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A PERSPECTIVE ON PRESENT AND FUTURE ALLOY DEVELOPMENT EFFORTS ON AUSTENITIC STAINLESS STEELS FOR FUSION APPLICATION*

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ABSTRACT

High fluence neutron data on advanced titanium-modified alloys are only now becoming available from more recently instituted (1975–78) systematic alloy development programs. Experiments on helium effects in these alloys are currently in progress. Studies thus far indicate that helium enhanced void swelling and grain boundary embrittlement potentially impose the most severe temperature and lifetime limits for fusion first walls. However, resistance to both swelling and embrittlement can be improved by controlled production of fine, stable dispersions of helium bubbles attached to MC precipitate particles in the matrix and grain boundaries of the advanced titanium-modified alloys. Fine, stable bubble microstructures also cause dilution of radiation-induced solute segregation (RIS); RIS plays a major role in the phenomenon of void swelling and possibly affects embrittlement and/or corrosion under irradiation as well. Lower fluence data from the High Flux Isotope Reactor (HFIR) and the Oak Ridge Research Reactor (ORR) suggest that helium does not degrade the properties of fatigue and creep under irradiation, but further study is required. An additional challenge is to impart similar radiation resistance characteristics to new alloys being developed for lower induced radioactivity; these are alloys in which manganese replaces nickel while the concentrations of molybdenum, copper, and nitrogen are severely restricted.

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1.0 INTRODUCTION

Because we have been studying radiation damage to austenitic stainless steels for over twenty years, one may feel that enough has been learned in order to optimize and use these alloys, and that further study is fruitless. However, a historical perspective indicates that the first ten or more years of study (1965–1975) were spent discovering the basic effects of fast neutron irradiation, which are so familiar to us today. For example, radiation-enhanced thermal precipitation,[1] radiation-induced void and bubble formation,[2–4] radiation-induced creep,[5] radiation-induced precipitation,[6,7] and radiation-induced solute segregation[8] were all discovered or first studied in austenitic stainless steels. Furthermore, systematic alloy development programs in the United States did not begin for fast breeder reactor (FBR) materials until 1974, and for magnetic fusion reactor (MFR) materials until 1977. Selections of compositions for the FBR D9 type alloys and MFR prime candidate alloy (PCA) were not made until 1977–78. Therefore, high fluence FBR data on the D9 type alloys are only currently becoming available. Thus, despite a fairly long, general study of the effects of irradiation on austenitic alloys, we have only recently been characterizing and understanding irradiation resistance in these advanced titanium-modified alloys, particularly for MFR conditions. Furthermore, because interest in void swelling has dominated the irradiation effects efforts, there has been far less study of degradation of other materials properties under irradiation at conditions relevant to fusion.

With this background, the purpose of this paper is to address important questions concerning how to effect further alloy development of austenitic stainless steels for swelling resistance, and to what extent we can or must influence the behavior of other properties under irradiation, such as (a) strength/embrittlement, (b) fatigue/irradiation creep, (c) corrosion (under irradiation), and (d) radiation-induced activation. To summarize our current understanding, helium has been found to have major effects on swelling and embrittlement, but several metallurgical avenues are available for significant improvement relative to type 316 stainless steel. Studies on fatigue and irradiation creep, particularly including helium effects, are preliminary but have yet to reveal engineering problems requiring additional alloy development remedies. The effects of irradiation on corrosion behavior are unknown, but higher alloy nickel contents make thermal corrosion in lithium worse. The newest challenge will be preserving the irradiation resistance already achieved in the advanced austenitic alloys while making major compositional changes (replacing nickel with manganese and removing molybdenum) in order to lower latent radioactivity of the MFR structure after shutdown. The following sections elaborate and expand upon these points.

2.0 CAVITY SWELLING

Swelling under fast neutron irradiation, with concurrent helium produced by transmutations, is caused by formation of either voids (cavities with bias-driven growth) or bubbles (cavities with helium-driven growth); below 650°C, voids undergo unstable growth and produce far more swelling at a given helium level than do stable bubbles.[9,10] Void swelling develops slowly with fluence during an initial transient period, during which voids, RIS, and precipitation compatible with RIS develop concurrently.[10–12] After the transient, swelling rapidly accelerates into a regime with a nearly linear rate, often approaching ~1%/dpa (ref. 13), characterized by void growth and coalescence. This behavior can be seen for FBR-irradiated CW 316 in Fig. 1. Our theoretical understanding is that voids form from unstable bubbles which exceed either their critical size or critical gas contents during the transient regime.[14-16] Therefore, the objective of alloy development efforts is to extend the transient regime by delaying void formation. Cold working (CW) together with adjustments of Ni and Cr and additions of Ti, Nb, or P relative to normal type 316 are effective metallurgical avenues for extending the transient regime under FBR irradiation.[11,17,18] Advanced modified austenitic alloy compositions are being investigated by several countries for MFR and/or FBR application (Table 1). Figure 1 shows the approximately fourfold increase in transient period achieved through MC formation together with delayed void and RIS development in CW D9 (similar to PCA) relative to type 316 in EBR-II.[11] An important question for fusion is how increased helium generation affects the void swelling behavior normally observed under FBR irradiation.

