the operative sources and sinks are different grain boundaries and the defects diffuse through the interior of the individual grains, as originally proposed by Nabarro [4] and more precisely analyzed by Herring [5]. Coble [6] expanded the model to allow for diffusion along the boundaries. Several analyses have been presented [7,8], all of which have the form.[†]

$$\dot{\epsilon}_{\text{HNC}} = [AD/d^2 + B D_B/d^3] \sinh (\sigma\Omega/kT) \quad (1)$$

where $\dot{\epsilon}$ = strain rate, σ = applied stress, Ω = atomic volume, D = volume diffusivity, d = grain diameter, D_B = grain boundary diffusivity. The different analyses yield slightly different values for the constants A and B, dependent on details of the approximations invoked. Except for these rather minor uncertainties, Eq. (1) is unique in that all parameters are well defined and so quantitative predictions are possible for this model. Indeed it has received considerable confirmation from carefully controlled experiments with pure metals, especially in the thin-film and small-wire geometries [9]. However, considerable questions have recently arisen concerning the existence and explanation of a so-called threshold stress [10] below which creep has been shown to essentially cease in a number of bulk polycrystalline materials. In such cases the linear stress dependence predicted by Eq. (1) at low σ 's ($\sigma\Omega \ll kT$) is evident only

[†]The sinh-dependence is what in fact arises from the model; the linear dependence usually quoted arises from assuming that $\sigma\Omega \ll kT$

$$v_p = \frac{2 \pi D}{b} [\exp(\frac{\sigma^* \Omega}{kT}) - 1] / \ln(R/b) \quad (2)$$

when it climbs by emitting vacancies, and

$$v_a = \frac{2 \pi D}{b} [1 - \exp(-\frac{\sigma^* \Omega}{kT})] / \ln(R/b) \quad (3)$$

when it climbs by absorbing vacancies. Here, b = Burgers vector, σ^* = local normal stress acting in the direction of the Burgers vector and R = radial distance from the dislocation at which the vacancy concentration is assumed to reach the equilibrium value. For processes such as Nabarro creep discussed above, the total creep rate is determined by additive contributions from (2) and (3) thus producing the hyperbolic dependence on σ or σ^* . When the climb process proceeds for a certain critical climb height, following which the dislocation undergoes a certain amount of glide, creep strain occurs during glide even if the two climb events are not in concert, i.e., either positive or negative climb, if it allows overcoming the glide obstacle, will produce creep strain from the subsequent glide which always occurs in the direction determined by the stress. This will be seen to be a critical distinction when we consider radiation effects.

The diffusional movement of jogs on a screw dislocation follows relations analogous to (2) and (3) for vacancy-producing and vacancy-absorbing jogs [13], respectively. For low stresses, both show identical, linear stress dependencies. However, at high stresses, (2) predicts an exponential stress dependence whereas (3) predicts a limiting finite velocity. If we assume roughly equivalent numbers of the two types of jogs on a given screw dislocation, it is clear that a steady-state, diffusion-controlled velocity requires both types to move at the same velocity. This in turn means that the glide velocity of such a screw dislocation can never exceed the intrinsic limiting velocity of the less mobile vacancy-absorbing jog. The analysis by Barrett and Nix [14], in which the overall

velocity was set equal to the average of the independent velocities of the two types of jogs, is clearly incorrect. But in spite of this and even though disclaimed subsequently by Nix [15], this incorrect formulation is still quoted widely. The correct formulation was first given by Holmes [16] and discussed in more detail by Duffin and Nichols [17]. The resulting velocity is

$$v_s = \frac{4 \pi D}{b} \tanh \left(\frac{\sigma^* b^2 \ell}{kT} \right) \quad (4)$$

where σ^* is now the local shear stress acting on the dislocation and ℓ is the average spacing between jogs. Thus it is clear that this process cannot be rate-controlling at high stresses where an exponential stress dependence is observed experimentally.

