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THE MECHANISTIC MODELING OF ZIRCALOY DEFORMATION AND FRACTURE IN FUEL ELEMENT ANALYSIS*

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Abstract

A review is given of the comprehensive model developed in the 1960's at the Bettis Atomic Power Laboratory to explain the creep of Zircaloy during neutron irradiation and applied to fuel element analysis and design. The in-pile softening observed at low stresses was hypothesized to be due to a combination of the growth-directed Roberts-Cottrell yielding creep originally proposed for α-uranium and the formation of point defect loops preferentially on certain planes in response to the applied stress, with the second process being of relatively greater importance. The in-pile hardening obsrved at high stresses (or strain-rates) was proposed to be due to the cutting by dislocations of radiation-produced obstacles. In this stress (strain-rate) region, in-pile behavior was proposed to be identical to post-irradiation behavior. At intermediate stresses (strain-rates) a mechanism of radiation-enhanced climb around obstacles was suggested as being rate-controlling. As the stress is decreased, the climb process becomes easier and the rate was then predicted to be controlled by glide at a flow stress characteristic of unirradiated, annealed material, where radiation-enhanced diffusion enabled climbing around the normal strain-hardening obstacles. At still lower stresses, this glide process became negligibly slow compared with the growth-connected creep mechanism which was presumed to operate independently. The overall scheme was shown to be good agreement with all the in-pile data then available and implemented into the computer analysis of fuel element behavior.

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Introduction

I accepted the 1984 W. J. Kroll Zirconium Medal with great pleasure and with pride for what it represented. But I also realized that, although it was for my accomplishments that I was singled out for this honor, many people and circumstances had contributed in very significant fashion. I certainly did not do my work in a vacuum or in isolation. I profited from those going before me, including Admiral Rickover, the first Kroll Medal recipient and founder of the Bettis Atomic Power Laboratory where I was privileged to work for some eighteen years. I profited very significantly from Ben Lustman, also a Kroll Medal recipient and for many years metallurgist par excellence of the Bettis Lab. And I profited, perhaps most significantly, from having a boss, Ralph Frederickson, who had confidence in me, supported me and allowed me to do my thing. Last but certainly not least, my co-worker Eliot Duncombe had the ability to translate my materials-mechanistic theory into efficient computer alogorithms suitable for use in fuel element analysis.

It is of course impossible to separate circumstances and atmosphere from personalities, but I also profited from the particular point in time when Ralph Frederickson turned me loose as a rather new PhD who had done a theoretical thesis under Bill Mullins at Carnegie Institute of Technology. That appealed to Ralph Frederickson, an analysis-oriented mechanical engineer who had difficulty in using the voluminous tables of data usually handed him by other metallurgists. Carl Friedrich had just written (I'm told in one day) the specifications for the first CYGRO computer program for analyzing fuel elements, but the Zircaloy mechanical properties algorithms were much too limited for a general-purpose program. Realizing that it had never been done and that most metallurgists thought it could not be done, but being willing to sacrifice precise detail for overall consistency, I proceeded blithely to concoct mechanical deformation and radiation-effects models for anisotropic, textured Zircaloy at all levels of stress, strain, strain rate, temperature, flux and fluence, then employed data from many diverse sources to which to fit my models and thus back out important parameters, using the theoretical models to describe functional forms.

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And so, together with many analogous models for fuel behavior, we began the game of analyzing, in much more detail than previously possible, the behavior of fuel elements under actual anticipated in-reactor histories. Prior to the CYGRO program, and I am well aware that some probably feel it should have remained so, the approach to fuel element design was to irradiate many hundreds of fuel elements with multitudinous variations and see which ones survived some presumably strenuous sets of conditions. From that point on, analysis played a key if not leading role in fuel element design, at least at the Bettis Laboratory.

It is also of great fascination to me personally to examine how our modeling efforts greatly altered the understanding of the basic mechanical properties of Zircaloy. Adopting a type of mechanical equation of state for my Zircaloy modeling, but being willing to use (imprecisely to be sure) plastic strain as an approximate state parameter (more correctly, an "effective" strain, i.e., corrected or decreased to account for recovery), we could utilize and unify results from the various mechanical tests labeled "tensile," "creep," "relaxation," etc., often treated as almost independent measurements of mechanical behavior.

We were also able to rationalize the seemingly contradictory phenomena people were measuring at that time, namely "radiation hardening," i.e., the increase in yield strength as measured in a tensile test <u>following</u> irradiation and "radiation creep," i.e., the decrease in "creep" strength (or stress to give a certain strain rate) sometimes (and sometimes not) reported when fixed-load or "creep" tests were run <u>during</u> irradiation. It also allowed us to interpret several "constant-strain-rate" tests conducted during irradiation which indicated a transition between softening and hardening and to relate these quantitatively to both the in-reactor creep and the post-irradiation tensile tests.

I want to give here a brief synopsis of some of the key experimental results available at the time and follow with a condensed account of the theoretical models developed and implemented into CYGRO and subsequent analysis programs. I shall discuss also the broad concepts of tensile instability and fracture which flowed from this analysis and led to design approaches to predictions of failure as well as deformation.

Experimental and Theoretical Background

The most-studied effect of neutron irradiation at temperatures >300°C on the mechanical properties of Zircaloy and many other metals and alloys, i.e. increase in hardness and tensile yield stress, had been well-established as a general characteristic by many post-irradiation tests[1]. Much more limited information on Zircaloy also indicated that the creep strength, i.e. flow stress for relatively slow strain-rates, is also greatly increased (at least for strain-rates as low as $\sim 10^{-7}$/h) following irradiation[2]. That is to say, the strengthening effect is evident from strain-rates typical of tensile tests $\sim 10^{c}$/h down to very low strain-rates ($\sim 10^{-7}$/h).

Some of the initial in-pile experiments indicated a decrease of creep rate (i.e., an in pile strengthening) under very special conditions[3]. On the other hand, other studies indicated an increase in creep-rate (i.e. an in-pile weakening)[4], again under very specific conditions. It became clear from later work that either result may be obtained, depending critically upon the exact experimental conditions.

Theoretical arguments had been presented which predicted that creep rates may be enhanced during neutron irradiation[5,6]. This would imply an in-pile weakening effect, rather than a strengthening one. The creep enhancement argument derived from the fact that irradition produces vacancies and interstitials. Since normal diffusion occurs via the movement of vacancies, their increased concentration during neutron irradiation should produce an enhanced diffusion coefficient. Then, since many creep processes are controlled by diffusion rates, the argument was that the enhanced diffusion coefficient is sufficient to produce an enhanced creep rate. Schoeck[5] originally applied the discussion to a dislocation climb mechanism and suggested that since climb is not normally a dominant creep mechanism (out-of-pile) below $\sim 1/2T_m$ where T_m = melting point on abosolute scale) the in-pile enhancement should not occur at temperatures much below $1/2T_m$. This argument seemed clearly incorrect since for various alternative creep mechanisms the one which will dominate is the one producing the largest strain-rate under the given conditions. Thus, the reason diffusion-controlled creep does not normally dominate at low temperatures out-of-pile may simply be that its rate is reduced below that of alternative mechanisms not depending on diffusion. However, if by some means (say by irradiation) the diffusion rate could be increased to values which normally obtain thermally only at temperatures $>1/2T_m$ then diffusion-controlled creep may well become dominant. The flux level required to yield enhanced diffusion rates of this magnitude can be roughly estimated to be $\sim 10^{14}$ nv (>1 MeV) [5,6]). Finally, it was noted that although the $1/2T_m$ rule is a useful guide, it is by no means absolute; e.g., the lower the strain-rate, the lower the required diffusion rate and hence the lower the temperature (or flux) required. This had to be kept in mind when considering the very low strain-rates of importance in nuclear reactors.

Schoeck's argument was challenged by Mosedale[7]) who interpreted it explicitly in terms of the Weertman pile-up mechanism of dislocation-climb-controlled creep. Hesketh[8]) extended Mosedale's argument to include any diffusional creep mechanism and arrived at the same conclusion: although irradiation enhances diffusion rates in between sinks and sources (such as dislocations and grain boundaries) it has no effect on diffusional creep rates since thermal equilibrium defect concentrations are maintained in the vicinity of the dislocations and boundaries, which must move in order to produce strain. Thus, the argument seemed to be quite general, not depending on any specific mechanism. However, it depended upon the assumption that dislocations and grain boundaries are truly perfect sinks and that equilibrium defect concentrations are maintained in their neighborhood. Another factor, the relative concentrations of sinks for vacancies and interstitials, is probably more important and will be discussed below.