In lieu of a high fluence source of fusion neutrons, mixed spectrum reactor irradiations (like HFIR or ORR) or helium preinjection have been used to survey the effects of helium relative to a base line of FBR irradiation. Two basic and opposite effects of helium have been found. Increased helium generation (~10^2 times more in HFIR) has been found to shorten the transient period of some alloys while lengthening it for others.[9-11,19] In general, increased helium generation always increases early bubble nucleation compared to the FBR case. However, if nucleation rates are not increased above a critical threshold so that these bubbles are not the dominant sinks, the bubbles either coalesce or quickly convert to voids while RIS and compatible precipitate development are enhanced,[10,11,14,15] thus shortening the transient regime. This effect has been observed in most SA alloys (including PCA types) and some CW 316s. Conversely, very high bubble nucleation rates can produce bubble-sink-dominated systems in which void formation is suppressed, RIS is diluted, and thermal precipitation is enhanced, thus lengthening the transient regime.[9-12,14,15] This behavior has been observed primarily in some CW 316s and most CW titanium-modified alloys. Because much higher fluence irradiations are required to test the effects of helium-extended transients in swelling-resistant alloys, the duration of the stable bubble structures and the degree of extension are open questions at this point. Furthermore, will these extensions be manifest at helium generation rates that are somewhat lower in MFRs than found under HFIR irradiation of these steels? Several factors appear to be involved in the swelling suppression of the 20 to 25% CW PCA/D9 type alloys with increased helium. The precipitation of a dispersion of MC also enhances bubble nucleation per increment of generated helium. The association of fine MC particles and helium bubbles further hinders bubble coarsening by coalescence. The resulting high density of bubble/precipitate sinks also suppresses RIS and thereby enhances MC stability. Currently, these ideas and mechanisms about helium-extended transients are being tested by unique experiments in which normal FFTF-irradiated samples are being compared to samples of the same steels preirradiated to various degrees of microstructural evolution in HFIR prior to FFTF irradiation.[20]

Several questions still remain that make further alloy development and experimental investigations desirable: (a) Can swelling transients in a commercial MFR confidently be extended to and beyond 150 to 200 dpa at 500 to 600°C? (b) Can helium-enhanced void formation at lower temperature (300–450°C) be suppressed? (c) How does stress influence helium effects?

and (d) Can swelling resistance be achieved in SA material (easier to weld)? Regarding the first question (a), small melts of PCA further modified with varying combinations of Ti, V, Nb, P, B, and C (intended to further stabilize the MC phase and to minimize undesirable phases) are currently being irradiated in various experiments.[20,21] Recent work on P-modified steels demonstrated efficient helium trapping along lath-shaped phosphide particles;[18] combined MC/phosphide precipitation may prove particularly effective. Regarding question (b), Fig. 2 illustrates the relative shifts in peak void swelling temperatures for various steels irradiated in HFIR; the behavior is not completely understood, but may involve differences in loop formation and radiation-induced precipitation.[10] Swelling is unlikely to be a problem below 300°C. A legitimate goal of alloy development may be to simply ensure that the swelling peak remains above 550 to 600°C, where embrittlement and corrosion (in lithium) also limit first wall lifetimes. The last questions [(c) and (d)] demand more experimental work.

3.0 STRENGTH AND EMBRITTLEMENT

Generally, stress-free fast neutron irradiation reduces ductility relative to unirradiated material through irradiation hardening at lower temperatures, and through grain boundary embrittlement (induced by segregation, precipitation, or helium) at higher temperatures. Severe embrittlement causes grain boundary separation with little or no plastic flow and concurrent loss of strength. For fusion, ductility must be sufficient to prevent fracture of the first wall, which would cause a breach of vacuum integrity and a loss of coolant into the plasma chamber. However, strength must also be adequate to prevent plastic deformation, particularly as swelling gradients (due to temperature dependence) build up stress gradients with time.[23] Helium embrittlement is a major concern for fusion at the higher gas generation rates relative to FBR behavior; helium embrittlement primarily imposes an upper temperature limit, but can also limit lifetimes when materials are swelling resistant at 500 to 600°C.

Most austenitic stainless steels become quite strong, but show no evidence of helium embrittlement (i.e., ductile, transgranular failure) at temperatures of 400°C and below,[22,24–28] even after the accumulation of ~3000 to 4000 at. ppm He. Figure 3 shows tensile yield and total elongation trend curves as functions of irradiation temperature for a variety of austenitic stainless steels after HFIR irradiation to 10 to 22 dpa and 1400 to 1750 at. ppm He;[22,24–28] EBR-II data for CW 316 irradiated to ~20 dpa and 5 to 15 at. ppm He are also included to gage the effects of helium.[29] Above 400°C, SA materials harden progressively less with increasing irradiation temperature, but CW materials recover and soften. In general, strength and ductility changes saturate with fluence at temperatures below 500 to 550°C. Above 550°C, however, the loss of both ductility and strength combined with grain boundary bubble coarsening and the onset of intergranular failure all suggest helium embrittlement in HFIR. The embrittlement becomes worse with fluence in many heats. Disk bend data are also consistent with these trends, but reveal severe embrittlement at 500 to 600°C for some alloys.[28,30] Above ~650°C, ductility and strength rapidly fall during HFIR irradiation.[31] Helium has little effect on strength at 600°C and below, whereas in HFIR, helium enhances embrittlement above 550°C relative to EBR-II irradiation. However, below ~650°C, heat-to-heat and/or pretreatment variations have larger effects on strength and ductility than does helium (strong materials, like PCA, have less ductility). Furthermore, these property changes correlate reasonably with microstructural changes,[28,32,33] to reveal underlying mechanisms and alloy development avenues for improvement. Therefore, several goals for embrittlement resistance for fusion are: (a) to reproducibly achieve the properties of the better heats of steel, (b) to increase the embrittlement cutoff to ~600°C or above, and (c) to extend the duration of resistance at 500 to 600°C.