Thus it is likely that at least at higher stresses, high-temperature thermal creep is controlled by some type of climb process. Various specific models have been presented, notably by Weertman [2], which may have merit in specific cases but have not met with universal acceptance. There are many authors who espouse a "recovery-controlled" process but surely steady state demands that the glide process proceed at exactly the same rate as the recovery process so that these are alternative approaches. Phrased another way, the recovery rate sets the steady-state dislocation structure which in turn sets the glide rate. A more comprehensive approach would consider some strain-producing glide process (possibly involving climb) followed by some recovery process (probably involving climb) where the two processes must occur in sequence and hence the overall rate will be dictated by the inherently more sluggish process. A specific example is Weertman's model [12], in which the leading dislocations of opposite-type pileups on parallel planes annihilate each other at a rate set by their climb velocities, thus enabling the production of new dislocations and the glide of

all the dislocations in those pileups to new equilibrium positions thus producing the creep strain. Here the source activation and glide process is considered to be much easier than the climb-recovery process so that this is in fact a recovery-controlled model, though sometimes labeled simply as climb-controlled. Even here it is of course implicitly assumed that the movement of screw dislocations occurs easily at the rate required to maintain the steady state. Since the various processes of edge glide and climb as well as screw glide and possibly cross-slip will in general have different stress- and temperature-dependencies, it is unrealistic to expect that such simplified models would be applicable over wide ranges of these variables, but only rarely does one see attempts to model a complete set of processes involved in steady-state creep, each with its own dependencies on σ , T , etc., and each controlling the overall rate in certain regimes of variables.

Diffusion-Controlled Creep during Irradiation

Perhaps the earliest suggestion for explaining an irradiation-enhanced creep rate was by Schoeck [18]. He suggested that the supersaturated concentration of vacancies maintained during irradiation would produce an enhanced diffusivity

$$D_{\text{rad}} = D_v (C_e + C^*) \quad (4)$$

where D_{rad} = effective diffusivity during irradiation, D_v = vacancy diffusion coefficient, C_e = equilibrium concentration of vacancies and C^* = excess vacancy concentration due to irradiation. He then assumed that diffusion-controlled creep models could be invoked directly simply by replacing D with the D_{rad} so defined. We now know that such enhanced diffusivities do in fact obtain but Schoeck erred in applying them directly to predict creep rates in that he

ignored the self-cancelling tendency of the interstitials also produced during irradiation. A more subtle effect enters which renders some types of diffusion-controlled creep processes unaffected by irradiation even if this self-cancelling effect is off-set in some manner.

Duffin and Nichols [19] showed that there are two requirements for irradiation to enhance diffusion-controlled creep processes:

1. The creep strain must occur as a separate process from the absorption of defects at sinks.

2. A net mis-match must exist between vacancy and interstitial currents into these sinks which must operate to permit the creep strain.

The first requirement rules out the radiation enhancement of Herring-Nabarro or Nabarro creep, since both of these occur by defect absorption alone. It does not rule out a climb-controlled mechanism where the creep strain occurs by glide following the climb process. The second requirement merely expresses the fact that if equal numbers of vacancies or interstitials enter a sink, there is no net effect.

The reason for the non-enhancement of mechanisms such as Herring-Nabarro or Nabarro creep is easily demonstrated. For simplicity, consider one-dimensional diffusion in a region bounded by two sinks at $x = \pm L/2$. If recombination of vacancies and interstitials is negligible, we can treat the two types of defects independently. For either one we have, by conservation,

$$\nabla \cdot J = \frac{dJ}{dx} = \dot{a}, \quad -\frac{L}{2} < x < \frac{L}{2} \quad (5)$$

which integrates directly to yield

$$J = \dot{a}x + b. \quad (6)$$

Now these creep processes require the effective transport of defects from one sink to the other, i.e., for our assumed geometry,

$$\dot{\epsilon} \propto \left| J \left(x = -\frac{L}{2} \right) + J \left(x = +\frac{L}{2} \right) \right|, \quad (7)$$

and from the expression for J we have

$$\dot{\epsilon} \propto |2b|. \quad (8)$$

We previously concluded [46] from Eq. (8) that $\dot{\epsilon} \neq f(\dot{\alpha})$ even where an asymmetrical interaction energy existed between the two sinks. Thus, for example, there may exist interaction energies between the stress/strain fields of the point defect and the sinks. The local diffusional flux will be due to the gradients of both that interaction energy and the concentration of the defect and, by assuming that $b \neq f(\dot{\alpha})$, we concluded that, even with an asymmetric interaction energy field, when steady state is reached the concentration gradients will have adjusted themselves so that the net overall effect is that the radiation-produced defects are absorbed by the two sinks in equal numbers; the creep rate would then be exactly that dictated by the boundary conditions and is identical to what would occur without irradiation.

