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SWELLING AND SWELLING RESISTANCE POSSIBILITIES OF AUSTENITIC STAINLESS STEELS IN FUSION REACTORS*

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Fusion reactor helium generation rates in stainless steels are intermediate to those found in EBR-II and HFIR, and swelling in fusion reactors may differ from the fission swelling behavior. Advanced titanium-modified austenitic stainless steels exhibit much better void swelling resistance than AISI 316 under both EBR-II (up to ~120 dpa) and HFIR (up to ~44 dpa) irradiations. The stability of fine titanium carbide (MC) precipitates plays an important role in void swelling resistance for the cold-worked titanium-modified steels irradiated in EBR-II. Furthermore, increased helium generation in these steels can (a) suppress void conversion, (b) suppress radiation-induced solute segregation (RIS), and (c) stabilize fine MC particles, if sufficient bubble nucleation occurs early in the irradiation. The combined effects of helium-enhanced MC stability and helium-suppressed RIS suggest better void swelling resistance in these steels for fusion service than under EBR-II irradiation.

1. INTRODUCTION

Soon after voids were discovered[1] in FBR-irradiated type 316 stainless steel (316), almost eighteen years ago, the materials and breeder reactor design communities realized the adverse consequences of unpredictably high swelling.[2] More recently, high void swelling has also been found to limit first wall lifetimes in conceptual designs of blankets for magnetic fusion reactors (MFR).[3,4] Because fusion is still conceptual, swelling must be studied using existing irradiation facilities, particularly mixed spectrum and fast breeder reactors. Extrapolation of fission reactor data to fusion reactor conditions must account for the different neutron energy spectra.

Fusion, being a more recent technology, derives some aspects of its materials programs from previous FBR programs, particularly in the area of alloy development. In the U.S. Breeder Program, attempts to improve the swelling resistance of the 300-series stainless steels resulted in the development of the "D9" type alloy in 1975 to 1977 (refs. 5–7). These alloys are basically (in wt %) 14.5 Cr, 14.5 Ni steels which also contain ~0.25 Ti (ref. 7). Both neutron and heavy ion irradiations have demonstrated the improved swelling resistance of these alloys compared to AISI 316 (refs. 7–13).

In 1974 to 1976, comparative irradiations of type 316 in EBR-II and HFIR (a mixed fast and thermal neutron spectrum light-water reactor) suggested that large changes in helium generation rate could affect void swelling,[14,15] as anticipated by theoretical studies.[16,17] However, in spite of the higher helium generation, HFIR data also suggested that cold work and titanium additions offered metallurgical avenues toward swelling resistance, as had been found under the FBR irradiations.[15,18,19] Therefore in 1977 to 1978, the Alloy Development for Irradiation Performance (ADIP) program initiated work to develop irradiation-resistant structural materials for MFRs.[20,21] The ADIP program was able to immediately select a Prime Candidate Alloy (PCA) for its austenitic stainless steel path (Path A) on the basis of the FBR program's "D9" type alloy, without the need for scoping studies. The PCA is a 14 Cr, 16 Ni, 0.24 Ti steel, otherwise similar to AISI 316.

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This paper summarizes fission reactor swelling data that show how helium affects swelling resistance in austenitic stainless steels. Mechanistic understandings from these data are also summarized. These data suggest several possible metallurgical avenues for further improvements to the swelling resistance of PCA.

2. SWELLING RESISTANCE UNDER FBR IRRADIATION

The fluence dependence of swelling can conveniently be described as a low-swelling transient period followed by an acceleration to a regime of near linear swelling. Such behavior is shown schematically in Fig. 1. Designers often approximate swelling curves with a bilinear model; the point at which the linear swelling curve extrapolates to intercept the fluence axis is termed the incubation fluence (τ).[22,23] Commercial stainless steels tend to have much longer low-swelling transients relative to high-purity alloys[9,22,24,25] (see Fig. 1). Cold work further delays swelling.