In addition to the questions of what irradiation may or may not do to normal (or thermal) creep mechanisms, it seemed that the possiblity existed of radiation-creep mechanisms completely unrelated to thermal creep and operating quite independently. One model had been proposed by Roberts and Cottrell to account for radiation-creep of α-uranium[9]), which depended upon the radiation-growth (i.e. dimensional changes with no applied stress) which is known to occur in this material. Since Zircaloy also showed radiation growth[10]) it was felt that the Roberts-Cottrell mechanism may apply for it as well.

A number of careful experimental studies had also been reported, primarily involving cold-worked Zircaloy. One study produced no significant differences between in-pile and out-of-pile creep rates and the investigators concluded that irradiation does not significantly affect the creep rate of cold-worked Zircaloy[11]). It was noted, however, that the tests were conducted at relatively high temperatures and stresses ranging from σ>30,000 psi at 300°C to σ>20,000 psi at 375°C (and hence high thermal creep rates) and relatively low neutron flux levels (and hence low radiation-enhanced effects). Thus, these results were deemed rather inconclusive. It was also noted that these tests were conducted at temperatures near and above the recovery range for post-irradiation-strengthening defects. Since all other studies to be discussed were conducted below this recovery region, we shall restrict our subsequent discussion primarily to temperatures below this recovery range, i.e. <300°C.

Another series of in-pile uniaxial creep tests of cold-worked Zircaloy at intermediate neutron flux levels [~0.5 to 1 x 10^{13} nv (>1 MeV)], at a temperature of 300°C and extending to significantly lower stresses (>11,000 psi) was reported by Fidleris and Williams[12]. In these studies, a definitely higher in-pile creep rate obtained during neutron irradiation, and during reactor outages the creep rate decreased to a value comparable with the out-of-pile control tests. The enhanced rate was generally not in evidence during the early (i.e. relatively high strain-rate) regions of the creep curves, but at long times, the in-pile creep rate became quite steady whereas the out-of-pile control tests showed a continuously decreasing creep rate. In general, the early stages of in-pile and out-of-pile creep were quite comparable.

A third series of in-pile tests of Zircaloy was performed by Azzarto et al.[13], at a still higher flux level of 1.2 x 10^{14} nv (>1 MeV) at a temperature of 282°C. These tests employed an approximately constant-strain-rate technique with pre-irradiation-hardened specimens and it was found that the in-pile creep strength (as evidenced by the stress at a plastic strain of 0.2%) of the initially annealed Zircaloy coincided with that of material with the same integrated neutron exposure but tested out of the reactor for relatively high strain-rates ($\sim 10^{-4}$ to $\sim 10^{3}$/h) whereas in-pile tests conducted at < 10^{-4}/h showed a very pronounced weakening (compared with post-irradiation tests at the same strain-rate) although the strength was still greater than the pre-irradiation value. At the lowest strain-rate employed in these tests (5 x 10^{-6}/h) the stress attained in-pile at a plastic strain of 0.2% was as low as 21,000 psi compared with ~50,000 psi for a post-irradiation control test. Thus, enhanced in-pile creep at <300°C was demonstrated for both annealed and cold-worked Zircaloy at sufficiently low strain-rates (or stresses) and sufficiently high flux levels.

Another series of results reported by Ross-Ross[14]) involved a still different type of creep measurement, diametral changes of cold-worked Zircaloy reactor pressure tubes. These results covered a moderate range in flux levels of the order of 10^{13} nv (>1 MeV), a small temperature range in the neighborhood of 300°C and a circumferential stress ranging from 9,000 to 19,000 psi. The author correlated his results by a formula of the type

$$\dot{\varepsilon} = A\sigma\phi(T - T_0), \qquad (1)$$

where $\dot{\varepsilon}$ is the long-time, steady-state, in-pile creep rate;

A, T_0 are constants;

σ = applied stress;

ϕ = neutron flux;

T = temperature.

The linear stress and flux dependencies seemed fairly well-established, but the temperature-dependence was stated by the author to be quite crude and only valid as a rough correlation for the temperature range employed (~260 to ~325°C). In fact, a comparison of his in-pile rates versus expected thermal creep rates at comparable stress and temperature indicated that the entire temperature-dependence observed could well be due simply to the thermal creep component.

Subsequently, Ross-Ross and Hunt[15]) reviewed the studies on cold-worked Zircaloy-2 previously reported by Ross-Ross[14]) and retained the same empirical representation of the data, eq. (1), stating that they attached no significance to the equation with regard to the mechanism of creep. These authors also reported results on pressure tubes made of 20% cold-worked Zr + 2.5 wt.% Nb which indicated this material to have a lower in-pile creep rate than does 20% cold-worked Zircaloy. Another significant difference, discussed further below, was that the strain rate in-pile is actually lower for low strains than that of the unirradiated control but at longer exposures (higher strains) the opposite obtains and eventually the in-pile creep curve becomes linear and the strain exceeds that of the unirradiated specimen, which shows a continuously-decreasing strain rate. Ross-Ross and Hunt reported reasonable agreement between their Zircaloy results at 11,000 psi and one test reported by Fidleris for this stress, after converting their stress and strain to effective values defined by

$$\sigma_{eff} = \{1/2[(\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2]\}^{1/2}, \qquad (2)$$

and

$$\varepsilon_{eff} = 1/3\{2[(\varepsilon_1 - \varepsilon_2)^2 + (\varepsilon_2 - \varepsilon_3)^2 + (\varepsilon_3 - \varepsilon_1)^2]\}^{1/2}, \qquad (3)$$

where σ_1, σ_2, σ_3 and ε_1, ε_2, ε_3 are the stresses and strains in the three principal directions. For Fidleris's uniaxial test, $\sigma_{eff} \equiv \sigma_1$ and $\varepsilon_{eff} \equiv \varepsilon_1$, where the subscript 1 designates the longitudinal direction. For the pressure tubes of Ross-Ross and Hunt $\sigma_{eff} = 1/2\sigma_t\sqrt{3}$ and $\varepsilon_{eff} = 2\varepsilon_t\sqrt{3}$, where the subscript t signifies the tangential ("hoop") direction. For an isotropic material this measn that if σ_t and ε_t are replaced by σ_{eff} and ε_{eff} in eq. (1) the constant A must be increasd by a factor of 4/3. Although Zircaloy displays anisotropic creep properties, we followed Ross-Ross and Hunt and employed the effective stress-strain relations for comparison purposes as though the material behaved isotropically. These authors found that for higher stresses, say σ_t = 20,000 psi in eq. (1), their prediction is about 3 to 6 times lower than the uniaxial results of Fidleris but since their tests did not in fact extend above σ_t = 17,300 psi (σ_{eff} = 15,000 psi) the comparison could not be considered conclusive.

Another paper by Fidleris[16]) reviewed the work on cold-worked Zircaloy-2 previously reported by Fidleris and Williams[12]) and included additional results. Two series of similar studies with cold-worked and heat-treated Zr + 2.5 wt.% Nb with annealed zirconium were also reported. The Zr + 2.5 wt.% Nb tests showed similar trends to those discussed previously for Zircaloy, but the creep rates for Zr + 2.5 wt.% Nb were significantly less than those for Zircaloy under comparable conditions. Fidleris also found that irradiation prior to in-pile testing had negligible effect on the results for low stresses but significantly reduced the (initial) in-pile rates for high stresses comparable to macroscopic yield values (unirradiated), and concluded that the radiation-induced obstacles are more effective at the higher stresses in blocking dislocation movement. He found that if the in-pile creep rate $\dot{\varepsilon}$ is assumed proportional to the stress σ raised to some power n, then the 20% cold-worked Zircaloy samples show an average value of n = 3.3.

Fidleris also reported a decided break in temperature-dependence of in-pile rates in the vicinity of 350°C. Above this temperature the implied activation energy was comparable to that of unirradiated material (50-60,000 cal/mole) whereas below this temperature the value dropped to the order of 20,000 cal/more or less. He gave somewhat conflicting information on flux-effect but concluded that a linear dependence between creep rate and flux is not unreasonable.