Fundamental studies of helium embrittlement reveal that helium embrittlement is primarily caused by grain boundary bubbles exceeding their critical radii (either by applied stress or by coalescence or growth).[34-36] Titanium-modified austenitic stainless steels exhibit better embrittlement resistance than unmodified steels due to the direct benefits of bubble refinement by MC-helium trapping.[27,28,30,37-40] The ductility maxima at 500 to 550°C for HFIR-irradiated type 316s and 316 + Ti correlate with formation and stability of $M_{23}C_6/M_6C$ at the grain boundaries; at higher fluences the carbides dissolve, and at higher temperatures they do not form or are replaced with intermetallics (Laves and σ phases).[32,33,41] Figure 4 shows the particular embrittlement resistance imparted to the strong CW PCA by bubble refinement induced by stable MC particles formed at the grain boundaries via thermal treatment prior to HFIR irradiation.[27] The current United States strategy for improved embrittlement resistance involves compositional variations of the PCA to further promote consistent and controllable MC formation at the grain boundaries and to increase its stability under irradiation, possibly in conjunction with $M_{23}C_6/M_6C$ as well. However, questions still remain about welding such pretreated alloys and about the effects of stress and creep deformation on microstructural stability under irradiation.

4.0 FATIGUE/IRRADIATION CREEP

Fatigue failure and cracking due to thermal stresses are obvious concerns for MFRs with cyclic plasma operation and may still even be a factor, considering plasma disruptions, in steady-state tokamaks. Data on the effects of irradiation on fatigue life are primarily on FBR-irradiated types 304 and 316 stainless steel.[42–44] The most relevant helium effects data are that by Grossbeck and Liu[43,45] for one heat of CW 316 irradiated in HFIR at 430 and 550°C to ~900 at. ppm He and 15 dpa; fatigue was reduced at 430°C, possibly due to irradiation hardening, but was unaffected by irradiation at 550°C (even for tests at 650°C) relative to unirradiated material. It should be noted, however, that most HFIR-irradiated steels are not severely embrittled after such irradiation (see Fig. 3) and that this particular heat of type 316 (MFE reference heat X-15893) shows less embrittlement than other CW 316s (DO-heat, N-lot) at higher fluence.[22,28] Microstructurally, the CW 316 (X-15893) also has stable grain boundary carbides ($M_6C/M_{23}C_6$) after HFIR irradiation to ~15 dpa at both temperatures. Similarly, Serpan et al.[46] see no embrittlement after helium preinjection (10–100 at. ppm) and actually observe improved fatigue life in the CW 316 tested at 816°C. Ermi and Chin[47] have also conducted low-fluence, in situ fatigue tests for CW 316 in ORR and found no degradation at 460°C. However, the introduction of tensile hold times can reduce the fatigue life of irradiated material.[48] Batra et al.[49] also note that severe embrittlement appears in helium preinjected type 316 (800 at. ppm) at 600°C when the fatigue cycle frequency is reduced from 5 to 0.5 Hz.

Fatigue studies thus far do not indicate first wall temperature or lifetime limits more severe than those imposed by swelling or normal helium embrittlement. Therefore, it appears that fatigue does not pose problems requiring new alloy development solutions. More study, however, is needed for the effects of higher fluence irradiations, particularly with higher helium contents and possibly lower temperatures, and for the effects of lower frequencies and/or longer hold times (including possibly in situ studies). Finally, care must be taken to prevent alloy development efforts aimed at other properties from impairing fatigue behavior, particularly since there are virtually no data on titanium-modified steels.

Irradiation can enhance thermal creep, or induce creep at lower temperatures and stresses due to the ongoing point defect production. Irradiation creep is expected to relieve swelling generated stresses in the MFR first wall.[23] There is an abundance of irradiation creep data, but most were generated via FBR irradiations prior to 1977.[50,51] There are very little data for the effects of helium on irradiation creep. Two aspects of creep under irradiation are important — the actual deformation behavior and the time to eventual rupture. Under FBR irradiation, creep parallels the fluence evolution of void swelling, increases linearly with stress, and is weakly temperature dependent, but increases at lower temperature.[51] Irradiation, however, can also decrease rupture life and strain, but not as severely as suggested by early postirradiation creep testing.[52–54] However, in-reactor testing still shows the same intergranular failure.[52] Helium may alter irradiation creep via its effects on matrix microstructural evolution,[55] and may further reduce rupture life through enhanced grain boundary embrittlement. Reactor data currently are insufficient to address the latter question, but can address the first.

Irradiation creep rates appear to be lower for CW 316 and CW PCA irradiated in the ORR at 500°C to ~5 dpa relative to CW 316 similarly irradiated in EBR-II, as shown in Fig. 5 (ref. 44), but no difference is observed between reactors at 330°C. Helium may reduce irradiation creep via microstructural refinement.[55] Hishinuma et al.[57] clearly show that pre-injected helium (20–60 at. ppm) suppresses both void swelling and irradiation creep for CW 316 irradiated in EBR-II at 525°C to 23 dpa. Microstructurally, the suppression correlated with a dense population of fine bubbles rather than coarse voids. Because irradiation creep depends on the biased flow of interstitials to dislocations, irradiation creep suppression is consistent with similar stress-free suppressions of void formation and RIS by bubble sink-dominated structures.[9–12] However, this is simply a consistent effect of helium on both swelling and creep rather than a change in their relationship. Irradiation creep may equally well be enhanced when helium enhances void swelling (and RIS) at lower temperatures, as shown in Fig. 2.

