PNL-SA-21235

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FACE-CENTERED CUBIC METALS**

F. A. Garner
M. B. Toloczko

June 1993

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JUL 27 1993
OSTI

Presented at the
Time Dependence of Radiation
Damage Accumulation and Its Impact
on Material Properties
October 14-21, 1993
Montreal, Switzerland

Prepared for
the U.S. Department of Energy
Contract DE-AC06-76RLO 1830

Pacific Northwest Laboratory
Richland, Washington 99352

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HIGH DOSE EFFECTS IN NEUTRON IRRADIATED FACE-CENTERED CUBIC METALS

F.A. Garner¹ and M.B. Toloczko²

¹ Pacific Northwest Laboratory, Richland WA, USA.

² University of California, Santa Barbara CA, USA.

Abstract

During neutron irradiation, most face-centered cubic metals and alloys develop saturation or quasi-steady state microstructures. This, in turn, leads to saturation levels in mechanical properties and quasi-steady state rates of swelling and creep deformation. Swelling initially plays only a small role in determining these saturation states, but as swelling rises to higher levels, it exerts strong feedback on the microstructure and its response to environmental variables. The influence of swelling, either directly or indirectly via second order mechanisms, such as elemental segregation to void surfaces, eventually causes major changes, not only in irradiation creep and mechanical properties, but also on swelling itself. The feedback effects of swelling on irradiation creep are particularly complex and lead to problems in applying creep data derived from highly pressurized creep tubes to low stress situations, such as fuel pins in liquid metal reactors.

1. Introduction

The microstructure of metals is strongly altered during neutron irradiation to high displacement levels, especially at higher temperatures where radiation-produced vacancies and interstitials are both highly mobile. At relatively low atomic displacement levels, the evolution of microstructure is transient in nature as new microstructural components appear, and pre-

existing components are altered in distribution and density. At some point, however, a quasi-steady state or "saturation" microstructure is reached in which most important microstructural components reach saturation densities.(1) For the purpose of this review, high displacement levels are considered to be those that place the metal beyond the transient and into the quasi-steady state or saturation regime. For relatively simple metals, this dose may be several displacements per atom (dpa) or less, depending on the irradiation temperature, while more complex solute-bearing alloys may require doses on the order of 50-150 dpa to reach the "high dose" regime. The evolution in complex alloys is usually more protracted due the additional complexity associated with the interaction of both microstructural and microchemical processes. The microchemical evolution is not only more sluggish, but often requires the prior formation of microstructural components such as network dislocations, Frank dislocation loops, or precipitates.

In a recent review of this subject, it was shown that when face-centered cubic (FCC) metals and alloys are irradiated at elevated temperatures, they evolve toward saturation microstructures that are independent of the starting state of the metal and also independent of the early details of irradiation history (1). Early detours in temperature or flux history are usually quickly forgotten by the microstructure, with the saturation state determined primarily by the final irradiation environment. This tendency toward saturation leads to macroscopic property changes and rates of macroscopic dimensional change that also eventually become independent of the starting microstructure and the temperature history.

Until recently, however, it has generally been assumed that the saturation states reached both in macroscopic properties and in rates of

dimensional change would persist indefinitely. It now appears that these saturation levels are actually metastable, as second order processes operating over long exposure periods begin to overwhelm the first order factors that determined the original saturation conditions. These initially second-order processes exert feedback on the microstructure that can lead to a new late-term saturation state or even to the loss of saturation behavior. The majority of these feedback mechanisms appear to be associated either directly or indirectly with "large levels" of void swelling, although the stress state and various transmutation processes may also be involved. The term "large swelling levels" is also relative, with the critical swelling magnitude varying for different feedback processes and for different irradiation environments. Since FCC metals and alloys are more prone to high swelling rates than are most body centered cubic (BCC) and hexagonal close packed (HCP) metals and alloys, this review will concentrate primarily on the evolution of FCC materials at high dpa levels.