Kreyns and Burkart[17]) reported on the in-pile decrease in free curvature of initially bent beams of annealed and cold-worked Zircaloy (as well as Zr-2.5 wt.% Nb-0.5 wt.% Cu in the quenched and aged condition) when irradiated in a fixture so that the beams were held in the flattened position. These tests were conducted at both 310 and 60°C at a flux level of 2×10^{14} nv (>1 MeV). The initial maximum fiber stress employed in these relaxation tests ranged from ~500 to ~20,000 psi. Interpretation of these results was based on an assumption of a linear stress-dependence, which appeared to be approximately valid for the conditions studied.

The tests at 310°C displayed a greater relaxation rate than those at 60°C, but I felt that this could have been simply a reflection of the temperature-dependence of thermal creep, which out-of-pile control relaxation tests showed to be appreciable (especially at the higher temperature) but which is not separable in any straight-forward manner from the total in-pile relaxation rates observed in the bent-beam relaxation samples[18]).

The results showd significantly greater in-pile relaxation rates than those of the out-of-pile control tests. Of further interest is the fact that the cold-worked Zircaloy specimens displayed significantly greater rates of relaxation than did the annealed specimens. In fact, employment of the authors' correlation for long times (when thermal creep should be negligible) indicated that 15% cold-worked Zircaloy has an in-pile creep rate approximately five times that of annealed Zircaloy. A similar comparison indicated that 78% cold-worked material has an in-pile creep rate roughly seven times that of annealed. These were only crude estimates but the empirical observation seemed clear: cold-worked material shows greater in-pile relaxation than does annealed.

This apparent difference between annealed and cold-worked material was displayed again when one examined the large body of data obtained from many different ceramic fuel rods irradiated in various tests of the Bettis Atomic Power Laboratory[19]. In general, these fuel rods are assembled with radial gaps existing at operating temperatures between the fuel and the clad tubing, and so the outside diameter of a rod decreased by creep due to the coolant pressure. In all these tests, the clad temperature was approximately 325°C and the neutron flux $\sim 10^{14}$ nv (>1 MeV). The stresses involved were at two levels, ~12,000 and ~19,000 psi. A large amount of scatter reflected inaccuracies in measurement of the small diametral changes involved as well as the small variations in clad temperature and flux levels for the various tests. Additional uncertainties arose from the fact that the rods had differing initial fuel-clad gaps and operated at various heat-flux levels so that fuel swelling may have closed the gaps at various exposures and thereby inhibited diametral shrinkage. Thus, except for measurement errors, the true creep rates should be reflected by the largest shrinkage values. In spite of these uncertainties, which made quantitative estimates very difficult, it seemed clear that the cold-worked tubing shrinks more rapidly during irradiation than does the annealed. In fact, factors of 5 to 7, as estimated from the relaxation tests, seemed quite reasonable in estimating the increased long-time shrinkage rates of 15% and 78% cold-worked material, respectively, when compared with annealed material. Short-time shrinkage was in good agreement with thermal (out-of-pile) creep data, for both the annealed and the cold-worked tubing; also, long-time, out-of-pile comparative tests demonstrated clearly the enhancements due to the neutron flux.

An observation which seemed very pertinent was reported by Kreyns and Burkart: a sizeable fraction (~70%) of the relaxation strain introduced during irradiation recovered upon post-irradiation annealing at temperatures >450°C. This observation seemed to rule out, as the only operative mechanism, any in-pile creep mechanism for which such recovery is not expected. On the other hand, since the recovery is apparently not complete, it seemed quite likely that there are at least two independent strain contributions, one recoverable and one not. Any strain introduced by Roberts-Cottell yielding creep would not be expected to be recoverable; on the other hand, the production of aligned dislocation loops as mentioned first by Hesketh for cubic materials would be expected to show recovery when the loops shrink during postirradiation annealing.

A similar recovery of growth strains upon annealing of samples irradiated in the unstressed condition led us to conclude that at low stress (which dominated the relaxation behavior) a creep mechanism very similar to the growth phenomenon predominates, with a somewhat smaller contribution from yielding creep.

Gilbert[20]) discussed a series of uniaxial, fixed-stress, in-pile and unirradiated creep tests of Zr + 2.5 at.% Nb specimens at temperatures from 300 to 400°C and stresses from 25.6 kg/mm^2 (36,500 psi) to 38.7 kg/mm^2 (55,000 psi). In agreement with the results of Ross-Ross and Hunt, Gilbert found significantly higher in-pile rates below 350°C but no significant difference between unirradiated and in-pile results at 350°C and above. He concluded that the total creep rate observed in-pile, $\dot{\varepsilon}$, may be expressed as the sum of the unirradiated rate, $\dot{\varepsilon}_u$, and an irradiation component, $\dot{\varepsilon}_i$, in which case it follows that one should utilize only $\dot{\varepsilon}_i$ in discussions of flux, stress and temperature effects for the irradiation-induced creep. This point led him to the conclusion that $\dot{\varepsilon}_i$ shows little if any dependence on stress or temperature. It seemed to me that the conclusion regarding little temperature-dependence was quite reasonable, especially since (as pointed out above) the results of Ross-Ross and Hunt will support the same contention. However, the conclusion regarding little if any stress dependence had to be modified since Ross-Ross and Hunt as well as Fidleris showed a definite stress-dependence.

Hesketh[21]) proposed linear stress and flux dependencies for a model involving preferential alignment of radiation-produced dislocation loops, which he believed to operate in cubic materials (but not Zircaloy) independently of any thermal creep component. He also found it to increase with decreasing temperature and with increasing amounts of cold-work. On the other hand, he presented a generalized theory[22]) of yielding creep based upon the theory of Blackburn[23]) to account for his results obtained with annealed Zircaloy-2 helical springs in a fast flux of 1.3 x 10^{13} nv at 43°C using maximum shear stresses of 3500, 7400 and 8700 psi (corresponding to generalized stresses of 6100, 12,800 and 15,100 psi). His theory proposed a yielding creep due to the buildup of an internal stress from the irradiation growth process. Clearly, as Hesketh pointed out, if irradiation growth occurs in Zircaloy, then yielding creep must of necessity occur. Certainly, irradiation growth had been shown to occur and so should yielding creep. The question of its magnitude compared with other mechanisms was an entirely different matter. It was my opinion that the observation by Kreyns and Burkart of extensive strain recovery for the low-stress region studied by Hesketh ruled out yielding creep as the dominant mechanism of the in-reactor creep of Zircaloy although the occurrence of irradiation growth and the apparent absence of complete strain recovery made it likely that it does contribute some of the in-pile creep strain. Hesketh also pointed out that it was equally possible to resolve the total irradiation creep into additive thermal and irradiation components. I preferred this approach over his yielding creep mechanism since it agreed with the results of other studies as mentioned above. Hesketh also gave a discussion of how irradiation might affect a diffusion-controlled creep process. He discussed the cases of no sinks for point defects other than those responsible for the creep, [grain boundries for Herring-Nabarro creep[24]), climbing edge dislocations for Weertman creep[25]) or jogs in gliding screw dislocations for Barrett-Nix creep[26])], annihilation at fixed sinks and mutual recombination. Only for the last case did he find an effect of irradiation on the concentration gradients of either vacancies or interstitials at the sinks responsible for creep. From this he concuded that irradiation cannot possibly affect diffusion-controlled creep, except when most point defects disappear via mutual recombination, in which case he predicted a strain rate proportional to $(\text{flux})^{1/4}$.

Hesketh's model for Zircaloy produced a complex behavior in which applied stress, temperature, time and neutron flux become interdependent functions of the creep rate. In contrast with this, Ross-Ross and Hunt's results showed a strain rate linearly related to applied stress and flux and almost independent of temperature. These difficulties, plus the presence of extensive strain recovery not predictable by a yielding creep model led me to reject Hesketh's yielding creep proposal as a dominant mechanism for the low-stress region in favor of a straight irradiation growth plus a stress-oriented growth of dislocation loops as originally proposed by Hesketh[21]) for cubic alloys, both acting independently of the thermal creep. This explained the linear dependence on stress and flux; the small temperature dependence also seemed quite plausible.