More data on the effects of helium on irradiation creep are needed, both at lower temperatures and at higher fluences; creep embrittlement may be more severe than tensile embrittlement, further reducing the maximum first wall temperature. Another question is the relationship between swelling and creep when swelling is due to bubbles driven by helium rather than bias-driven voids. Irradiation creep, however, does not currently impose new problems requiring alloy development solutions, although it appears to benefit from alloy development efforts already directed toward swelling and embrittlement resistance.

5.0 CORROSION UNDER IRRADIATION

Corrosion by the coolant thins the first wall and can aggravate mechanical failure. Corrosion is a property that limits the maximum first wall temperature at the coolant interface. It is also a property that is generally not the focus of alloy development to the same degree as swelling and embrittlement resistance. One reason may be that coolants change with MFR design (and particular corrosion problems differ with each coolant), unlike the fusion neutron damage spectrum which remains the same. For example, the problem facing stainless steel in water at 200 to 350°C is stress corrosion cracking, exacerbated generally by the formation of chromium-rich carbides at grain boundaries in heat-affected zones of welds. But this problem may be alleviated by the same alloy development efforts which seek to encourage MC rather than $M_{23}C_6$ at the grain boundaries for helium embrittlement resistance. However, the problem becomes quite different for a stainless steel first wall and a liquid lithium coolant.

In lithium, thermal corrosion increases with increasing nickel content of the alloy, with type 316 corroding less than PCA type alloys,[58] as shown in Fig. 6(a). The initially rapid corrosion rate is reduced by the formation of a nickel-depleted ferrite layer, as seen in Fig. 6(b), which then corrodes more slowly than the original austenite. By itself, this behavior would suggest selection of lower nickel austenitics, whereas alloys like the PCA have higher alloy nickel contents to improve resistances to swelling and excessive intermetallic phase (Laves, σ) formation. However, corrosion behavior under irradiation may be different if RIS causes alloy modification of the surface at rates more rapid than the corrosion kinetics. Among its various effects, RIS causes the nickel content at surfaces and grain boundaries to be considerably enriched. Rapid nickel transport to the surface could frustrate formation of the ferrite layer, or worse, cause a higher local nickel content to enhance corrosion. The effects of RIS on the surface (depth of composition gradients and rate of development) are not well known for the metal-liquid metal interfaces of reactor-irradiated samples. Use of the advanced PCA/D9 type alloys, already designed for swelling resistance, may eliminate some uncertainty because a key to helium-suppression of void swelling in them involves helium bubble dilution of RIS.[11] When RIS is suppressed, normal thermal diffusion appears enhanced,[10] which may at least then lead to the more predictable thermal behavior observed in Fig. 6, and possibly provide another reason for employing the advanced irradiation-resistant austenitic stainless steels.

Experiments on the effects of irradiation on corrosion may be of interest; they would, however, be necessary to prove a problem exists before any additional alloy development effort would be required. Adding inhibitors, like aluminum, to the lithium may also help reduce corrosion.[59] However, corrosion in lithium offers a reason for not seeking swelling resistance by dramatically increasing the alloy nickel content.

6.0 REDUCED ACTIVATION AUSTENITICS

This new area of alloy design is a response to the need to reduce the induced radioactive decay of first wall and blanket components, so that after service they can be buried as Class C waste.[60,61] For austenitic stainless steels, the compositional guidelines are still evolving, but do entail replacing nickel completely with manganese, eliminating niobium, and severely restricting Mo, Cu, and N. Some commercial manganese-stabilized steels do exist, like the AISI 200-series austenitic stainless steels, but they are not generally used at elevated temperature and do not fit compositionally into the low activation guidelines. The challenge, therefore, is to develop a new class of Fe-Mn-Cr steels while maintaining the radiation resistance already achieved in the advanced modified Fe-Ni-Cr austenitic stainless steels.

There are currently several efforts under way to study or develop Cr-Mn austenitic steels for fusion. Studies at the Joint Research Center (JRC)-Ispra are focused on a broad characterization of many properties for several commercial steels in the range of 17 to 20 wt % Mn and 10 to 14% Cr (ref. 62). They find many properties comparable to equivalent nickel-stabilized alloys, except for increased strength and work hardening, deformation-induced martensite and lower thermal conductivity in the manganese-stabilized alloys. Under electron irradiation they find that helium enhances void swelling. Alloy development and studies at the Hanford Engineering Development Laboratory (HEDL) center on achieving a low swelling base composition in the Fe-Mn-Cr system equivalent to the anomalous swelling resistance demonstrated under FBR irradiation in the INVAR region of the Fe-Ni-Cr system (7–15 wt % Cr, 35–60% Ni) (ref. 63). These studies include a range of pure Fe-Mn-Cr alloys with 15 to 35% Mn and 0 to 15% Cr and steels with various solute additions in the range of 15 to 30% Mn and 2 to 15% Cr; (refs. 63,64); there is also work on type 216 stainless steel.[65] In EBR-II or FFTF they find swelling behavior similar to that found in nickel-stabilized austenitics, but with little chromium dependence and less reduction in swelling for increased manganese compared to nickel. Finally, alloys are being studied and developed at the Oak Ridge National Laboratory (ORNL) with the intent to produce a manganese-stabilized equivalent of the PCA, with a microstructural basis for irradiation resistance.[61,66,67] These alloys have 10 to 20 wt % Cr, 15 to 20% Mn, and 0.1 to 0.4% C. These studies, including thermal aging, show manganese to be less than half as effective an austenitizing agent as nickel. Precipitate phases are similar to those found in nickel-stabilized austenitics ($M_{23}C_6$, σ, Laves), but the manganese-stabilized steels seem more prone to intermetallic phases than the nickel-stabilized steels, despite higher carbon contents.