We concluded then that this rules out not only the more conventional Herring-Nabarro types of mechanisms but also the so-called Stress-Induced-Preferential-Absorption (SIPA) model espoused recently by several authors [20,21] as a mechanism for radiation-enhanced creep. In this concept, the interaction energy with point defects differs for two neighboring edge dislocations of different orientations with respect to the applied stress. This produces a net preference of say interstitials for one dislocation and the idea is that this then produces the required net effective transport from one

dislocation to the other. We argued, from the above reasoning, that this preferential attraction is offset by adjustments in concentration profiles to produce no net effect. However, Wiedersich has pointed out [47] that we did not prove that b is independent of a . Thus, our conclusion concerning SIPA should be held in abeyance until a more complete analysis can be done. Such an analysis will be reported separately.

As an example of a mechanism which can be enhanced by irradiation, Duffin and Nichols [19] treated two nearby dislocations held up by obstacles which without irradiation climb in opposite senses by an effective vacancy transport from one to the other but after reaching a critical climb height then glide until they encounter another obstacle. The glide of course produces strain now in addition to any Nabarro-type strain induced by their climb. To make the model quantitative, they chose parameters felt appropriate for zirconium-base alloys at 300°C and 20,000 psi tensile stress under a neutron flux of $6 \times 10^{12} \text{ n-cm}^{-2}\text{-sec}^{-1}$ ($> 1 \text{ MeV}$). They assumed the depleted zones formed by each damage cascade to be the dominant dislocation obstacles and used the effective mis-match between interstitial and vacancy currents into the dislocations as a variable. Their results are shown in Fig. 1, along with data obtained by Fidleris [22] under the assumed conditions. There is a flux or damage-production level below which there is predicted no effect of irradiation. This is a manifestation of the assumption that neighboring dislocations climb in opposite senses in the absence of irradiation. Thus, if we impose upon that situation a net current of one type of point defect (either vacancy or interstitial) then one dislocation is slowed in its climb by exactly the same amount as the other is speeded up, yielding a null effect. This remains true until a sufficient current of defects to completely stop the climb of one dislocation is attained. For currents (neutron fluxes) above this critical value that dislocation increases its climb velocity (linearly with increasing neutron flux) but in

the opposite sense so that both climb in the same sense. The essential point here is that the climb sense is immaterial since the dislocation can escape the obstacle by either positive or negative climb. The difference in their climb rates remains fixed at all flux levels (a manifestation of the nil-effect of irradiation on Nabarro creep) but since both produce additive creep strains in the glide to the next obstacle, the net effect is an increased creep rate directly proportional to the climb rate. (These authors assume the glide process to be relatively unhindered).

With reasonable choices of parameters, the predicted creep rates are seen to be compatible with observed steady-state rates if one assumes the effective mis-match in point defect currents to be of the order of 1%, a value which is not inconsistent with some theoretical estimates which assume that the preferential attraction of interstitials to dislocations provides the required bias in currents. The plausibility of the model, coupled with this consistency check, make this a viable candidate for explaining the radiation-enhanced creep of irradiated metals at stresses high enough that glide in between obstacles is very rapid. The glide process, however, is expected to have an intrinsically stronger stress dependence than the climb process, so that at sufficiently low stress we would expect the glide process to control the overall rate of creep.

Experimental support for this concept is provided by data of Fidleris [23], shown in Fig. 2. In this experiment, a single specimen of a zirconium-base alloy was crept at different stress levels. Each creep-rate value shown was a steady-state value following the transient region occasioned by each stress change. That steady-state rates were in fact attained is clearly demonstrated by the coincidence of points when the stress range was traced out a second time