2. Feedback of swelling into the instantaneous swelling rate

For most FCC alloys, the void growth process at a given set of irradiation conditions approaches a saturation rate that persists to swelling levels beyond that considered to be relevant for either fundamental or technological considerations. It has been postulated, however, that large volume fractions of either voids or bubbles will alter the balance of microstructural sinks so as to eventually induce a decline in the swelling rate, possibly even reducing it to zero.(2,3) In simple ternary Fe-Cr-Ni alloys and most typical austenitic commercial steels (see Fig. 1 [top] and Fig. 2 for examples), however, there is no indication that neutron-induced

swelling over most of the typical temperature regime of void growth will either decline in rate or attain saturation. (An exception to this general rule will be shown in the next section, however.)

There is some indication, however, that saturation of swelling can occur more easily during ion bombardment under certain conditions. In one proton irradiation study conducted on 316 stainless steel, saturation was attained, but only after ~260% swelling had occurred.(6) As shown in Fig. 3, this same steel also reached saturation at only 20% swelling in another study. This relatively low level of saturation occurred in a Ni^+ ion bombardment experiment conducted with high levels of simultaneous implantation of both helium and deuterium.(7) This suggests that higher cavity densities formed under such conditions tend to promote the onset of saturation. Since the sink strength of voids is defined in terms of their surface area rather than their volume, gas-enhanced cavity densities and their associated larger surface areas would indeed be expected to induce saturation at lower swelling levels.(2,3)

One other set of conditions exists under which saturation might occur during neutron irradiation in the absence of large gas levels. At lower irradiation temperatures, the cavity densities are usually much higher, and saturation-prone cavity-dominated microstructures might evolve for sufficiently high enough void levels. Fig. 1 (bottom) illustrates this possibility in relatively soft Fe-Cr-Ni ternary alloys.

Other methods exist by which voids can participate in limiting or reducing the swelling rate at higher displacement doses. If void surfaces at higher swelling levels serve as internal annihilation sites for dislocations, and thereby cause a collapse in the dislocation density, swelling will tend to

saturate at relatively low levels. Collapse can occur most easily in metals with a high stacking fault energy, which precludes the easy formation of partial dislocations with their higher level of pinning. This makes it harder to stabilize dislocation networks.(8) An example is shown in Fig. 4 in a study involving neutron-irradiated pure nickel. As shown elsewhere, saturation of swelling in pure nickel is a very general occurrence, observed during both neutron and charged particle irradiations.(9) Saturation does not occur as easily in metals with lower stacking fault energies.

The formation of ordered void lattices also appears to lead to saturation of swelling, and under some circumstances, this can actually cause swelling to decrease via void shrinkage,(10) as shown in Fig. 5 for molybdenum. The saturation level appears to be independent of the microstructural starting state, however. The overshoot above the saturation level of swelling under some conditions appears to be associated with establishing a high density of dislocation subgrain walls by prior deformation. This induces early formation of randomly distributed voids within the small subgrains, followed by void shrinkage during ordered lattice formation.(11)

Void lattices in general do not form as easily in FCC metals but they have been observed in several FCC metals during ion bombardment at displacement rates much greater than those encountered in typical reactor irradiations.(12) Numerous mechanisms have been proposed for void lattice formation and the associated saturation of swelling (see review by Krishan [12] but Evans has suggested that formation of technetium by transmutation and its subsequent segregation to void surfaces may be responsible for void shrinkage in molybdenum.(13)

In other experiments, segregation of transmutants to void surfaces has been shown to reduce the swelling rate. The best example was provided by Farrell and coworkers when silicon formed and segregated to void surfaces in aluminum alloys during irradiation in the High Flux Isotope Reactor (HFIR). (14,15) Segregation of transmutants *per se* to void surfaces does not imply that swelling rates will be reduced, however. Nickel formed by transmutation of pure copper is known to segregate to void surfaces during irradiation (16,17), but neither its formation nor its segregation appear to affect the swelling rate of copper. (18)

3. Feedback of Swelling into Irradiation Creep Rates

Irradiation creep offers the best demonstration of feedback effects, with feedback arising from both void swelling and dislocation development in response to imposed stress states. As a consequence of the relatively recent recognition of feedback's very strong role on irradiation creep, it is now known that the creep phenomenon is much more complex than previously envisioned. Four separate regimes of irradiation creep appear to exist, described here as "primary" or transient creep, "secondary" or steady state creep in the absence of swelling, swelling-enhanced creep, and feedback-dominated or "disappearing" creep. "Tertiary" creep associated with high temperature thermal creep will not be discussed in this paper.