Piercy[27]) proposed a model based upon the radiation-enhanced gliding of screw dislocations containing jogs and his analysis demonstrated two very important points. First, he showed that for this mechanism an effect of irradiation occurs if and only if the concentrations of sinks for vacancies and interstitials are different. This conflicted with Hesketh's analysis since Hesketh did not allow for this distinct possibility but looked only at the defect concentration gradients at the jogs. This result of Piercy's, though obtained for this specific mechanism, seemed to be a necessity for any diffusion-controlled mechanism. Once steady-state is reached, the rate of disappearance of defects must equal their rate of production; since irradiation produces equal numbers of vacancies and interstitials, this means that unless their sink concentrations differ they must on the average be absorbed in equal numbers at the sinks producing the creep. Any jog, or edge dislocation or grain boundary receiving equal numbers of vacancies and interstitials will undergo no net motion and therefore no enhanced creep occurs.

The second conclusion to which Piercy's analysis led him was that to explain the magnitude of enhanced in-pile creep by his mechanism, he had to assume that the interstitial sink concentration exceeds that of vacancy sinks. (The reverse condition leads to a reduction in creep rate.) This occurs physically because the vacancy-absorbing job is inherently less mobile than the vacancy-producing job and so the former sets the overall dislocation velocity. It follows then that to increase the dislocation velocity requires the net absorption of more vacancies than interstitials; an excess of interstitial sink concentrations assures this. A further related point was that the difference in sink concentrations for the two types of defects must be extremely small ($<0.05\%$) to account for the observations. This likewise seemed an intuitively reasonable result, since for the temperature region $\approx 300^0$C the number of radiation-produced defects greatly exceeds that obtaining thermally.

Although Piercy favored the jog-dragging mechanism, he discussed qualitatively the case of diffusion-controlled climb of edge dislocations. This model possessed the very important characteristic that the creep rate should be enhanced for a difference in sink concentrations, regardless of whether vacancy or interstitial-type sinks are in excess. This is because an equivalent contribution to creep strain occurs whether the dislocations surmount their obstacles by enlarging or reducing the area of their half-planes.

Piercy seemed to base his preference for a jog-dragging model over a climb model on his assumption that this mechanism is applicable thermally. It was difficult for me to see how this can possibly be true. First of all, as Holmes[28]) had shown, a correct analysis of the jog-dragging mechanism leads to a dislocation velocity which reaches a limiting value with increasing stress in strong contradiction to the exponential increase of creep rate with stress displayed by Zircaloy at high stresses at ~300°C, in the annealed, cold-worked and irradiated conditions. Secondly, post-irradiation tests[29]) showed clearly that at ~300°C radiation damage greatly increases the flow stress of Zircaloy down to strain rates as low as $\sim 10^{-8}$/h whether the material is initially annealed or cold-worked. So I concurred with Piercy that a dislocation slip mechanism is required to explain the intermediate-stress results of Fidleris at 300°C but rejected his model of jog-dragging in favor of one involving climb over radiation-produced obstacles.

A final point by Piercy of great importance to me was that a sample which had been stressed in-pile at 270°C to a plastic strain of 0.2% failed to show any strain recovery when annealed up to 380°C. In fact, it showed a small positive strain, thus eliminating preferential dislocation loop alignment as a dominant mechanism. I agreed with Piercy that this mechanism cannot govern the irradiation creep for the stress range in which the Chalk-River tests were run (mostly >15,000 psi). On the other hand, the strain recovery exhibited by the relaxation samples of Kreyns and Burkart whose stress history was mostly in the range <15,000 psi seemed strong evidence for this mechanism at the lower stresses. I was thus led to propose different dominant irradiation creep mechanisms in the low and intermediate stress regions, coupled with a behavior equivalent to post-irradiation behavior (with radiation hardening) in the high stress (strain-rate) region explored by Azzarto et al.[13]). The results of Azzarto et al. also seemed to lend support for a dislocation slip mechanism in the intermediate stress region since they showed the in-pile "strength" of transverse specimens to be significantly higher than that of longitudinal specimens, a phenomenon well-documented for unirradiated, annealed Zircaloy and attributed to the difference in average resolved shear stresses on active slip planes brought about by the texture.

Model Development

a. Low Stress Region

From the fact that the early stages (i.e. low strain, high strain-rate) of low-stress, in-pile creep appear to coincide approximately with out-of-pile creep tests at the same temperature and stress, I assumed that during irradiation the total observed creep rate for low stresses is the summation of that which would occur out-of-pile (i.e. thermal creep), plus an irradiation-growth term, plus an independent radiation creep rate. The in-pile strain rates reported by the various investigators become comparable with that expected for free (i.e. stress-free) growth of Zircaloy as the stress is lowered. Thus, I predicted the form to be

$$\dot{\varepsilon} = \dot{\varepsilon}_{th} + (A + B\sigma)\phi, \text{ (low stresses)}, \qquad (4)$$

where $\dot{\varepsilon}$ = total in-pile creep rate;

$\dot{\varepsilon}_{th}$ = thermal creep rate;

A, B = constants;

σ = stress;

ϕ = fast neutron flux (>1 MeV).

I assumed that $\dot{\varepsilon}_{th}$ is a function of σ, ε, T, nvt, metallurgical structure and texture, and that the constants A and B similarly were functions of T, ε, metallurgical structure and texture. The product $A\phi$ was interpreted as the strain rate in the absence of an applied stress, i.e. $A\phi$ = growth rate. Presumably A, but not B, would be zero for a fully randomly oriented grain structure. In practice, however, this is not attainable and A is relatable to the "f" parameter discussed by Kearns[30]) where f is defined as the fraction of basal poles in a particular direction of the specimen (one-third for a random texture).

he term $B\sigma\phi$ was assumed to arise from the same basic phenomena as give rise to growth. Various growth mechanisms had been proposed but the most plausible appeared at the time to be that due to Buckley[31]). In this model, the growth stains are assumed to arise from the formation of interstitial and vacancy loops preferentially on particular crystallographic planes in response to the anisotropic thermal expansion stresses built up in the spike region of a high-energy damage cascade. For zirconium the assumption that interstitial loops form preferentially on the $\{10\bar{1}0\}$ prism planes and vacancy loops on the $\{0001\}$ basal planes appeared to be the most plausible, as this assumption afforded a good correlation with texture data[30] and also the observation of Bernstein and Gulden[32] of loops tentatively identified as interstitial on the $\{10\bar{1}0\}$ planes of zirconium following bombardment by fast krypton ions. The calculational model developed did not require the validity of these assumptions in detail.

I assumed two contributing components of the $B\sigma\phi$ term. The first was that due to the yielding-creep mechanism originally proposed by Roberts and Cottrell to account for radiation creep of α-uranium[9]).

In its simplest form

$$B\sigma\phi = K(\sigma/\sigma_y)\dot{\varepsilon}_g, \tag{5}$$

where K is a numerical factor, $\dot{\varepsilon}_g$ is the magnitude of the single-crystal growth tensor and σ_y is "the appropriate yield strength of the crystal, i.e. the smallest stress needed to operate those plastic modes that are capable of accommodating the growth strains." In this model, the anisotropic growth strains build up in different directions in the different grains. Due to the restraint of neighboring grains, then, each grain builds up internal stresses until the yield point σ_y is reached. The grains then plastically deform to relieve these stresses but the stresses are continuously replenished by the growth process. The result is that each grain remains stressed to σ_y. In randomly oriented polycrystals, no external dimensional changes occur in the absence of applied stress although with any degree of preferred orientation growth occurs. Now if an external stress is applied it simply "adds" to the σ_y already existing in each grain and so a small stress is capable of "biasing" the dimensional changes by plastic deformation.