To summarize, the manganese-stabilized steels are different from the nickel-stabilized steels, but not so different as to preclude application of the same general alloy development principles used to impart irradiation resistance to the latter. Work is ongoing, but much more characterization is necessary before specific alloy development directions become clear.

7.0 SUMMARY AND PERSPECTIVE

Helium-enhanced void swelling and induced embrittlement limit fusion first wall operating temperatures and lifetimes and are the logical first priorities for alloy development efforts aimed at improved performance. Further alloy development for swelling resistance is worthwhile to confidently eliminate helium-enhanced void swelling at lower temperatures and to prolong the transient regime at higher temperatures. This effect is amplified by the degree to which helium helps to further prolong the transient in CW PCA/D9 type alloys. Swelling resistance also needs to be demonstrated during irradiation with applied stresses and would be desirable in SA materials from the standpoint of welding. Efforts to achieve helium embrittlement resistance are particularly worthwhile in void swelling resistant alloys, and grain boundary precipitate (MC) tailoring appears to be the most effective alloy development avenue. The aim should be to increase the temperature limits as much as possible beyond 550 to 600°C and to confidently increase the duration of embrittlement resistance at 500 to 600°C, particularly with stress conditions producing creep or fatigue. The effects of welding, however, still need to be studied on preirradiation microstructures that include cold worked and precipitate structures for irradiation resistance.

Other properties such as fatigue or creep under irradiation require further investigation, particularly for the effects of helium. Currently they do not appear to pose new problems requiring additional alloy development fixes, and indeed appear to benefit from the alloy development measures already outlined above. Corrosion concerns include the potential problems of stress corrosion cracking sensitization in water or RIS-enhanced corrosion in lithium under irradiation, both of which require more experimental study. But again, these corrosion problems may benefit from alloy development solutions already applied toward achieving swelling and helium embrittlement resistance.

The newest area of challenge is making major alloy compositional modifications to reduce radioactive decay problems while still preserving the degree of irradiation resistance already achieved in current PCA/D9 type steels. There appears to be enough general metallurgical similarity between the manganese- and nickel-stabilized steels to make these efforts worth pursuing.

The worth of alloy development efforts for austenitic stainless steels can also be affected by choices and tradeoffs that must be made by fusion reactor system design studies or by competition with other alloy classes, regardless of the improvements that the austenites demonstrate under irradiation. However, this author believes that it is important to continue to pursue the current strategy of developing multiple alloy classes, coupled with a broader, holistic view of many properties in each alloy class, to offer a choice among several good candidate materials for design optimization.

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LIST OF FIGURES

Fig. 1. Scatter bands of swelling versus fluence for various steels indicated after EBR-II irradiation. Data references are in ref. 11. Microstructural evolution that correlates with swelling behavior in indicated (RIS - radiation-induced solute segregation, RIP - radiation-induced precipitation).

Fig. 2. Approximate swelling versus temperature for steels irradiated in HFIR to illustrate relative heat-to-heat variation of peak swelling. Fluences are not the same for all steels. PCA and N-lot 316 were irradiated to ~44 dpa [P. J. Maziasz and D. N. Braski, J. Nucl. Mater. 122&123 (1984) 311–316]; MFE ref. 316 to ~55 dpa (ref. 22), and DO-heat 316 to ~47 to 60 dpa (ref. 9). CW 316 + Ti was irradiated only to 10 to 17 dpa [P. J. Maziasz and M. L. Grossbeck, DOE/ER-0045/6 (1981) 28–56], but was extrapolated to higher fluence.

Fig. 3. Schematic trend curves and scatter bands for (a) tensile yield stress and (b) total tensile elongation as functions of irradiation temperature for various steels irradiated in HFIR to 10 to 22 dpa and 500 to 1750 at. ppm He, and tested at or near the irradiation temperature (refs. 22, 24–28). A low helium (5–15 at. ppm) base line of EBR-II irradiated CW 316 is included (ref. 29). All HFIR specimens (except PCA at 300 to 600°C) were tested at 75 to 100°C below their actual irradiation temperatures. However, the properties were little affected by raising the test temperatures except at ~600°C and above, where ductility was reduced (ref. 22).

Fig. 4. Correlation of (a) total tensile elongation with (b)–(d) differences in grain boundary microstructure for PCA that had been either solution annealed (A1), 25%-cold-worked (A3), or aged (8 h at 800°C) and 25%-cold-worked (B3) prior to HFIR irradiation at 300 to 600°C to ~22 dpa and 1750 at. ppm (ref. 27). Stable grain boundary MC, produced via pretreatments, improves ductility by refining bubble size (ref. 28). (c) MC precipitate dark-field of (b).