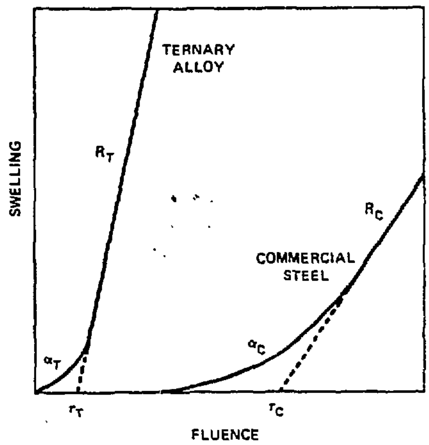

FIGURE 1
Schematic relative fluence dependence of void swelling for a commercial austenitic stainless steel (i.e., AISI 316) and a related ternary (Fe-Cr-Ni) alloy irradiated in an FBR. R_T, R_C designate regions of near linear swelling, α_T, α_C denote the accelerated swelling regimes at the end of the low-swelling transient, and τ_T, τ_C denote linearly extrapolated incubation fluence.

After the transient regime, swelling in commercial alloys may be similar to high-purity alloys;[26] however, the high swelling rate in the linear regime is of little interest to alloy development. Extended transient regimes are the goal of alloy development.

The general swelling behavior of several heats of 20%-CW AISI 316 [first core heats for the Fast Flux Test Facility (FFTF)] irradiated in EBR-II are shown in Fig. 2. Trend bands are drawn from data by Bates and Korenko,[23] Yang and Garner,[27] and Brager and Garner;[28] the bands also approximately describe the behavior of N-lot 316 (refs. 23,27). The data fall into two

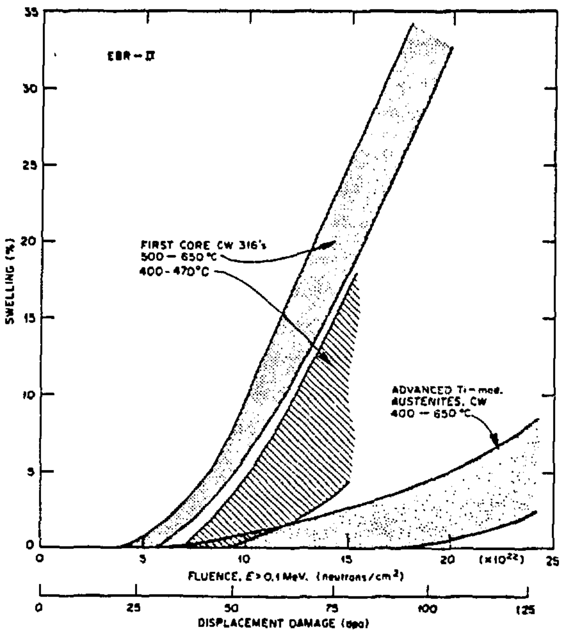

FIGURE 2
Scatter bands representing the average swelling behavior of several heats of 20%-CW AISI 316 (first core candidates for FTF fuel clad) as functions of fluence in EBR-II at 400 to 650°C after data by Bates and Korenko,[23] Yang and Garner,[27] and Brager and Garner.[28] Also included is a scatter band for the relative behavior of 20 to 25%-CW advanced titanium-modified austenitic stainless steels constructed for schematic representations of data by Laidler and Bennett[6] and Chin et al.[7]

distinct bands, covering the temperature ranges of 500 to 650°C and 400 to 470°C.

The swelling behavior of several CW "D9" type alloys is also shown in Fig. 2 by a scatter band, following schematic treatments of the data by Chin et al.[7] and Laidler and Bennett.[6] Swelling is clearly much lower in the CW "D9" type alloys than in the first core heats of 20%-CW AISI 316 because the low-swelling transient regime is extended. Higher fluence data are required to determine the onset of the linear swelling regime.

3. MICROSTRUCTURAL DEVELOPMENT FOR FBR SWELLING RESISTANCE

Swelling in an FBR is the result of microstructural nucleation and growth of voids (bias-driven cavities). Precipitation in austenitic stainless steels can also cause small amounts of densification (carbides) or swelling (intermetallics).[29] Swelling resistance is therefore obtained when void nucleation and growth are suppressed. In this section we will examine data and mechanisms explaining void suppression in the CW "D9" type steels relative to 20%-CW AISI 316.