When compared to thermal creep in the absence of irradiation, the transient creep regime during irradiation is in general characterized by a very high creep rate, usually referred to as primary creep, that declines with continued irradiation. The creep rate eventually approaches a "steady-state" rate that persists as long as void swelling does not occur. Earlier

literature usually refers to this stage as secondary creep. Fig. 6 shows the primary and secondary regimes for pure copper and aluminum irradiated in the annealed condition with neutrons at relatively low displacement rates.(19) Fig. 7 shows that similar behavior was observed in a solute-modified commercial steel irradiated in the cold worked condition in two different stress states.(20)

In each case, the initial microstructure undergoes a stress-affected evolution, eventually reaching a saturation state. The duration of the approach to the saturation state depends not only on the stress level, but also on its orientation with respect to any pre-existing microstructure. The deformation-induced texture of cold worked steels and its orientation to the stress state is quite different for bent beam and pressurized tube studies, and therefore transient regimes of different magnitude occur as shown in Fig. 7, but the secondary creep rates at a given stress level are quite similar.

In both annealed and cold worked steels, the transient regime is characterized by the development of a microstructure that is specific to both the stress state and the irradiation conditions. As described elsewhere, the primary feature of this evolution is the replacement of the original microstructure with a dislocation and loop microstructure characterized by an anisotropic distribution of Burger's vectors.(21,22)

In materials that begin to swell without exhibiting a significant void incubation period, it is very obvious that the irradiation creep rate is enhanced whenever swelling occurs. Fig. 8 demonstrates such behavior for nickel at two purity levels. Although the swelling behavior of the two starting conditions is very different, the instantaneous creep rate appears to be directly proportional to the instantaneous swelling rate.(23) When the

swelling rate of 99.99% pure nickel declines as swelling approaches its saturation level, the creep rate also declines, indicating that a strong coupling exists between void swelling and irradiation creep. This coupling is maintained regardless of whether the swelling rate is increasing or decreasing.(24)

A similar coupling during periods of both increasing or decreasing swelling rates was observed in a more complex alloy designated DIN 1.4864.(25) This alloy was observed to exhibit an incubation period before swelling and then to begin swelling at a significant rate. Swelling thereafter began to decline in rate and to tend toward saturation upon further irradiation. As shown in Fig. 9, the creep rate exhibited the same behavior. Note that the swelling strain is somewhat sensitive to stress, but saturation appears to be at ~1% for all stress levels studied. This corresponds to only ~3% volumetric swelling. It is not clear whether this relatively low level of swelling exerted feedback directly on the termination of the swelling process or whether swelling was terminated by some other process. Austenitic alloys with this nickel level (~35%) have exhibited some very unusual spinodal-like segregation behavior during irradiation, and this may have contributed to the saturation of swelling.(26,27)

In most commercial steels, however, an extended incubation period precedes the onset of steady-state swelling, which then proceeds to very high swelling levels. Irradiation creep accelerates in turn as the swelling rate accelerates and saturates in rate as swelling reaches its saturation rate,(28) as shown in Fig. 10. Although the width of the stress-normalized creep bands in Fig. 10 appear to vary slightly with increasing stress level, this behavior does not reflect a non-linear dependence of creep rate on stress level.

Instead, it is a consequence of the well-known stress-induced shortening of the incubation period of swelling.(29,30) The actual swelling rate in these experiments cannot be measured without destroying the pressurized tubes, but it is always somewhat larger than the stress-free level used to derive Fig. 10. In this particular incidence, where the stress-free and stress-affected swelling rates are far below the swelling rate saturation level, availability of the actual stress-affected swelling rates would show the calculated creep rate to be proportional to the applied stress, and consequently, the width of the stress-normalized creep band would be much smaller.