Fig. 5. Effective creep strain as a function of effective stress for pressurized tubes of 25%-cold-worked PCA and 20%-cold-worked (MFE ref.-heat) 316 irradiated in the ORR to ~5 dpa at 500°C (ref. 44) and of 20%-cold-worked 316 similarly irradiated in EBR-II (ref. 56).

Fig. 6. (a) A plot of weight loss as a function of exposure time for various austenitic stainless steels and (b) metallography of the nickel-depleted ferrite region after 6700 h of PCA-A3 for corrosion testing in lithium at 600°C (ref. 58). PCA has more nickel than type 316 and corrodes more readily.

Table 1. Current Advanced Modified Austenitic Stainless Steel Candidates from Several National FBR and/or MFR Programs

COUNTRY	ALLOY	COMPOSITION, wt.%							
		Ni	Cr	Mo	C	Si	Ti	Nb	P
USA	PCA	16.2	14	2.3	0.05	0.4	0.25	–	0.01
JAPAN	JPCA	16	15	2.5	0.06	0.5	0.25	–	0.025
W. GERMANY	DIN 1.4970	15	15.3	1.3	0.1	0.3	0.3	–	–
UK	FV 548	11.5	16.4	1.4	0.1	0.4	–	1.0	0.01
FRANCE	316 + Ti	13	17.5	2.4	0.06	0.8	0.3	–	0.02

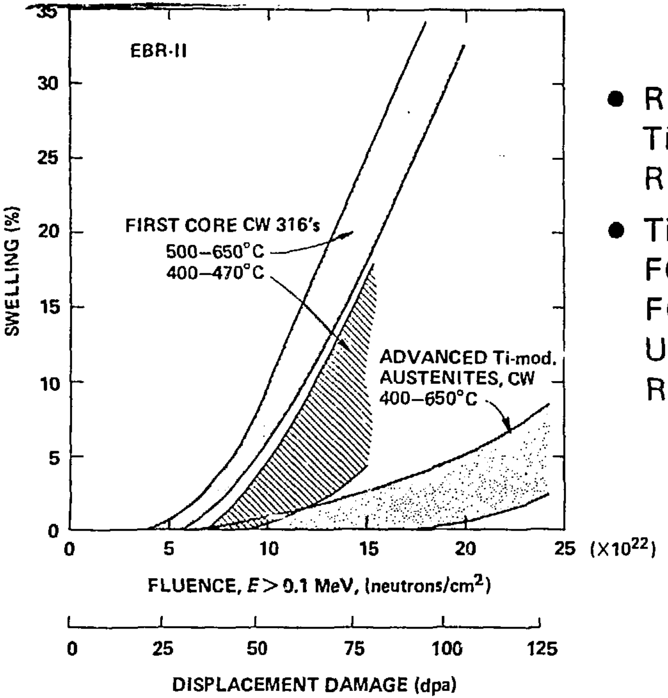


Fig. 1

- RIS, RIP AND VOIDS DEVELOP DURING THE LOW SWELLING TRANSIENT REGIME
- Ti ADDITIONS INITIALLY CAUSE MC FORMATION TO DELAY VOID FORMATION. AS MC BECOMES UNSTABLE, DUE TO RIS, VOIDS AND RIPs DEVELOP

RELATIVE SWELLING BEHAVIOR IN HFIR

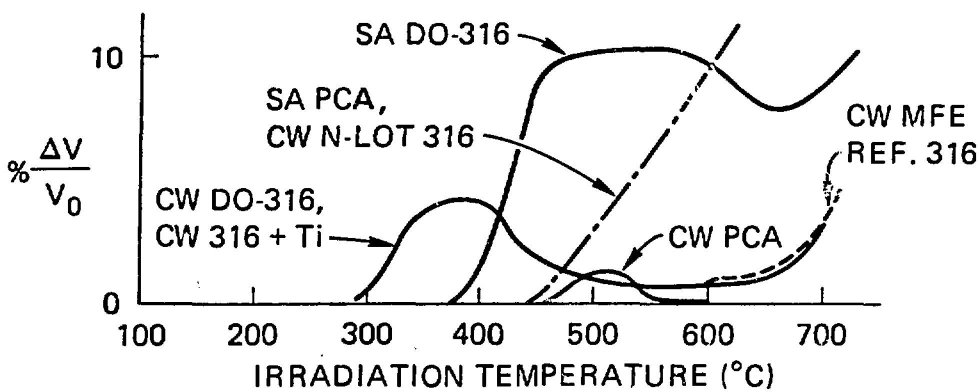
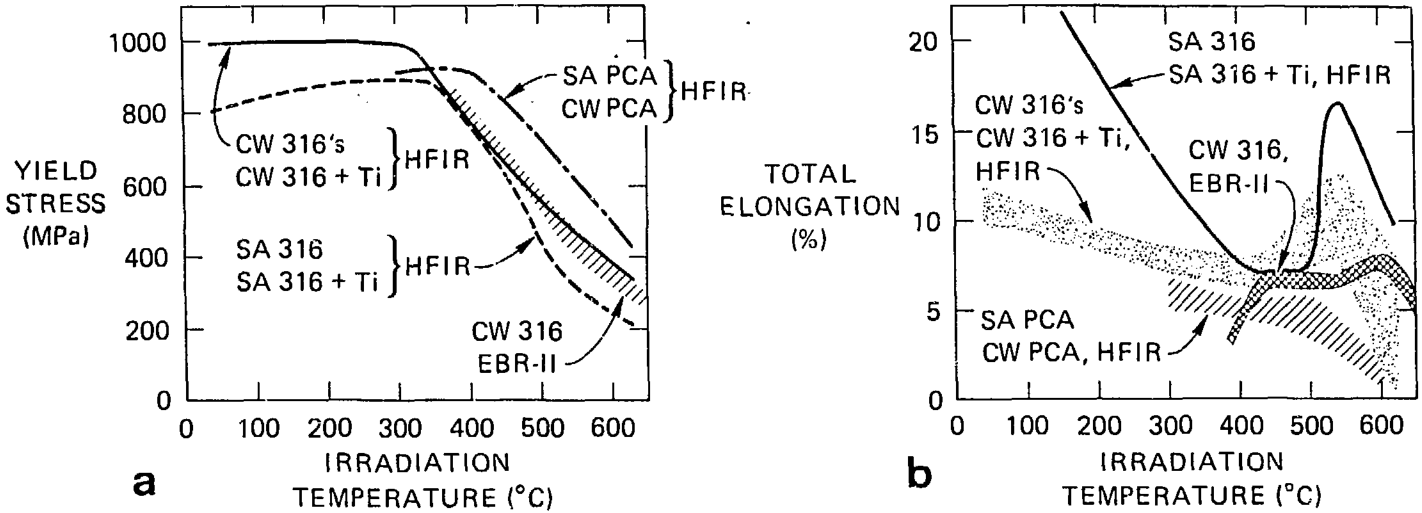