Obviously, the Roberts-Cottrell mechanism requires a crystal structure with anisotropic thermal expansion coefficients. Zircaloy met this requirement. However, in-pile creep at very low stresses had been observed for cubic materials as well[4,33,34,35] and in fact cubic nickel for example showed a faster in-pile creep rate than did the hexagonal zirconium. I favored then a mechanism similar to that proposed by Hesketh for cubic materials as dominating the low-stress creep, thereby explaining the nearly linear stress dependence. In this model, the applied stress simply serves to bias the particular planes on which interstitial and vacancy loops form during irradiation, those planes being preferred which give dimensional changes tending to relieve the applied stress. (Hesketh discussed only vacancy loops forming by collapse of damage cascades; I allowed the possibility of both vacancy and interstitial loops as envisioned by Buckley.) For example, if in Zircaloy a tensile stress is applied along the [10$\bar{1}$0] direction the interstitial loops, rather than forming indiscriminately on all three sets of prism planes could form preferentially on the (10$\bar{1}$0) planes which are most nearly normal to the applied stress. I assumed then the relation [eq. (4)] to hold and evaluated A and B from available in-pile data. Hesketh[33]) had proposed a mechanism for a transient in-pile creep effect which could be important in some cases, but there was at the time no evidence for any significant transient phenomenon in Zircaloy, at least above room temperature[8]).

b. High Stress Region

In the discussion above, we centered our attention on low stresses and discussed the in-pile creep behavior as stress is increased from zero. We now wish to consider the opposite extreme in stress, i.e. the macroscopic yield stress at high strain-rate, and examine the effect of lowering the stress. In the following section we shall consider the intermediate stress region.

As mentioned above, the effect of neutron irradiation at T < 300°C on post-irradiation properties is to greatly increase the macroscopic yield stress[36,37]). From the variation of the post-irradiation yield stress (defined as the stress at 0.2% plastic strain) at 290°C as a function of strain-rate compared with results reported by Azzarto et al.[13]) for in-pile tensile-testing of annealed Zircaloy, I concluded that for their flux of 1.2 x 10^{14} nv the in-pile results coincided with the post-irradiation results for strain rates >10^{-4}/h but showed a marked decrease in strength in the vicinity of 10^{-4}/h. Therefore, I assumed that for strain-rates greater than some critical value $\dot{\varepsilon}_c$ the in-pile creep properties approach the thermal (out-of-pile) creep behavior at the same neutron exposure. That is to say, the in-pile properties in this high strain-rate region reflect the radiation-strengthening effect discussed above and thus post-irradiation properties are equivalent to in-pile properties. Specifically, I assumed

$$\dot{\varepsilon} = \dot{\varepsilon}_{th} + (A + B\sigma)\phi, \quad (\dot{\varepsilon} >> \dot{\varepsilon}_c), \qquad (6)$$

which is, of course, identical to eq. (4) for the low-stress region. The different behaviors in these two regions were envisioned as due to the fact that $\dot{\varepsilon}_{th}$ falls off so rapidly as σ is decreased that in the low-stress region the second term becomes dominant except for the early stages of creep; conversely, as σ is increased, $\dot{\varepsilon}_{th}$ increases so rapidly that in the high-stress region the first term becomes dominant. Thus, I assumed that growth and the loop-alignment mechanism discussed above operate at all times, independently of thermal creep, but become completely negligible at high strain-rates (or stresses). Thus, I proposed an in-pile weakening at low stresses and an in-pile strengthening with increasing exposure at high stresses, with exactly the same mechanisms assumed to be operative in both regions.

c. Intermediate Stress Region

Extrapolation of eq. (4) using the data of Ross-Ross and Hunt predicted, for a stress of 40,000 psi, a strain-rate of $\sim 10^{-6}$/h for the flux level of Azzarto's in-pi .ests (1.2×10^{14} nv > 1 MeV), which was approximately two orders-of-magnituu. 'ower than the observed value which suggested a new mechanism to be operative at these highεr stress levels. Also, the variation of $\dot{\varepsilon}$ with σ in this intermediate region was clearly stronger than a liner dependence. In addition to this, the strain recovery phenomenon which exists for lower stresses and not for higher stresses was considered evidence for different mechanisms.

I assumed that the high-stress (or high-strain-rate) strengthening effect was due to the accumulation of damage regions which dislocations must cut through in order for deformation to occur. Since zirconium presumably has a rather high stacking-fault energy[38]) cross-slip may well enable screw segments to easily avoid the obstacles with or without an irradiation flux. However, the only manner in which edge segments may do so is by climb which will be very slow at the temperature of these experiments in the absence of irradiation. The presence of irradiation opens the possibility of enhancing diffusion rates sufficiently for the edge components to climb around the obstacles. Whether or not a certain diffusion rate is sufficient to initiate this mechanism will depend obviously on the rate of deformation. Thus, the decrease in strength in-pile (when compared with post-irradiation properties) at intermediate strain-rate was in qualitative agreement with this concept.

The creep rate to be expected when the rate-controlling process is the climb of dislocations over fixed obstacles had been given by Weertman in a discussion of dispersion strengthening. Assuming that the dislocations pile up at the barriers, Lenel and Ansel[39] gave

$$\dot{\varepsilon} = (\pi\sigma^4\lambda^2 D)/\{8\surd 2)h\mu^3 kT\} \tag{7}$$

where σ = unresolved applied stress;

λ = average spacing between barriers;

D = diffusion coefficient;

h = height of barrier;

μ = shear modulus;

k = Boltzmann constant;

T = absolute temperature.

Similarly, for the case of isolated dislocations climbing over fixed obstacles, Weertman gave[40]

$$\dot{\varepsilon} = (\pi\sigma b^3 D)\{(4\surd 2)kTh^2\}, \tag{8}$$

where b = Burger's vector.

Examination of the experimental results for the intermediate stress region indicated that a σ^4-dependence described the data quite adequately and was definitely superior to a linear relation. I therefore assumed that eq. (7) described the data in this region and then examined the reasonableness of implied values of the parameters entering into the equation.

The theory of radiation-enhanced diffusion had been discussed by several authors[41,42]. The simplest model involved the assumption that the enhanced diffusion is due to the increased steady-state concentration of vacancies created by the damaging flux. The concentration was in turn set by assuming that these vacancies must diffuse to randomly distributed sinks with fixed average inter-sink distance. This gave

$$C^* \approx \dot{N}\tau_v \approx \dot{N}\overline{x^2}/D_v, \tag{9}$$

where C^* = enhanced vacancy concentration;

$\dot{N}$ = rate of vacancy production due to damaging flux;

τ_v = average lifetime of vacancies;

$\overline{x^2}$ = mean-square distance between sinks;

D_v = vacancy diffusion coefficient.

Then, since the atomic diffusion coefficient $D = D_v C_v$, where C_v is the vacancy concentration, one obtained

$$D^* \approx D_{th} + \dot{N}\overline{x^2} \approx D_{th} + \dot{R}\phi, \tag{10}$$

where D^* = in-pile diffusion coefficient;

D_{th} = thermal (equilibrium) diffusion coefficient;

$\dot{R} = \dot{N}\overline{x^2}/\phi$,

and where the assumption was made that the rate of production of vacancies is directly proportional to the flux. I retained the much simpler "effective" diffusion coefficient approach but recognized that D^* is due to combined effects of vacancies and interstitials and should be lower than the value which should be obtained by a straight diffusion measurement where vacancy and interstitial effects add rather than cancelling one another. Inserting eq. (10) into eq. (7) then, my predicted strain-rate for this intermediate region became in the absence of significant thermal creep,

$$\dot{\varepsilon} = (A + B\sigma)\phi + \frac{C\pi\sigma^4\lambda^2 D^*(\phi)}{(8\sqrt{2})h\mu^3 kT}, \tag{11}$$

(intermediate stress and strain-rate),

where C = constant to be experimentally determined and D^* (hence $\dot{\varepsilon}$) was assumed to vary linearly with ϕ. The factor C allowed for the inexactness of the model. The retention of the stress- and flux-dependencies predicted by the model was justified by the fact that they agreed reasonably well with experimental results.

It is important to note that the mechanism proposed in this intermediate region was an alternative or parallel one to that proposed for the high-stress region. That is, the dislocations may either cut through the radiation obstacles or climb around them. Thus, the second term of eq. (11) dominates when it results in a calculated $\dot{\varepsilon}$ much greater than that obtainable by thermal creep (as evidenced by post-irradiation testing) and vice versa.