The maximum swelling shown in Fig. 10a is only ~7.5%, and the swelling rate ($\leq 0.1\%/dpa$) in this temperature range (380-415°C) is still far below the 1%/dpa maximum swelling rate of this steel observed at higher temperatures.(31) As the swelling rate and swelling level increase, however, the feedback effects of swelling and stress become even more pronounced. At some point, however, the feedback begins to reverse in sign, and creep seems to "disappear." The creep rate first increases, but then begins to fall toward zero (32) as shown in Fig. 11, and the creep strain appears to saturate (33) as shown in Fig. 12. In the latter figure, the volumetric swelling rate at ~530°C has reached ~0.7%/dpa, and the total diametral (linear) strain rate is 0.24%/dpa.(32) The tendency for the creep strain to lose its dependence on stress and to saturate is also shown in Fig. 13. In this case, the volumetric swelling rate has reached the 1%/dpa limit and accounts fully for the observed 0.33% diametral strain rate at the highest dpa levels.(32,34)

One consequence of the creep disappearance phenomenon is that creep predictions based on pressurized tubes at high stress levels will always fail to predict creep strains in fuel pin cladding, which operates at initially low

but slowly increasing stress levels. In the highly pressurized tube case, creep occurs before swelling starts, while in the fuel pin, swelling starts before significant creep occurs. This is an important distinction, especially in its consequences on dislocation development and the anisotropy of Burger's vectors.

Fig. 14 shows that creep correlations developed from pressurized tubes irradiated in the French liquid metal reactor (LMR) program predict reasonably well the strains observed in hexagonal ducts irradiated in PHENIX at moderately high and relatively constant stress levels, but Fig. 15 demonstrates that the predicted strains of fuel pin cladding constructed from the same steel and irradiated in the same subassembly are roughly a factor of three too low.(35) A similar problem in predicting of cladding strains was also observed in the U.S. LMR program, as shown in Fig. 16.(36) In this case, the observed creep strains are much smaller than expected, however. The apparent difference in the sign of the misprediction in the French and U.S. creep data is a consequence of differences in the two programs in (a) the manner in which the creep experiments were conducted and (b) the way in which the creep correlations were developed. The origin of both types of mispredictions is the disappearing creep phenomena, however.

The strongly coupled but varying relationship between swelling and creep can be observed even when the temperature and flux profiles along fuel pins yield very complex swelling profiles (37), as shown in Fig. 17. Note, however, that the relative magnitudes of the swelling and the creep modulus of the French cold-worked fuel pin cladding are distorted as the temperature increases. This distortion was also observed to a different degree in annealed fuel pin cladding and led to an early incorrect assumption that the

creep-swelling coupling coefficient depended on alloy starting condition, irradiation temperature, and other variables. An explanation of the origin of the distortion along with a reanalysis of both French and U.S. creep data are presented elsewhere.(38-43) The major conclusions of this reanalysis are that the coupling coefficient is independent of all relevant variables, being $\sim 0.6 \times 10^{-2} \text{ MPa}^{-1}$, and that the apparent distortion of the coupling coefficient reflects (a) the dependence of swelling on these variables and (b) the manner in which the average swelling rate was determined in the French LMR program.

A microstructurally-based explanation of (a) the role of swelling in first accelerating irradiation creep and then causing the creep rate to decrease, and (b) for creep later to completely "disappear" has been recently provided by Woo, Garner, and Holt.(42,43) The dislocation and loop microstructures that evolve in the absence of void swelling are known to become progressively more anisotropic in the distribution of Burger's vectors as the stress increases.(21,22) This anisotropy is important to the operation of the stress-induced preferential absorption (SIPA) mechanism of irradiation creep. When voids begin to form, the interstitials matching the vacancies accumulating in voids feed this previously developed anisotropic microstructure at a rate significantly larger than that associated with a void-free dislocation-dominated microstructure. This accounts for the initial acceleration of irradiation creep when swelling begins. As swelling increases, however, one consequence is the progressive development of a more isotropic dislocation microstructure, thus overwhelming the anisotropy that underlies the SIPA creep process.

Thus creep "disappears" as swelling progresses to higher levels and the total diametral strain rate approaches that due to swelling alone. Since the

swelling of typical FCC steels never exceeds 1%/dpa (volumetric), this limits the diametral strain rate to 0.33%/dpa in the feedback-dominated regime. An interesting prediction of Woo et al.'s model, however, is that the 0.33%/dpa strain rate limitation would be reached long before the swelling rate reached 1%/dpa. Thus, the 0.33%/dpa limitation applies to the total of the creep and swelling strain rates. In this case, the creep rate falls even after the strain rate limitation is reached in order to maintain the strain rate limitation as the swelling rate increases. This prediction was later proved to be valid by analyzing the strains of tubes that had reached the 0.33%/dpa limit, but then lost pressure, continuing to swell thereafter (44).