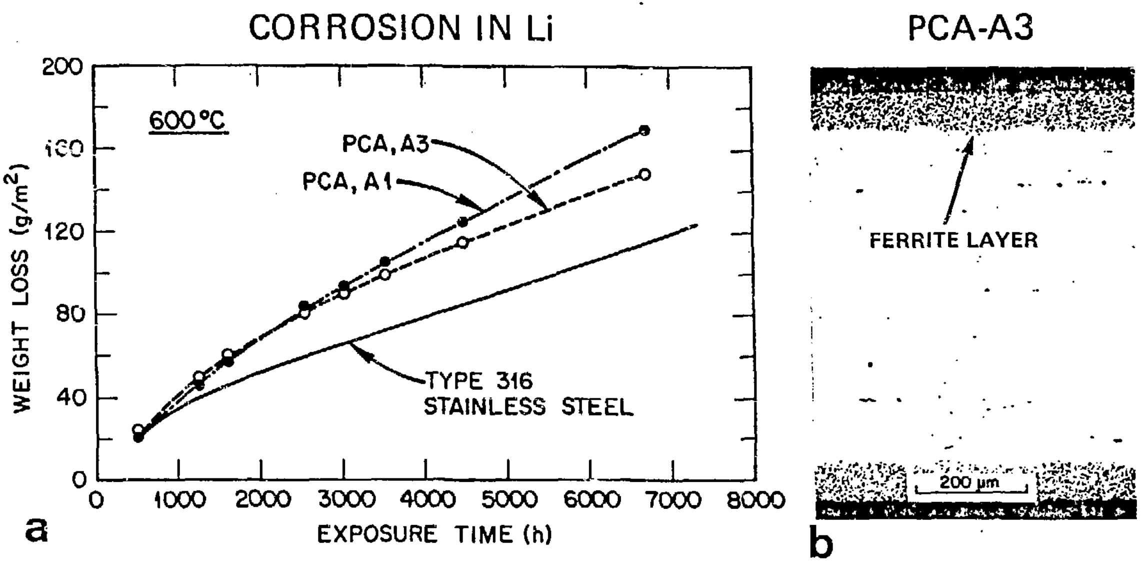
Fig. 2



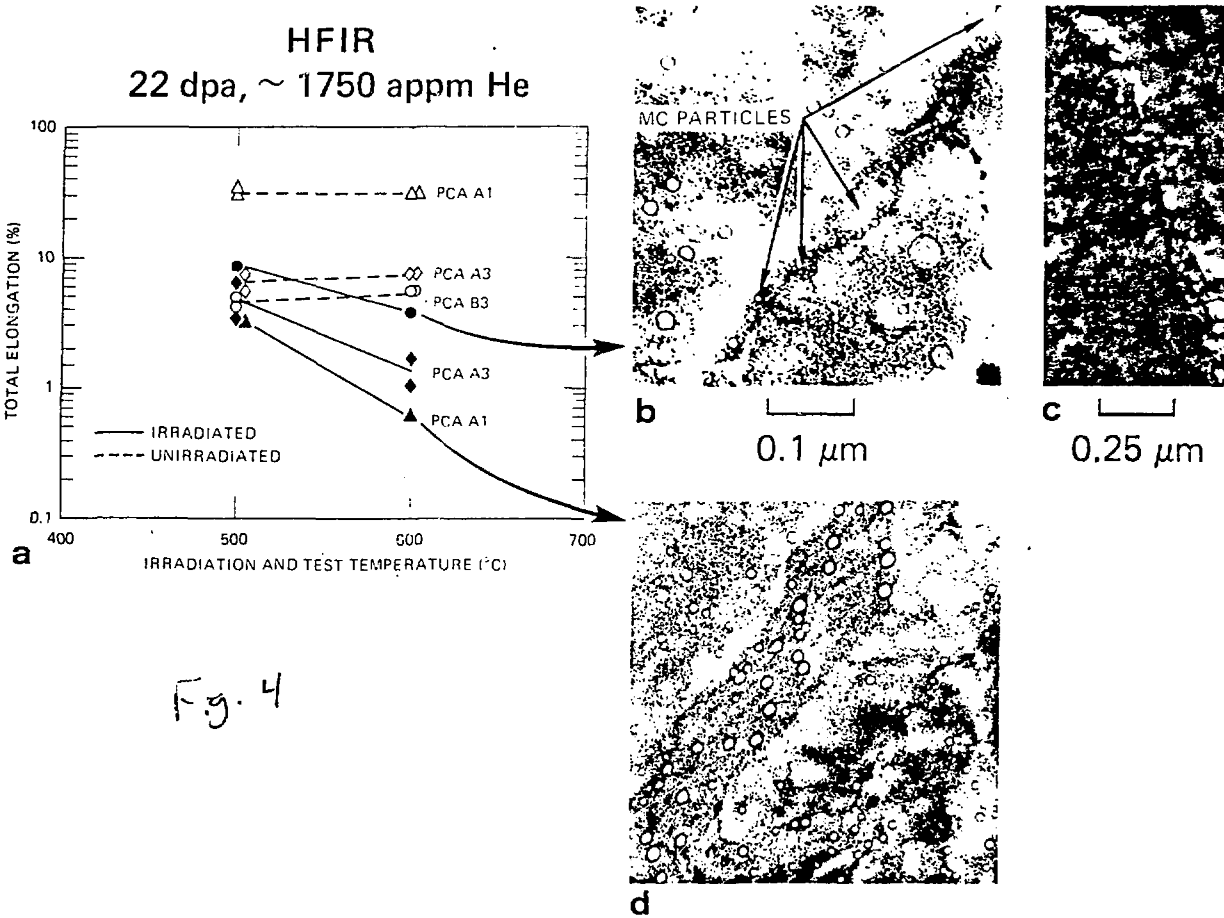
YIELD STRESS (MPa)
1000
800
600
400
200
0
0
100
200
300
400
500
600
IRRADIATION TEMPERATURE (°C)
CW 316's
CW 316 + Ti
HFIR
SA 316
SA 316 + Ti
HFIR
SA PCA
CW PCA
HFIR
CW 316
EBR-II
a
TOTAL ELONGATION (%)
20
15
10
5
0
0
100
200
300
400
500
600
IRRADIATION TEMPERATURE (°C)
SA 316
SA 316 + Ti, HFIR
CW 316's
CW 316 + Ti,
HFIR
CW 316,
EBR-II
SA PCA
CW PCA, HFIR
b

Fig. 3

Fig. 6