Also, it is clear that if conditions are such that the climb around radiation obstacles becomes quite rapid, the overall strain-rate cannot be set by the second term of eq. (11). This is because climb per se does not effect a strain in the sample but the dislocations must glide once they have climbed around the obstacles, in order to produce a plastic strain. If follows that if the climb process is rendered "easy" enough (say by radiation-enhancement of diffusion and by sufficiently low overall strain-rate) it ceases to be the rate-controlling step in the climb-glide series. The overall strain-rate will then be determined primarily by the normal glide process of the metal just as if no irradiation obstacles were present.* However, the enhanced diffusion rate, if sufficient to overcome irradiation obstacles, should also be sufficient to overcome the "weaker" strain-hardening obstacles. (We assumed the overcoming of strain-hardening obstacles to be much easier and hence not rate-controlling except at very low exposures.) The resulting strength then should be that characteristic of the metal with some minimal amount of plastic strain. To account for this effect, we introduced σ_{int} which we called the intrinsic strength of the alloy, but which was expected to be representative of the flow stress of unirradiated, annealed Zircaloy at a very low plastic strain value to be determined by experiment. When a specimen arrived at a stress and strain-rate characteristic of this intrinsic strength, its subsequent strain-rate contribution [over and above the $(A + B\sigma)\phi$-term] would remain at the same value if the stress, temperature and flux values remained unaltered. This effect then operated in series with the parallel cutting-climbing process, and the intrinsic strength was viewed as that flow stress required for the glide of dislocations in the long-range internal stress field alone, with strain-hardening obstacles being rapidly eliminated (or overcome) by the enhanced diffusion rates.

The in-pile data of Azzarto et al.[13]) suggested that, for the same stress, annealed Zircaloy tested in the direction parallel to the rolling direction (longitudinal) has a creep rate approximately four times that of material tested in the transverse direction. This was assumed to be due to texture-induced differences in resolved shear stresses on active slip planes and as such was considered additional evidence for a dislocation slip mechanism; a similar directionality occurs in unirradiated material[44]).

c. General Proposed Model

Collecting the results discussed above, the prediction for the in-pile stress, strain, strain-rate dependencies of either annealed or cold-worked Zircaloy, at all flux levels and temperatures near 300°C over the complete

*The situation was considered similar to that proposed by Ardell and Sherby[43]) in interpreting the out-of-pile, high temperature creep of zirconium to be diffusion-controlled at low stresses and glide-controlled at high stresses. The nature of their climb obstacles was of course different and diffusion was *not* enhanced in their case, so that *normal* strain-hardening could occur.

range of stress from zero to the post-irradiation, fast strain-rate flow stress of highly irradiated Zircaloy was developed in the following form

$$\dot{\varepsilon} = [A(\text{structure, orientation, } T, nvt) + B(\text{structure, orientation, } T, nvt)\sigma]\phi + \dot{\varepsilon}_{int}(\dot{\varepsilon}_{th} + \dot{\varepsilon}_{climb})/(\dot{\varepsilon}_{th} + \dot{\varepsilon}_{climb} + \dot{\varepsilon}_{int}), \quad (12)$$

where $\dot{\varepsilon}_{int} = \dot{\varepsilon}_{int}(\text{orientation, } T, \sigma)$;

$\dot{\varepsilon}_{climb} = C'\sigma^4\phi$;

$\dot{\varepsilon}_{th} = \dot{\varepsilon}_{th}$ (structure, orientation, T, nvt, ε, σ),

and where A, B, C', $\dot{\varepsilon}_{int}$ and $\dot{\varepsilon}_{th}$ were to be determined experimentally. I dropped the explicit formulation of the climb model in view of its inexactness; thus, for example, no significance was attached to the T^{-1} relationship of eq. (7), since effects not accounted for may well have stronger T-dependencies.

Eq. (12) was formulated into a computer program, FLIC (FLux-Induced-Creep),* to treat an arbitrary (σ, T, ϕ)-history and determine the resulting deformations under arbitrary uniaxial, biaxial or triaxial loading conditions. The origin of the manner in which $\dot{\varepsilon}_{int}$, $\dot{\varepsilon}_{th}$ and $\dot{\varepsilon}_{climb}$ were combined in eq. (12) lay in the assumption, discussed above, that the thermal cutting process and the climb process were alternative creep mechanisms so that $\dot{\varepsilon}_{th}$ and $\dot{\varepsilon}_{climb}$ add (like currents in a parallel circuit). This parallel combination must however act in series with the "in-between" glide process since the overall rate can never exceed $\dot{\varepsilon}_{int}$. This led to the assumption that $(\dot{\varepsilon}_{int})^{-1}$ which measures the "resistance" of this process must be added to the effective "resistance" of the cut-climb parallel network $(\dot{\varepsilon}_{th} + \dot{\varepsilon}_{climb})^{-1}$ to obtain the overall "resistance" $(\dot{\varepsilon}_{overall})^{-1}$. Thus

$$\frac{1}{\dot{\varepsilon}_{overall}} = \frac{1}{\dot{\varepsilon}_{int}} + \frac{1}{\dot{\varepsilon}_{th} + \dot{\varepsilon}_{climb}}, \quad (13)$$

which when solved for $\dot{\varepsilon}_{overall}$ and combined with the independent contribution $(A + B\sigma)\phi$ gives eq. (12). Such a formulation was essential in order to follow

*Later a modular form called ZIRMOD was written and served as a general-purpose properties module for various structural analysis programs.

wide variations in the various parameters. When one process in the parallel network, say $\dot{\varepsilon}_{climb}$, greatly exceeds the other but is much smaller than $\dot{\varepsilon}_{int}$ then eq. (13) reduces to simply $\dot{\varepsilon}_{climb}$ which is the situation when the climb process is said to be rate-controlling. This will occur after some amount of strain hardening. On the other hand, when a sample is initially loaded in-pile the thermal component $\dot{\varepsilon}_{th}$ will exceep $\dot{\varepsilon}_{climb}$ and the overall rate will be essentially the same as that obtained out-of-pile at the same neutron exposure. As straining progresses $\dot{\varepsilon}_{th}$ decreases until finally $\dot{\varepsilon}_{climb}$ sets the overall rate, providing $\dot{\varepsilon}_{climb} \ll \dot{\varepsilon}_{int}$ obtains.

Summary

The overall behavior which emerged is schematically illustrated in the attached figure in the form of a log-stress vs. log-strain rate plot for a specific temperature, flux and fluence level. The line labeled "INT" (intrinsic) represents the behavior of unirradiated, annealed Zircaloy at low strains. Increasing strain (or cold-working) and/or fluence produces a family of roughly parallel curves at rising stress levels. After irradiation to a high fluence level, the line labeled "CUT" represents the cutting by dislocations of radiation-produced obstacles, with little effect of strain level, reflecting the drastic reduction in strain hardening following irradiation. This line then displays the well-known radiation-hardening effect. Radiation hardening as well as its thermal recovery were included, using available theoretical and empirical representations. Thermal deformation was included via an effective-strain-hardening model, also including thermal recovery and anisotropy.

All other lines reflect deformation processes (also anisotropic) which occur only <u>during</u> concomitant irradiation. The two-segment line* labeled "CLIMB" represents the flux-enhanced climb of dislocations to overcome radiation-produced obstacles whereas the "INT" line plays a secondary role representing the consequent glide of dislocations to the next obstacle. Since glide and climb must occur in sequence the "slower" of the two controls the overall rate. This contrasts with the "CUT" process where stress-induced escape from obstacles operates in parallel with climb-induced escape so that

*Although initially only a "σ^4 region was employed, later data[45]) indicated a linear-to-fourth-order transition using Eqs. (8) and (7) was more appropriate.

the "faster" sets the rate. The figure also shows the processes of low-stress radiation creep not involving long-range dislocation glide and hence acting independently and additively. The line representing this process is labeled "LOOP" because I proposed that the stress-induced preferential alignment of dislocation loops due to flux-produced point defects caused this creep component. Finally, the figure illustrates, by the line marked "GROWTH," the flux-induced dimensional change or radiation growth occurring in the absence of applied loads. The dashed curve shows how these various processes were combined via a series-parallel electrical-circuit analog to produce the expected overall behavior.

Data comparisons and setting of the various parameters was an extensive process and with the FLIC program was accomplished using actual in-reactor conditions which often vary significantly even in the best of in-pile tests. These data comparisons are available[42]) in the literature and will not be repeated here. Instead, we give only an indication of the various types of tests used to establish the magnitudes for the various deformation regimes.