4. Feedback of Swelling into Mechanical Properties

It is well known that irradiation of annealed FCC metals usually leads to an increase in strength and a reduction in ductility, as demonstrated in Fig. 18. Another characteristic of irradiated metals is that the yield strength approaches the ultimate strength as demonstrated in Fig. 19, but saturation of strength usually occurs before significant swelling is accumulated.

When large levels of swelling occur without a significant incubation period, however, a loss of strength is sometimes observed instead of an increase, although the ductility is still reduced. An example of such behavior is shown in Fig. 20 for annealed pure copper irradiated in the Fast Flux Test Facility (FFTF), which produces swelling at ~0.5% per dpa. These copper specimens also displayed a unique microscopic failure mode in tension tests, that demonstrated the voids probably played a large role in the failure mechanism (46). The irradiated tensile specimens also failed with their

fracture surfaces normal to the stress axis rather than at the usual 45° orientation, even though a significant amount of ductility was retained at these large swelling levels.

In commercial FCC alloys, swelling occurs only after a relatively long incubation period. Although annealed materials usually reach the same saturation strength level as do cold-worked alloys (47) (see Fig. 21), they usually begin to swell much sooner than cold-worked alloys. Thus, it appears that voids are initially a relatively small contributor to the hardening process. As swelling increases, however, it is clear that voids exert an increasing influence on mechanical properties. The tendency toward loss of both strength and ductility with increasing swelling are demonstrated in Fig. 22 and 23.

Fissolo et al. showed that the ductility of titanium-modified 316 stainless steel fell continuously with increasing swelling and that rupture occurred without reduction of area at $\geq 6\%$ swelling (49). In the swelling range above $\sim 10\%$, the material exhibited no ductility at all. A similar correlation between swelling and brittle failure in Charpy impact tests was also observed by Fissolo, as shown in Fig. 24. Note that the saturation in total absorbed energy also occurs at $\sim 6\%$ swelling. In both tensile and Charpy tests, the fracture surfaces were transgranular with mixtures of ductile zones and planar brittle zones. The fracture surfaces in Fissolo's studies were observed to rotate from 45° to 90° from the tensile axis as the swelling level increased, however.

This tendency for failure to occur normal to the stress axis has been observed in other studies. Neustrov and Sharmadin demonstrated this quantitatively for an annealed Fe-18Cr-10Ni-Ti steel irradiated in BOR-60 at

400-500°C.(50) Note in Fig. 25 that the rotation and plasticity loss corresponded with convergence of the yield and ultimate strengths. They also demonstrated the effects of voids on softening and elasticity loss at higher swelling levels (Fig. 26). In this work, however, the critical swelling levels for total loss of ductility are larger than those of Fissolo et al. The reason for the difference probably arises from the different precipitate microstructures and void densities developed in each steel. A similar explanation was advanced by Neustrov, Golovanov and Sharmadin to explain a difference in critical swelling levels for plasticity loss in two other steels, as shown in Fig. 27. Note also that each steel in this study was irradiated over a different temperature range, which produces a different void density in the peak swelling regime.

Microstructural examination by Neustrov et al. showed that an abrupt increase occurred in the stress concentration factor whenever the mean size of the various microstructural components exceeded 40% of the mean distance between them. In heavily precipitated alloys, a greater concentration of potential stress risers was present, and a smaller amount of swelling was therefore necessary to reach the critical level for zero ductility.

In each of the studies mentioned above, another facet of the microstructural evolution with increasing swelling was a progressive increase in flow localization, as evidenced in the observation of sheared void zones, and "channel" facets on the fracture surface.