Delayed void formation and growth cause extension of the low-swelling transient for the CW advanced titanium-modified alloys relative to 20%-CW AISI 316. Voids begin to form in 20%-CW AISI 316 from 30 to 40 dpa and swelling increases to ~11 to 13% at ~70 dpa [Figs. 2 and 3(a)]. By comparison, few voids form in the CW titanium-modified alloy, even at 66 dpa [Fig. 3(b)].

Several factors appear to be involved in the suppression of void formation in the titanium-modified steels. Ion irradiation experiments demonstrate that with titanium additions helium is necessary for void nucleation,[10,11] whereas helium aids, but is not required, for void nucleation in unmodified AISI 316 (ref. 30). This suggests that titanium strongly getters residual gases[12] (like oxygen). Therefore, under FBR irradiation, void formation may be retarded in the titanium-modified steels because helium generation alone must supply gas to nucleate bubbles as potential void embryos.

Titanium additions also cause precipitation of titanium (MC) carbide at elevated irradiation temperatures. Precipitation of MC occurs readily in heavily cold-worked titanium-modified austenitic steels during thermal aging at temperatures above 500 to 550°C, with little long-term coarsening below 700 to 750°C (refs. 31–33). A

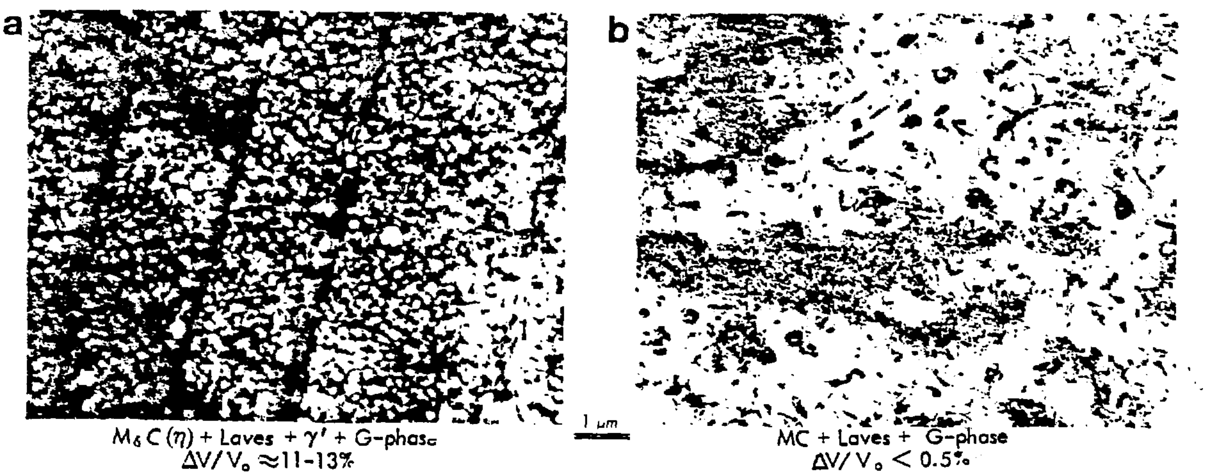

FIGURE 3
A comparison of void formation during EBR-II irradiation at 510 to 535°C of (a) CW 316 (DO-heat) to 69 dpa and (b) a CW advanced titanium-modified austenitic stainless steel to 66 dpa (courtesy E. H. Lee, ORNL).

typical structure of fine, dense MC particles decorating the dislocation structure after 10,000 h at 650°C is shown in Fig. 4(a). Fine MC also forms in the CW "D9" type alloys under EBR-II irradiation above 500°C; the MC microstructure after 37 dpa (~10,000 h) at 650°C is shown in Fig. 4(b). The MC is unstable below 500°C in EBR-II and will dissolve if introduced via preirradiation thermal treatments.[31] Fine MC particles, when stable, can trap helium at interfaces and the bubble distribution is substantially refined relative to unmodified austenites.[32,34,35] Fine MC particles can also pin dislocations,[13] increase the overall sink strength (aiding mutual defect annihilation),[36] and absorb excess vacancies (due to oversized misfit).[34] Together these factors hinder void formation and growth from bubbles.[37,38] Fine MC formation also tends to prevent development of coarser phases like M_6C, Laves, and G phases, which delays the formation of very large precipitate associated voids.[13,36] Under FBR irradiation, MC stability appears quite important to maintaining void swelling resistance.