as indicated by the numbering of the data points. The straight-line segments are drawn in to give indication of the approximate stress sensitivities n in the general "power-law" approximation $\dot{\epsilon} \propto \sigma^n$. The value of $n \approx 14$ is consistent with glide control and is very similar to values obtained in the unirradiated material for similar stresses at this rather low temperature. The switch to $n \approx 1$ to 4 as stress is increased is very consistent with expectations for a climb-controlled escape from obstacles followed by glide. In fact, the data could support a value of $n < 1$ in the lower regions of the climb-controlled regime. For thermal creep, $n < 1$ would not be expected but it may well occur under irradiation. This is because the climb velocity is set essentially by the defect production rate and hardly affected by stress. Thus the stress dependence of the creep rate arises only from the stress dependencies of the number of climbing dislocations, the spacing between obstacles along the dislocation, the critical climb height and the mean-free path of the glide in between obstacles. Apparently all of these give a combined stress dependency (at relatively low stresses) as low as unity and possibly slightly lower. In view of the paramount importance of the stress-sensitivity in the ductility of tensile specimens in steady-state creep as displayed by the analysis by Hart [24], we would expect that radiation creep in the region of $n \approx 1$ to 4 should result in superplasticity. These ideas have been explored previously. Data exist [25] in the $n \approx 4$ region showing at least 10% elongation with no indication of necking whereas unirradiated control specimens show necking instability at around 2% strain and fracture at 10 to 12% elongation. These effects are extremely germane to the design of nuclear reactor core structurals.

An experiment has been performed by MacEwen and Fidleris [26] which was seen by them as rather conclusive evidence for the validity of our assumed

mechanism. They observed transient creep rates following an abrupt insertion into the core and following the shutdown of an operating reactor. Figs. 3, 4, and 5 show their results obtained at about 300°C with 138 MPa (20,000 psi) tensile stress and $\sim 10^{17} \text{ n m}^{-2} \text{ sec}^{-1}$ ($> 1 \text{ MeV}$) neutron flux.

As shown by Figs. 3 and 4, following initial insertion creep rate enhancement was immediate. During an initial transient of $\sim 1 \text{ h}$ the rate decreased by about an order of magnitude. It then slowly decreased and eventually attained an apparent steady state. Fig. 5 shows that removal of the flux again caused an initial increase in creep rate with a decreasing rate lasting about 1 to 2 h after which the rate was comparable to the rate prior to insertion into the flux.

The authors' explanation appears quite reasonable. Upon insertion into flux, vacancy and interstitial production begins but interstitials, being more mobile, (in fact they are also more numerous since some vacancies are immobilized in depleted zones) reach dislocations in greater numbers and soon attain their steady-state concentration. They then drain to sinks as fast as they are formed but on a longer time-scale increasing numbers of vacancies reach the sinks and partially cancel out the interstitial flux. As the vacancies attain steady-state the numbers of vacancies and interstitials reaching dislocations are almost equal, with the net climb (and hence creep) rate at a given structure being set by the bias in point-defect fluxes to the dislocations. The slowly decreasing rate with exposure reflects the slower buildup of depleted zones which attain saturation or steady state on a much longer time scale and supports our contention that these are the dominant dislocation obstacles. We would concur but would add to the author's explanation the fact that the effective production rate of interstitials is in fact greater than that of vacancies prior to saturation of the depleted zones. This effect alone also contributes to the transient creep, as discussed in detail by Dollins and Tucker [27].

The observation of increased creep rate when the flux is removed is seen as strong evidence for a climb-plus-glide model. The interstitials anneal out rapidly giving an initially increased climb (hence creep) rate. The remaining vacancies anneal out more slowly but since such a model is insensitive to the direction of climb, they too produce an enhanced climb rate (albeit in the opposite direction) which is reflected in an enhanced creep rate. The rate of course decreases as the vacancy supersaturation is depleted, and eventually reaches a rate equivalent to the out-of-flux rate.

The shapes of the transients seem reasonable and the time constants seem plausible although our knowledge of the migration energies of point defects in zirconium is too uncertain to be more quantitative. It would be interesting to perform a similar experiment with a material for which these numbers are better known.

These above results also offer evidence against several alternative proposed mechanisms for the particular testing conditions employed. Thus, the immediate enhancement upon insertion into the flux argues against either the yielding creep model proposed by Hesketh [28] as the dominant mechanism for zirconium-base alloys at all stress levels, or the preferential loop alignment model proposed by Nichols [29,30,32] as a general mechanism for lower stresses. The increased rate after flux removal argues against the thermal spike mechanism of Brinkman and Wiedersich [31] and the SIPA model since vacancy interactions with dislocations should not be nearly so strong as those of interstitials.