It would be convenient at this point to assert that stress concentration effects arising from precipitates and voids was the sole source of both hardening and softening, flow localization, the associated ductility losses, and the rotation of the fracture surface, but there are other second order

effects of voidage that probably contribute. Hamilton et al. demonstrated that the onset of failure surface rotation and severe embrittlement in AISI 316 was accompanied by a sudden large increase in strength as opposed to a decrease (52). They showed that the microstructural origin of this increase was the segregation of nickel to void surfaces, which lowered the M_s temperature of the volumes between voids. The stacking fault energy of Fe-Cr-Ni alloys is also a strong function of temperature at lower nickel levels, decreasing with decreasing temperature. These factors predispose the matrix of low nickel alloys to massive stress-induced formation of fine ϵ -martensite platelets. These platelets form at the initiation of tensile tests conducted at relatively low temperatures. Their formation causes great difficulty in handling irradiated 316 stainless components after they are removed from reactor, but not while they are kept at liquid sodium temperatures. In the case cited by Hamilton et al., the critical swelling level was ~10% for irradiations conducted at ~400°C. Of particular significance was the concurrent drop of the tearing modulus to zero in low temperature tests (52,53), which leads to a very brittle and unforgiving response to applied stresses.

Another potential second order effect of voids may be related to their preference to form on some types of precipitates and not others, thus presenting different stress concentration and crack initiation opportunities. Yet another source of second order effects is the impact of voidage on elastic constants, Poisson's ratio, and thermal conductivities, as reviewed in ref. 54. Of particular significance to the softening effect at larger swelling levels is the voidage-induced reductions in shear, Young's and bulk elastic moduli, measured in neutron irradiated stainless steels to be approximately -

2.5% for each 1.0 percent of swelling (55,56). Theoretical treatments support the magnitude of these reductions as well as their applicability to other materials such as copper (54). This influence of swelling on elastic properties may also partially explain the softening behavior of copper shown earlier in Fig. 20.

It thus appears that void swelling has both strong first order and strong second order effects on the mechanical properties of irradiated FCC alloys. The attainment of the nil ductility state can lead to failure of structural components at room temperature with only the slightest of physical insults, an excellent example of which is shown in Fig. 28. Similar examples of brittle failure that occurred during not very strenuous out-of-reactor handling of "highly" swollen components are provided in other papers (32,52).

5. Conclusions

It appears that void swelling can exert significant influence on the late-term evolution of dislocation and precipitate microstructure and also on post-irradiation deformation-induced microstructure. Operating via both first-order and second-order effects (such as segregation at void surfaces or dislocation annihilation), the swelling phenomenon can exert feedback on the swelling process itself and also on in-reactor creep processes. Post-irradiation mechanical properties are also strongly influenced by large levels of voidage, although the critical swelling level depends on the irradiation and test temperatures and the presence of other microstructural components, especially precipitates.

Acknowledgements

This work was sponsored by the U.S. Department of Energy under two contracts, DE-AC06-76RLO 1830 with Pacific Northwest Laboratory (operated by Battelle Memorial Institute for the U.S. Department of Energy) and DE-FG06-89ER-75522 with the Northwest College and University Association for Science. The assistance of Jill Walters in preparing the manuscript is gratefully acknowledged.

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Figure Captions

1. Swelling observed in various Fe-15Cr-Ni alloys during irradiation in EBR-II.(4). Swelling at temperatures $\geq 538^{\circ}\text{C}$ (top) and 427°C (bottom). Saturation may be developing at 427°C for relatively large swelling levels.
2. Swelling observed in 20% cold worked AISI 316 during irradiation in EBR-II.(5)
3. Saturation of swelling and microstructure in annealed AISI 316 during 4 MeV Ni^{+} ion irradiation at 627°C and 6×10^{-3} dpa/sec with simultaneous injection of He and D atoms at rates of 10 appm/dpa and 25 appm/dpa, respectively.(7)
4. Saturation of swelling observed in pure nickel irradiated in EBR-II. The saturation level is dependent on irradiation temperature and starting microstructure.(9)