Radiation-induced solute segregation (RIS) leads to void swelling and to eventual MC instability. Due to several superimposed mechanisms, RIS in AISI 316 results in nickel and silicon enrichment and chromium and molybdenum depletion at point defect sinks.[39] Void swelling and RIS are strongly coupled phenomena in AISI 316 under FBR irradiation; the effects of RIS appear to be very strong at the end of the low-swelling transient regime,[40,41] as shown in Fig. 5. RIS results in the formation of phases highly enriched in nickel and silicon (and often lower in molybdenum and/or chromium than thermally produced phases) near the end of the low-swelling transient. These can be described as radiation-induced, -modified, or -enhanced phases [M_6C (η), high-nickel Laves, γ', G, and phosphide] which couple positively to RIS.[31,41] Possible effects of RIS on precipitation are illustrated in Fig. 6(a,b). Phases naturally rich in nickel and silicon can be aided by enrichment of these elements at their interfaces via RIS as the particles form under irradiation[41-44] [Fig. 6(a)]. Conversely, phases which are naturally nickel and silicon poor and unable to enrich substantially in these elements, may be hindered by RIS [Fig. 6(b)]. Two such phases appear to be $M_{23}C_6$ and MC (both titanium-and niobium-rich).[43] The $M_{23}C_6$ is not enhanced and is often retarded (particularly at higher fluences) for FBR irradiations of AISI 316 compared to thermal aging.[31,32,41] The MC composition and its relationship to the formation of RIS-induced phases suggest a negative RIS coupling for the MC phase as well.

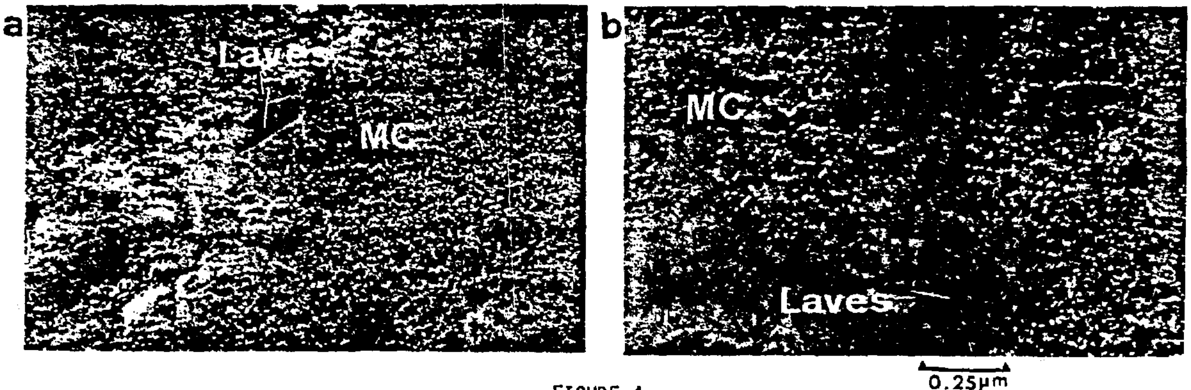

FIGURE 4
Comparison of MC precipitate microstructures (via TEM dark field) of a CW advanced titanium-modified austenitic stainless steel produced by (a) thermal aging at 650°C for 10,000 h and (b) EBR-II irradiation at 650°C to 37 dpa (~10,000 h)(courtesy E. H. Lee, ORNL).

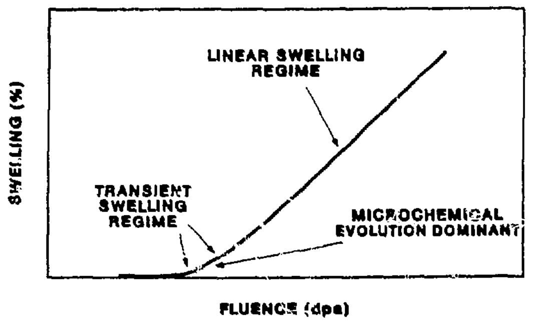

FIGURE 5
A schematic of the relationship of microchemical evolution (due to RIS) to the onset of steady state void swelling, after Garner and Wolfer (DAFS Quart. Prog. Rept. July—Sept., 1983, DOE/ER-0046/11, pp. 101—112).