The "CUT" line was supported at high strain rates by post-irradiation tensile tests and at low strain rates by post-irradiation creep tests. The "CLIMB" and transition "INT" regions were supported by in-pile creep tests and the transition from "CLIMB" to "CUT" was supported by in-pile "fixed-strain-rate" tests. The "LOOP" regime was supported by in-pile internally pressurized tube creep tests and bent-beam, stress-relaxation tests. Finally, the "GROWTH" regime was supported by in-pile dimensional changes of unloaded specimens.

Approximate values for the slopes of the various deformation processes are indicated in the figure by the m-values, the so-called strain-rate sensitivity. Building on an analysis by Hart[48]) showing the controlling influence of this parameter on tensile ductility during steady-state deformation I developed the "Silly-Putty" model[46]) for Zircaloy fracture predictions. Thus, at high strain rates following significant radiation $m \to 0$ and ductility is extremely limited so that a stress-limit design approach was appropriate. Conversely, the relatively high value of m during irradiation creep indicated a rather strong resistance to tensile instability ("necking") failures and hence large strains (>20%) were predicted before failure so that a large strain-limit approach to design was feasible. These concepts became the basis for failure analysis and design limits of the LWBR (Light Water Breeder Reactor).

Conclusions

I have neither the knowledge nor the space to give in this paper the details of the development of related computer programs and subsequent implementations of the concepts discussed here in various applications throughout the LWBR developmetnn program. Much of this work is not available in the open literature and I have appended a bibliography listing only a sampling of pertinent internal Bettis reports.

It was very exciting to kind of relive those rapidly changing events in the sixties which gave rise to my receiving the 1984 W. J. Kroll Zirconium Medal. I am greatly indebted to many other people who played very significant roles and to the combination of events which happily placed me in the right place at the right time. But in fact I personally give the praise to God who makes life worth living and provided (and continues to provide, for me at any rate) the motivation for trying to understand and put to use just a tiny bit of the marvelous world he created and, amazingly, put under our stewardship.

References

1) See, e.g., AIME Conf. Radiation Effects (September 1965, Asheville, NC, USA; Gordon Breach) 1969

2) Information obtained from P. W. Patterson, Bettis Atomic Power Laboratory, West Mifflin, PA, USA (1968)

3) Information obtained from R. H. Fillnow, Bettis Atomic Power Laboratory, West Mifflin, PA, USA (1968)

4) K. P. Dubrovin, S. T. Konobeyevskiy, B. M. Levitskiy, L. D. Panteleyev, P. A. Platanov and N. F. Pravdyuk, in Properties of Reactor Materials and the Effects of Radiation Damage (ed. D. J. Littler; Butterworths, London, 1962) p. 233

5) G. Schoeck, J. Appl. Phys. 29 (1958) 112

6) J. A. Brinkman and H. Wiedersich, in Flow and Fracture of Metals and Alloys in Nuclear Environments, ASTM, Spec. Techn. Publ. No. 380, Philadelphia (1965) 3

7) D. Mosedale, J. Appl. Phys. 33 (1962) 3142

8) Information obtained from R. V. Hesketh, Berkeley Nuclear Laboratories, England (1967)

9) A. C. Roberts and A. H. Cottrell, Phil. Mag. 1 (1956) 711

10) Information obtained from L. A. Tarshis, Bettis Atomic Power Laboratory, West Mifflin, PA, USA (1964)

11) J. J. Holmes, J. A. Williams, D. H. Nyman and J. C. Tobin, in Flow and Fracture of Metals and Alloys in Nuclear Environments, ASTM, Spec. Techn. Publ. No. 380, Philadelphia (1965) 385

12) V. Fidleris and C. D. Williams, in Zirconium and Its Alloys (Ed. J. P. Pemsler, E. C. W. Perryman and W. W. Smeltzer; Electrochemical Soc., NY, 1966) p. 258

13) F. J. Azzarto, E. E. Baldwin, F. W. Wiesinger and D. M. Lewis, J. Nucl. Mat. 30 (1969) 208

14) P. A. Ross-Ross, Trans. Amer. Nuclear Soc. 10 (1967) 114

15) P. A. Ross-Ross and C. E. L. Hunt, J. Nucl. Mat. 26 (1968) 2

16) V. Fidleris, J. Nucl. Mat. 26 (1968) 51

17) P. H. Kreyns and M. W. Burkart, J. Nucl. Mat. 26 (1968) 87

18) G. W. Lewthwaite and D. Mosedale, Brit. J. Appl. Phys. 17 (1966) 821

19) H. B. Meieran, Bettis Atomic Power Laboratory, West Mifflin, PA, USA (1967) unpublished results

20) E. R. Gilbert, J. Nucl. Mat. 26 (1968) 105

21) R. V. Hesketh, Proc. Internat. Conf. Solid State Physics Research with Accelerators, Brookhaven National Laboratory, Report BNL 500.83 (C-52), (1968) 389

22) R. V. Hesketh, J. Nucl. Mat. 26 (1968) 77

23) W. S. Blackburn, Phil. Mag. 6 (1961) 503

24) Conyers Herring, J. Appl. Phys. 21 (1950) 437

25) J. Weertman, J. Appl. Phys. 26 (1955) 1213

26) C.R. Barrett and W. D. Nix, Acta Met. 13 (1965) 1247

27) G. R. Piercy, J. Nucl. Mat. 26 (1968) 18

28) J. J. Holmes, Acta Met. 15 (1967) 570

29) P. W. Patterson, Bettis Atomic Power Laboratory, West Mifflin, PA, USA (1968) unpublished research

30) J. J. Kearns, Bettis Atomic Power Laboratory, (USA), Report WAPD-TM-472, (1965)

31) S. N. Buckley, in Properties of Reactor Materials and the Effects of Radiation Damage (ed. D. J. Littler; Butterworths, London, 1962) p. 413

32) I. M. Bernstein and T. D. Gulden, in Report of a Consultants Symposium (AERE, Harwell, July 1966) UKAEA Report AERE-R5269, 2, p. 441

33) R. V. Hesketh, Phil. Mag. 8 (1963) 1321

34) R. L. Mehan, J. Basic Eng. 81 (1959) 226

35) R. V. Hesketh, 10th Colloque de Métallurgie (1967) 186

36) A. L. Bement, Jr., AIME Conf. Radiation Effects (September 1965; in Asheville, NC, USA; Gordon-Breach) 1969

37) A. D. Whapham and M. J. Makin, Phil. Mag. 5 (1960) 237

38) R. B. Roy, Aktiebolaget Atomenergie (Sweden) Report AE-130 (1963)

39) F. V. Lenel and G. S. Ansell, in Powder Metallurgy (Interscience, NY, 1961) p. 267

40) J. Weertman, J. Appl. Phys. 26 (1955) 1213

41) G. J. Dienes and A. C. Damask, J. Appl. Phys. 29 (1958) 1713

42) R. E. Larsen and A. C. Damask, Acta Met. 12 (1964) 1131

43) A. J. Ardell and O. D. Sherby, Trans. Met. Soc. AIME 239 (1967) 1547

44) C. R. Woods, Bettis Atomic Power Laboratory, (USA) Report, WAPD-TM-585 (1966)

45) V. Fidleris, J. Nucl. Mater. 36 (1970) 343

46) F. A. Nichols, Mater. Sci. Eng. 6 (1970) 167

47) F. A. Nichols, J. Nucl. Mater. 30 (1969) 249

48) E. W. Hart, Acta Met. 15 (1967) 351

Bibliography

1. WAPD-TM-514, "CYGRO-Stress Analysis of the Growth of Concentric Cylinders," Oct. 1965.

2. WAPD-TM-514 Addendum 1, July 1965.

3. WAPD-TM-547, "CYGRO-2 A Fortran IV Computer Program for Stress Analysis of the Growth of Cylindrical Fuel Elements With Fission Gas Bubbles," Nov. 1966.

4. WAPD-TM-583, "Comparisons With Experiment of Calculated Dimensional Changes and Failure Analys . of Irradiated Bulk Oxide Fuel Test Rods Using the CYGRO-1 Comput.. Program," Sept. 1966.