A similar experiment at lower stresses where Nichols has also suggested a loop-alignment model should be very informative, but would presumably require either better sensitivity than present in-pile creep machines supply or the significantly higher neutron fluxes available in other reactors, particularly fast reactors. Ion bombardment may be very useful in this regard also, for which creep machines are being designed at several laboratories [33].

Radiation Hardening and Radiation Creep

At sufficiently high stresses, we expect the dislocations to pull free of (and partially or totally destroy) the depleted zones more rapidly than they can climb out of them. This process should occur exactly the same as in post-irradiation testing, i.e., the radiation-hardening and dislocation-channeling phenomena should occur. That this is in fact the case was demonstrated by the approximately fixed-strain-rate tests of Azzarto, et al [34], again on a zirconium-base alloy. Fig. 6 demonstrates their findings. At high strain rates, the stress-strain curves obtained in-flux are the same as those obtained out-of-flux after comparable neutron fluence. They display the usual region of strain-softening due to the dislocation channeling phenomenon. However, at lower strain rates (below $\sim 10^{-4} \text{ hr}^{-1}$ for their flux level of $\sim 10^{14} \text{ n-cm}^{-2}\text{-sec}^{-1}$ and temperature of $\sim 300^\circ\text{C}$), the curves show a gradual rise to a fixed stress level indicative of steady-state creep with no indication of strain softening. The stress levels attained are significantly lower than those reached in post-irradiation tests at the same strain rate and both the stress levels and the average strain-rate sensitivity of about 0.3 are in excellent agreement with the creep tests of Fidleris discussed above. Thus, these tests give independent support for the climb-controlled region and the radiation-hardening regime where dislocations cut through the obstacles. At the lowest strain-rates employed

($\sim 10^{-6} \text{ hr}^{-1}$) the steady-state stress level is about the same as the unirradiated flow stress so the glide-control regime was not covered by these tests.

Results on a zirconium-base alloy obtained by Gilbert and Mastel [35] and shown in Fig. 7 give additional support to the entire region of glide-control, climb-control and cut-control. The curve is predicted by a theoretical model developed by Dollins [36] which incorporates all the specifics detailed above of depleted zone buildup, their interaction with dislocations, the flux-enhanced climb of those dislocations in parallel with their activated cutting of the obstacles (treated as tiny voids) and both of these processes in series with the glide process in between which is treated in the model by assuming a fixed frictional stress. The zones are allowed to boil off vacancies and to capture both vacancies and interstitials by diffusion. In a subsequent modification [37], the biased flux of interstitials to cause dislocation climb at steady state was incorporated into the model via a preferential interstitial-dislocation interaction, with sub-grains of $1 \mu\text{m}$ diameter assumed to serve as the required neutral sinks at these low temperatures where voids do not form. The treatment employs the typical reaction-rate theory familiarly used to analyze void swelling.

The temperature- and flux-dependencies predicted by this model are displayed in Figs. 8 and 9 along with available data. The agreement is considered quite satisfactory. At low temperatures, the creep rate is essentially independent of temperature. At high temperatures the depleted zones have a short life time and therefore their steady-state density steadily decreases and the creep-rate correspondingly increases rapidly. The higher the damage rate the higher the temperature-insensitive region extends.

The flux-dependency for 300°C at low flux levels approximates to an $\dot{\epsilon} \propto (\text{flux})^{0.5}$ dependence but for fluxes $\gtrsim 10^{14}$ n-cm²-sec⁻¹ it becomes essentially linear. At higher temperatures ($\gtrsim 400^\circ\text{C}$), as seen in Fig. 8, the creep of course becomes independent of flux level whereas at lower temperatures the linear flux dependence extends to lower flux levels.

Although the above experimental results pertain to zirconium-base alloys at temperatures where depleted zones formed in-situ by displacement damage cascades comprise the major dislocation obstacles the basic concepts of a climb-plus-glide model should apply equally well for other metals and alloys at similar temperatures or for all materials at higher temperatures where, e.g., the radiation-induced voids should serve as analogous attractive barriers to dislocation motion. Although very few data exist for other materials either within or below the void-swelling regime, a climb-plus-glide model with the voids serving as dislocation barriers has in fact been proposed by other investigators [39].