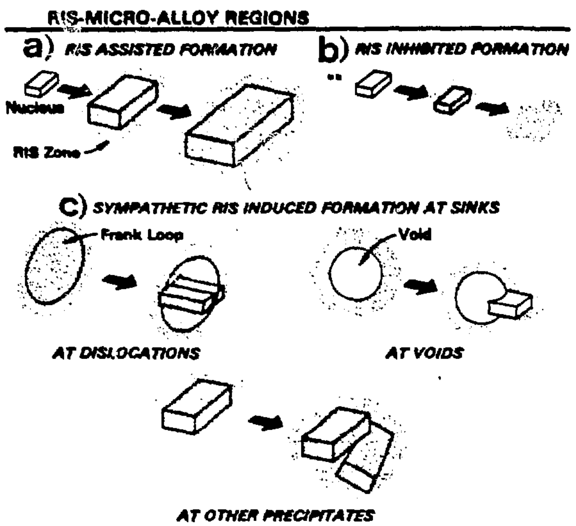

FIGURE 6
A conceptual illustration of the couplings between the development of individual precipitate phase particles and microcompositional regions produced by radiation-induced solute segregation (RIS) under irradiation. Examples are for the cases of (a) RIS-enhanced or -modified thermal phases like $M_6C(\eta)$ or Laves, (b) RIS-retarded phases like $M_{23}C_6(\tau)$ and possibly MC, and (c) RIS-induced phases like Ni_3Si (γ'), $M_6Ni_{16}Si_7$ (G) phosphides and possibly high-nickel Laves. Cases (a) and (c) represent positive RIS/phase couplings and (b) represents a negative coupling, based on combined ideas presented by Lee et al.[31] and Maziasz.[41]

Lee et al.[31] found MC did not form in a CW 14.5 Cr, 14.5 Ni titanium-modified steel irradiated in EBR-II below 500°C, coincident with RIS-induced $\gamma'(Ni_3Si)$ formation. The MC still remains stable in a similar alloy to ~66 dpa in EBR-II at 535°C [Fig. 3(b)], despite development of Laves and G phases;[45] higher fluence microstructural data on these alloys are unavailable. Ion irradiation studies of CW LS1B (a similar advanced titanium-modified steel) by Rowcliffe and Lee[46] indicate the MC eventually becomes unstable as G phase develops at high fluences [Fig. 7(a)]. Figure 7(b,c) shows the compositional contrast between the two phases.[47] The MC seems incompatible with the silicon and nickel enrichments and molybdenum depletion caused by RIS, whereas G phase is induced by them. The rapidly formed MC does not dissolve immediately upon formation of G phase, and Rowcliffe and Lee suggest that cascade dissolution of MC particles together with titanium absorption by the G phase lead to eventual instability of the MC phase. Earlier G phase

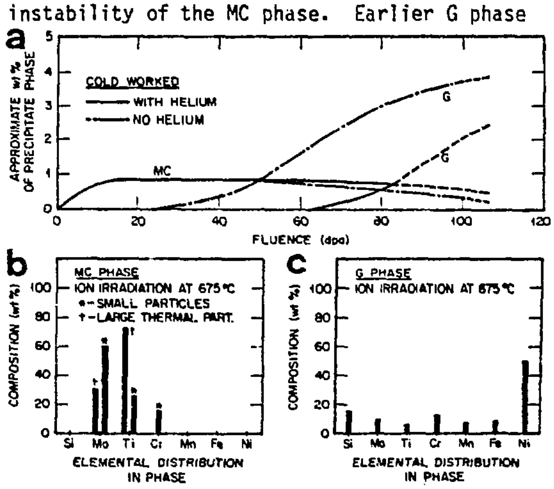

FIGURE 7
A comparison of relative evolutions of MC and G-phases in CW LS1B under ion irradiation at 675°C with and without 20 at. ppm/dpa of simultaneously injected helium. (a) Approximate fractions of phases (wt %) as functions of fluence, from Rowcliffe and Lee[46] and compositions of the (b) MC and (c) G phases, from Lee et al.[47]

formation in similarly irradiated solution-annealed (SA) LS1B does, however, affect MC formation and hastens its instability at higher fluence.[46] In summary, MC appears to couple negatively with RIS, either directly during its formation or indirectly through RIS-induced development of other phases. Once MC becomes unstable, void swelling resistance erodes as well.