CORROSION IN Li
600 °C
PCA, A3
PCA, A1
TYPE 316 STAINLESS STEEL
WEIGHT LOSS (g/m²)
EXPOSURE TIME (h)
0
40
80
120
160
200
1000
2000
3000
4000
5000
6000
7000
8000
a
PCA-A3
FERRITE LAYER
200 µm
b



HFIR
22 dpa, ~ 1750 appm He
TOTAL ELONGATION (%)
100
10
1
0.1
PCA A1
PCA A3
PCA B3
PCA A3
PCA A1
IRRADIATED
UNIRRADIATED
400
500
600
700
a
IRRADIATION AND TEST TEMPERATURE (°C)
MC PARTICLES
b
0.1 μm
c
0.25 μm
d

Fig. 4

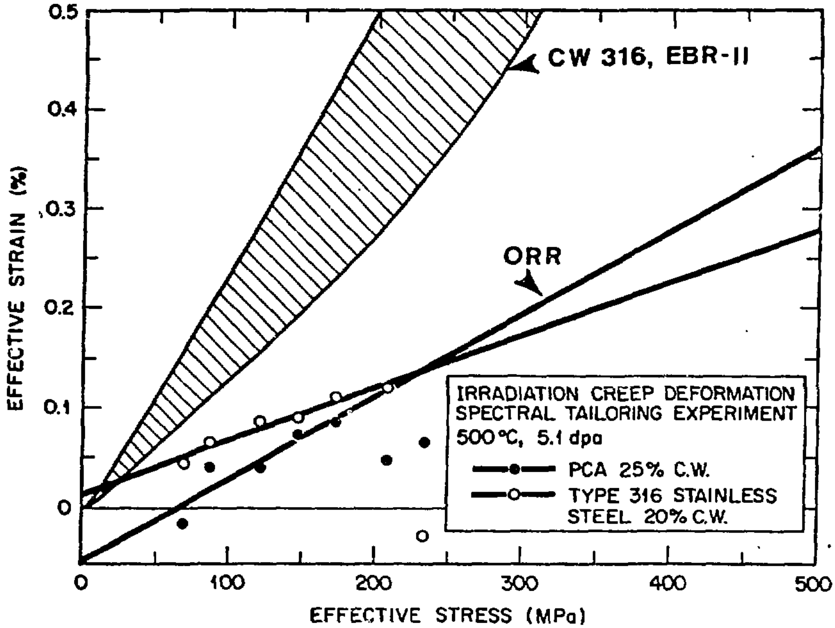


Fig. 5

DISCLAIMER