Low-Stress Creep during Irradiation

Except for the zirconium-base alloys, data exist primarily only at low stresses, i.e., below unirradiated flow or yield stresses. Furthermore, most of these data were obtained at low temperatures, i.e., below reactor operating temperatures. But these data exist for a very large number of materials, both pure metals and alloys, and a number of basic characteristics of the creep behavior under these conditions has been deduced [30,40]. It has been found that the low-stress creep rate is linearly related to stress and damage flux levels and weakly dependent on temperature. In fact, comparisons have shown creep rates to be typically higher (by about a factor of three) at 78°K than at room temperature. Perhaps even more remarkable is the fact that the radiation-enhanced

creep rates of a very large number of different metals (both annealed and cold-worked) and alloys (both precipitation-hardened and solution-annealed) whose thermal creep rates would differ by orders of magnitude, were all essentially the same (within an order of magnitude). This leads us to suspect that the mechanism is intimately involved with the damage production and essentially unrelated to the thermal (out-of-flux) creep properties. We thus believe that mechanisms such as yielding creep which depend directly on out-of-flux creep properties do not play a dominant role, especially since annealed cubic materials which do not display the radiation growth required by the yielding creep model show creep rates equivalent to or greater than those displayed by hexagonal metals such as zirconium.

As discussed above, we believe that long-range dislocation glide cannot occur at these low stresses so that we do not believe the climb-plus-glide model of higher stresses is applicable here. There are primarily two other models which have been proposed for this low-stress regime.

The first is that of preferential loop nucleation. In the presence of an applied normal stress σ , assuming for simplicity three orthogonal planes on which loops may form, straightforward homogeneous nucleation theory yields [32]

$$f = \frac{\exp(n\sigma\Omega/kT) - 1}{\exp(n\sigma\Omega/kT) + 2} \quad (9)$$

as the fractional excess of the total loop density on the plane favored by the applied stress, where n = the number of point defects in the critical loop nucleus. The subsequent growth of these loops by interstitial absorption produces the creep strain. Quantitative predictions are difficult due to the likelihood of heterogeneous nucleation and great uncertainty concerning the value of n . However,

one experimental study by Okamoto and Harkness [41] indeed revealed a preferential loop alignment in a stainless steel specimen which had been crept in-flux. A similar search in zirconium-base alloys has been reported to be negative [42]. However, the sample had been crept in the intermediate-stress regime where we would not expect dominance of this mechanism. More recently, Causey [45] has reported complete recovery of irradiation-induced strains in annealed zirconium alloys crept in-pile in the low-stress regime. We concur with him that this is strong evidence for a loop-alignment mechanism. The fact that the post-irradiation anneal of highly cold-worked samples did not produce complete recovery of the irradiation-induced strains may indicate a significant contribution by yielding creep, or possibly SIPA, for cold-worked material. The higher radiation creep and growth rates of cold-worked zirconium alloys would be consistent with yielding creep predictions.

Another proposal has been made recently by several authors [20,21]. This is the so-called stress-induced-preferential-absorption (SIPA) model. This is a modification of the Nabarro creep mechanism in which the net difference in climb rates of dislocations whose Burgers vectors are oriented differently with respect to the applied stress produces the creep strain. The new wrinkle here is an asymmetric interaction of interstitials with the two dislocations which is hypothesized to produce a net bias effect. We believe, as discussed above, that this type of mechanism may be possible but no reliable quantitative analysis is yet available with which to assess its magnitude. We shall discuss the shortcomings of the available analyses in a separate publication. We believe that some type of preferential loop alignment process is likely, at least in annealed Zircaloy, because of the observations of strain recovery discussed above, with possibly other mechanisms contributing in cold-worked materials, such as yielding creep due to radiation growth which occurs in cold-worked metals.