4. THE EFFECTS OF HELIUM ON VOID SWELLING RESISTANCE

The effects of helium on swelling have been reviewed.[37,48,49] Earlier FBR program studies found small amounts (30 ppm) of preinjected helium to have little effect on void swelling under either ion or neutron irradiation. However, Keefer and Pard[50] did find that helium accelerated void swelling in SA 316 under light ion irradiation. More importantly, Harkness et al.[51] (SA 304) and Maziasz[41] (SA 316) found cold preinjection of 80 to 110 at. ppm He to completely suppress void formation under EBR-II irradiation. Studies of the effects of higher levels of continuous helium generation confirm these two opposite responses of either enhanced or suppressed void swelling.[41,47-49,52]

Comparison of EBR-II and HFIR irradiations of AISI 316 has comprised a major portion of the fusion program's efforts to study helium effects. Irradiation of nickel-bearing alloys near the fuel in HFIR produces much more helium (15–70 at. ppm/dpa, due to thermal neutron reactions with ^{58}Ni) than EBR-II (0.5 at. ppm/dpa), but at fairly similar displacement damage rates. Not only does helium generation vary between EBR-II and HFIR, but neutron energy spectrum and solid transmutations differ as well (manganese burnup and vanadium production in HFIR).[49,53] However, no present facility perfectly simulates fusion. Several recent studies and reviews[41,49,54,55] consider these energy spectrum and transmutation effects and find them minimal; they conclude that the helium/dpa ratio is the major variable of influence in the EBR-II versus HFIR swelling comparison. Dual ion studies (single variable) find similar effects of helium on swelling to support this conclusion.[49,53,55]

Swelling data for comparison of EBR-II and HFIR irradiation of DO-heat type 316 irradiated at 425 to 650°C are summarized in Fig. 8. Increased helium generation in HFIR leads to accelerated void swelling for the SA (DO-heat)

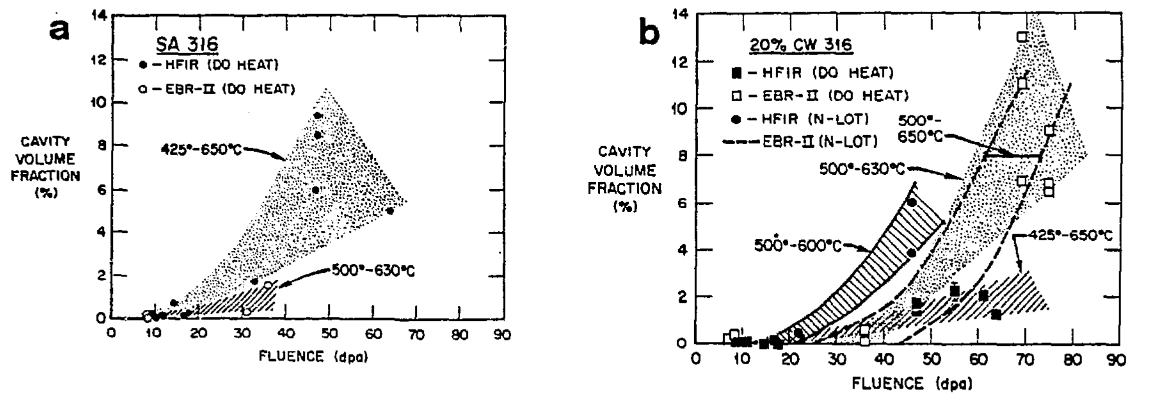

FIGURE 8
A comparison of cavity volume fraction data for (a) SA 316 and (b) CW 316 irradiated in EBR-II and HFIR at 400 to 650°C to evaluate the effects of helium on void swelling. Increased helium enhances void swelling for SA (DO-heat) 316 and CW (N-lot) 316 irradiated in HFIR compared to EBR-II whereas it suppresses void formation in CW (DO-heat) 316. The SA and CW (DO-heat) 316 data are taken from Maziasz,[63] the CW (N-lot) 316 HFIR data are from Maziasz and Braski,[66] and the scatter for CW (N-lot) 316 in EBR-II is taken from data by Yang and Garner.[32]