5. WAPD-TM-583 Addendum 1, April 1967.

6. WAPD-TM-591, "Deformation and Collapse of Fuel Rod Cladding Due to External Pressure," Jan. 1967.

7. WAPD-TM-756, "Estimated Creep Properties of Zircaloy-4 during Neutron Irradiation," May 1968.

8. WAPD-TM-794, "Analysis of Void Migration, Clad Collapse, and Fuel Cracking in Bulk Oxide Fuel Rods," July 1968.

9. WAPD-TM-961, "CYGRO-3 - A Computer Program to Determine Temperatures, Stresses and Deformations in Oxide Fuel Rods," March 1970.

10. WAPD-TM-973, "In-Pile Dimensional Changes of Zircaloy-4 Tubing Having Low Hoop Stress," July 1971.

11. WAPD-TM-1094, "Postirradiation Recovery of Irradiation Damage," Aug. 1973.

12. WAPD-TM-1300, "The CYGRO-4 Fuel Rod Analysis Computer Program, July 1977.

13. WAPD-TM-1338, "Cladding Deformation into Pellet End Voids," Nov. 1981.

14. WAPD-TM-1339, "Analyais of Cladding Deformation Over Plenum Axial Gaps in Zircaloy Fuel Rods," Feb. 1983.

15. WAPD-TM-1404, "Ex-Reactor Deformation of Externally Pressurized Short Lengths of Fuel Rod Cladding," May 1979.

16. WAPD-TM-1509, "Thermal Tests to Investigate Stability of Externally Pressurized Zircaloy-4 Tubing Over Axial Gaps," Aug. 1982.

17. WAPD-TM-697, "Large Deflection Theory for Inelastic Stresses and Strains," Aug. 1967.

18. WAPD-TM-712, "A Numerical Method for Determining Strain Distributions in a Plastic or Creep Material Under Incremental Load," Sept. 1967.

19. WAPD-TM-731, "GAPL-1, A Computer Program for the Plastic Stress Analysis of a Horizontal Plate With Pressure Loading Above and Deflection Restraints Below," Jan. 1970.

20. WAPD-TM-791, "GAPL-3: A Computer Program for the Large Deflection Plastic Stress Analysis of a Plate, Cylinder, or Disc With Pressure Loading Above and Deflection Restraints Below," June 1969.

21. WAPD-TM-890, "BUSHL - A Computer Program for the Inelastic Buckling of Shells of Revolution Under External Pressure and Axial Compression," Mar. 1971.

22. WAPD-TM-984, "Improved First Order Analysis of the Plastic Stability of Diffuse Patches in Tubes and Sheets," Aug. 1973.

23. WAPD-TM-1384, "Finite Deformation Analysis of Continuum Structures With Time-Dependent Anisotropic Elastic Plastic Behavior," Mar. 1980.

24. WAPD-TM-746, "Radiation Enhanced Relaxation in Zircaloy-4 and Zr/2.5 w/o Nb 10.5 w/o Cu Alloys," May 1968.

25. WAPD-TM-943, "Phenomenological Model of In-Pile Stress Relaxation," Aug. 1971.

26. WAPD-TM-1027, "Irradiation Creep Based on Dislocation Climb," Feb. 1973.

27. WAPD-TM-1043, "High Temperature, Time-Dependent Deformation in Internally Pressurized Zircaloy-4 Tubing," Oct. 1974.

28. WAPD-TM-1066, "Viscoelastic Analysis of In-Pile Stress Relaxation," Mar. 1973.

29. WAPD-TM-1123, "Irradiation Induced Primary Creep," Apr. 1974.

30. WAPD-TM-1396, "External Pressure Test and Deformation of a Zircaloy-4 Hexagonal Shell," Feb. 1984.

31. WAPD-TM-1536, "Irradiation Growth of Zircaloy," Nov. 1982.

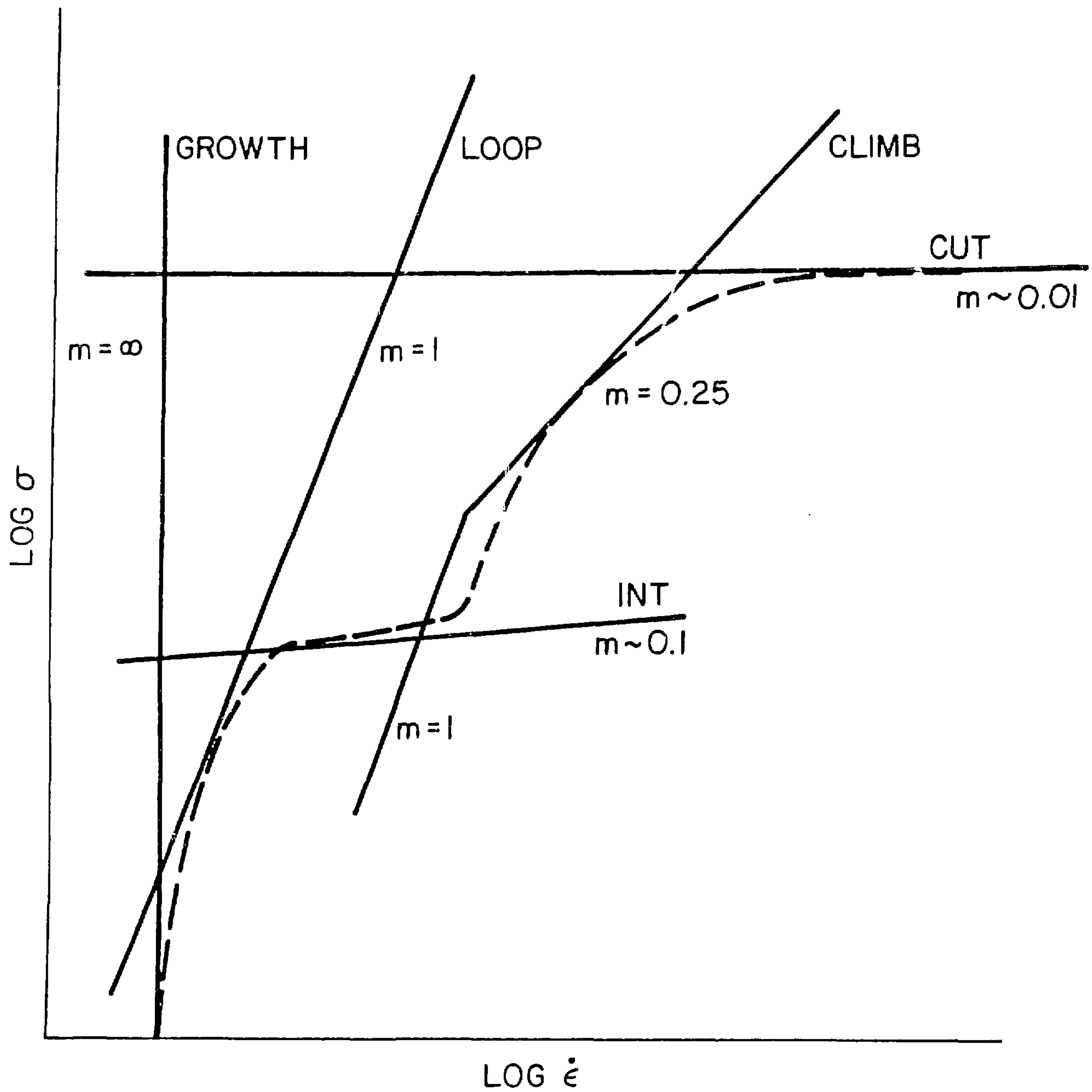
32. WAPD-TM-808(L), "Prediction of the Reactor Deformation of Zircaloy Cladding for Oxide Fuel Rods," Dec. 1969.

33. WAPD-TM-929(L), "Deformation and Failure of Zircaloy Cladding on Oxide Fuel Rods During Neutron Irradiation," Apr. 1970.

34. WAPD-TM-992(L), "Basis for Mechanistic Model for Irradiation Growth of Zircaloy," July 1971.

35. WAPD-TM-1166(L), "In-Pile Length Changes of Zircaloy-Clad Oxide Fuel Rods," Oct. 1973.

36. WAPD-TM-1167(L), "In-Pile Diameter Changes of Zircaloy-Clad Oxide Fuel Rods," Oct. 1973.



Schematic Representation of Mechanical Deformation Model for Zircaloy Including Radiation Effects.