5. Swelling behavior observed in pure molybdenum during irradiation in FFTF.(11)
6. Irradiation creep of annealed aluminum at 393K (left) and annealed copper at 523K (right) observed in the WWR-K reactor by Ibragimov et al.(19)
7. Irradiation creep of a single heat of 20% cold worked 316 stainless steel during irradiation in two experiments with different stress states in EBR-II.(20)
8. Swelling and irradiation creep observed in pure nickel pressurized tubes irradiated in the SILOE reactor.(23)
9. Saturation of swelling and creep in 20% cold worked DIN 1.4864 (Fe-15Cr-35Ni-1.16Mn-1.56Si-0.05C) pressurized tubes at 420°C in PFR.(25)
10. (a) Diametral strains resulting from void swelling at ~400°C in neutron irradiated stress-free tubes constructed from nine titanium-modified 316 stainless steels. (b) Stress-normalized midwall creep strains observed in three of these heats, showing a strong correlation of swelling and irradiation creep rates.(28)
11. Instantaneous creep coefficient derived from strain measurements on pressurized tubes constructed from a double-aged 316 stainless steel (heat treat D) irradiated in EBR-II at 823K.(32)
12. Creep and swelling strains observed in annealed 347 stainless clad fuel pins irradiated in EBR-II.(33) These data were originally explained in terms of fuel-clad interaction as the major source of stress in the cladding, with fuel contact and creep eventually terminated by the onset of clad swelling.

13. (a) Deformation observed in pressurized tubes of 20% cold worked AISI 316 irradiated in EBR-II at 550°C.(32,33) Note that doubling the hoop stress level from 103 MPa to 206 MPa does not double the creep rate, which never exceeds 0.33%/dpa. (b) Density measurements on the 30 Ksi (206 MPa) tube show that stress accelerates the rate of swelling, but also causes the creep rate to approach zero at high swelling levels.
14. Comparison of predicted and measured creep strains for cold worked 316Ti stainless steel ducts irradiated in PHENIX.(35)
15. Comparison of predicted and measured creep strains for cold worked 316Ti stainless steel cladding irradiated in PHENIX.(35)
16. Comparison of predicted and measured diametral strains of 20% cold worked 316 stainless steel cladding irradiated in FFTF. Note that the measured strains appear to arise primarily from swelling, with the creep strains much lower than predicted from pressurized tube studies.(36)
17. Comparison of creep and swelling behavior observed in two 20% cold worked 316 fuel pins irradiated in RAPSODIE. Note the relative distortion with increasing temperature of the magnitude of the creep profiles with respect to the respective profiles of swelling.(37)
18. Changes in room temperature tensile properties of Fe-15Cr-25Ni and Fe-15Cr-45Ni after irradiation in EBR-II to 12.5 dpa at 450°C. Previously unpublished tensile test curves from experiments described in ref. 45.
19. Saturation of irradiation-induced hardening of annealed AISI 304L irradiated at 370°C in EBR-II (46).
20. Changes in room temperature tensile behavior of annealed pure copper after irradiation in FFTF, showing an apparent reduction in strength with irradiation (47).

21. Evolution of yield strength in AISI 316 stainless in EBR-II at temperatures within the void swelling regime (48).
22. Onset of strength loss with swelling above ~5% in 20% cold-worked titanium-modified stainless steel hexagonal wrappers after irradiation in PHENIX (49).
23. Correlation between ductility loss and swelling measured in several heats of 20% cold-worked titanium-modified stainless steel irradiated in PHENIX to a variety of dpa levels over a range of irradiation temperatures (49).
24. Correlation of swelling and embrittlement induced in titanium-modified 316 Charpy specimens irradiated in PHENIX (49).
25. Influence of swelling on fracture properties of an annealed Fe-18Cr-10Ni-Ti steel irradiated in BOR-60 at 400-500°C (50).
26. Influence of swelling on tensile properties of the annealed Fe-18Cr-10Ni-Ti steel shown in Fig. 25, showing softening that occurs with increasing swelling (50).
27. Influence of swelling on ultimate tensile strength of OKh16N15M3B cladding and Kh18N10T hexagonal ducts irradiated in BOR-60 (51). Tests were performed at the irradiation temperature, using specimens cut from regions of maximum swelling.
28. Out-of-reactor failure during routine handling of a titanium-modified 316 clad fuel pin after irradiation in FFTF to a peak fluence of $\sim 2.4 \times 10^{23} \text{ n cm}^{-2}$ ($E > 0.1 \text{ MeV}$). The break occurred at ~32% swelling. (Courtesy of B.J. Makenas, Westinghouse Hanford Company.)

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