Summary

A truly complete model for long-term, steady-state dislocation creep must consider the production of dislocation loops, the movement and obstacle interactions of both edge and screw components and their eventual destruction. Normally, theoretical models focus on only one of these sequential processes and assume it to be rate-controlling. Usually either the overcoming of obstacles (climb-control) or the destruction of dislocations (recovery-control) is considered in high-temperature creep. Much remains to be learned but it seems fairly certain that such creep is normally edge-dislocation climb-controlled in some fashion but this may occur either in overcoming obstacles or in the recovery process. It seems clear that the glide of jogged screw dislocations cannot explain results at least at higher stresses where the correct analysis for such a model predicts a limiting rate rather than the exponentially increasing rates normally observed. It seems unlikely that stress dependency of the "mobile dislocation density" could overcome such grossly different qualitative behaviors.

Following high fluences of damage-producing particles such as neutrons or ions, we believe the uncertainties are alleviated considerably in that the damage process produces such a large density of small obstacles that these should essentially dominate the entire dislocation transport process. We consider, as discussed by Wiedersich [43], that since the climb rate of a dislocation for a given point defect flux increases as its edge character decreases the overall creep rate in steady state should be controlled by the movement of relatively straight edge dislocations. At high stresses or strain rates the dislocations must cut through the obstacles as in post-irradiation testing and so radiation hardening is seen even during irradiation. At somewhat lower stresses (strain rates) the climb of long edge dislocations at a rate set by the damage flux becomes faster than the cutting process and so sets the overall rate. At still

lower stresses (strain rates) the climb rate can become so fast that the glide of dislocations in between the obstacles is more difficult and thus sets the overall rate. This entire process can be expressed mathematically by

$$\dot{\epsilon} = \left[\frac{1}{\dot{\epsilon}_{\text{glide}}} + \frac{1}{\dot{\epsilon}_{\text{climb}} + \dot{\epsilon}_{\text{cut}}} \right]^{-1} \quad (10)$$

where $\dot{\epsilon}_{\text{glide}}$, $\dot{\epsilon}_{\text{climb}}$ and $\dot{\epsilon}_{\text{cut}}$ are the creep rates which would be predicted to occur by a model which considered that process alone to be rate-controlling. Since the climb and cut processes are alternative or parallel processes their intrinsic rates add (as parallel currents add in an electrical circuit). However the climb/cut combination must occur in sequence or in series with the glide in between obstacles, so that the inverses of these intrinsic rates are additive (analogous to adding resistances in series in an electrical circuit). The expected overall behavior is shown in Fig. 10. The high-stress region is the region of radiation hardening in which the cut process (with obstacle destruction and dislocation channeling) dominates. The intermediate stress region is where the flux-enhanced climb of dislocations allows them to escape their obstacles (depleted zones, loops or voids) with little or no destruction of them. The high strain-rate-sensitivity of this region implies very large (superplastic) straining capability which has been partially demonstrated experimentally. This contrasts strongly with the extremely low strain-rate-sensitivity of the high-stress regions which, coupled with the strain-softening effect of obstacle destruction, brings about a drastic reduction in uniform strain and very little necking resistance, results well established in post-irradiation testing of many metals and alloys. At the lower stresses of the intermediate stress region the glide in between obstacles, because of its larger expected

stress dependence, becomes sluggish compared with climb rates, and therefore determines the overall dislocation climb-plus-glide creep rate.

In the low-stress region of Fig. 10, we believe a new process not involving dislocation glide dominates which has been shown experimentally to have a strain-rate-sensitivity near one. We believe this to be controlled primarily by a preferential-loop-alignment mechanism. The possible contribution of the SIPA mechanism remains to be established.

For completeness, we have also indicated in Fig. 10 the phenomenon of radiation-growth which is observed in non-cubic materials (both annealed and cold-worked) and in cold-worked cubic materials. In annealed non-cubic materials, the thermal spike of a displacement cascade is believed to cause preferential strains due to precipitation of point defects in response to the orthogonal tensile and compressive stresses set up due to the anisotropic thermal expansivities of such structures. Since thermal expansivity is isotropic in cubic crystals, this mechanism does not operate. However, in cold-worked materials of either cubic or non-cubic structures the dominant direction of such working will induce anisotropic distributions of the Burgers vectors of the large number of dislocations produced and then any net climb induced by point-defect annihilation will produce anisotropic, constant-volume strains. These concepts have been developed into a quantitative model by Dollins [44] and good agreement with available data was obtained. The existence of radiation growth implies the occurrence of yielding creep and experimental results on zirconium alloys indicate it probably contributes a portion of the low-stress creep strain of cold-worked material.

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