316 [Fig. 8(a) and refs. 15, 42, 56], whereas helium appears to suppress void swelling for CW (DO-heat) 316 [Fig. 8(b) and refs. 41,49,52,54, 56]. Brager and Garner[28,55] disagree and reach different conclusions, but recent detailed microstructural data and insight help explain these swelling differences.[52] Another recent EBR-II versus HFIR comparison is available for CW (N-lot) 316 from Maziasz and Braski.[57] Figure 8(b) shows the EBR-II swelling trend band for CW (N-lot) 316, taken from data by Yang and Garner;[27] there is little swelling difference between CW (DO-heat) and (N-lot) 316s in EBR-II. The HFIR data on CW (N-lot) 316 indicate accelerated void swelling behavior compared to EBR-II data. The swelling variation between CW (DO-heat) and (N-lot) 316s is larger in HFIR than in EBR-II, indicating that increased helium generation also affects the sensitivity of swelling of CW AISI 316 to heat-to-heat compositional differences.

A variety of pretreatments of PCA have been irradiated in HFIR, and lowest swelling was found in 25%-CW (A3) material.[57] In the range 400 to 600°C, PCA-A3 was found to be much lower swelling than CW (N-lot) 316 (see Fig. 9). Swelling for PCA at these temperatures is similar or slightly lower than the already low swelling of CW (DO-heat) 316 (ref. 57). HFIR fluences (44 dpa) on CW PCA are not sufficient yet to judge whether the low-swelling transient period will be longer or shorter than that observed for the CW "D9" type alloys in EBR-II. Further, neither alloy has yet been irradiated in both reactors to allow an accurate assessment of helium effects on the transient behavior.

The effects of helium on the swelling behavior of advanced titanium-modified steels has been studied by Lee and coworkers[13,46,47] under heavy ion irradiation. As before, helium can enhance void swelling in SA material. Dual ion irradiations at 675°C and 0.4 at. ppm He/dpa to 70 dpa produce minimal void swelling in both SA

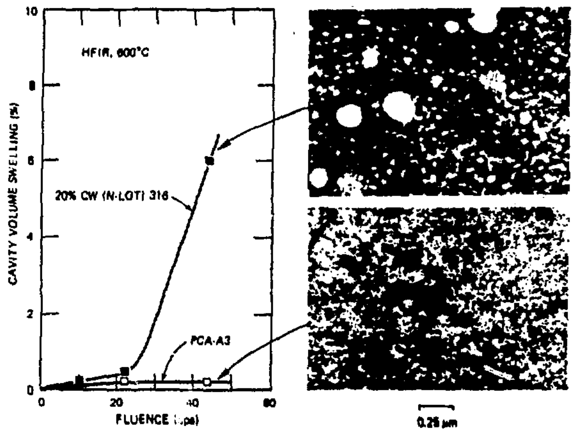

FIGURE 9
Comparison of cavity volume fractions as functions of fluence for 20%-CW (N-lot) 316 and 25%-CW (A3) PCA irradiated in HFIR at 600°C. Micrographs of both at the highest fluences show that void formation was suppressed in the PCA-A3. Data of Maziasz and Braski.[57]

and CW LS1B, as shown in Fig. 10(a,b). Similar irradiations at a higher helium/dpa ratio of 20 cause a large enhancement of void swelling in the SA material, but not in the CW LS1B [Fig. 10(c,d) compared to (a,b)]. The different responses of swelling to increased helium for the SA and CW LS1B parallel exactly the great swelling differences found between SA and CW DO-heat 316 (Fig. 3) and SA and CW PCA [Fig. 10(e,f)] irradiated in HFIR. These results emphasize the importance of cold work in achieving swelling resistance for fusion.

5. HELIUM EFFECTS ON MICROSTRUCTURAL DEVELOPMENT FOR VOID SWELLING RESISTANCE

Increased helium generation usually increases the rate of bubble nucleation. However, sufficiently increased bubble nucleation can lead to (a) suppressed void formation, (b) suppressed RIS effects on precipitation, and (c) stabilization of fine MC phase particles. These features can be recognized at lower fluences and appear to correlate with helium-suppressed void swelling at higher fluences.

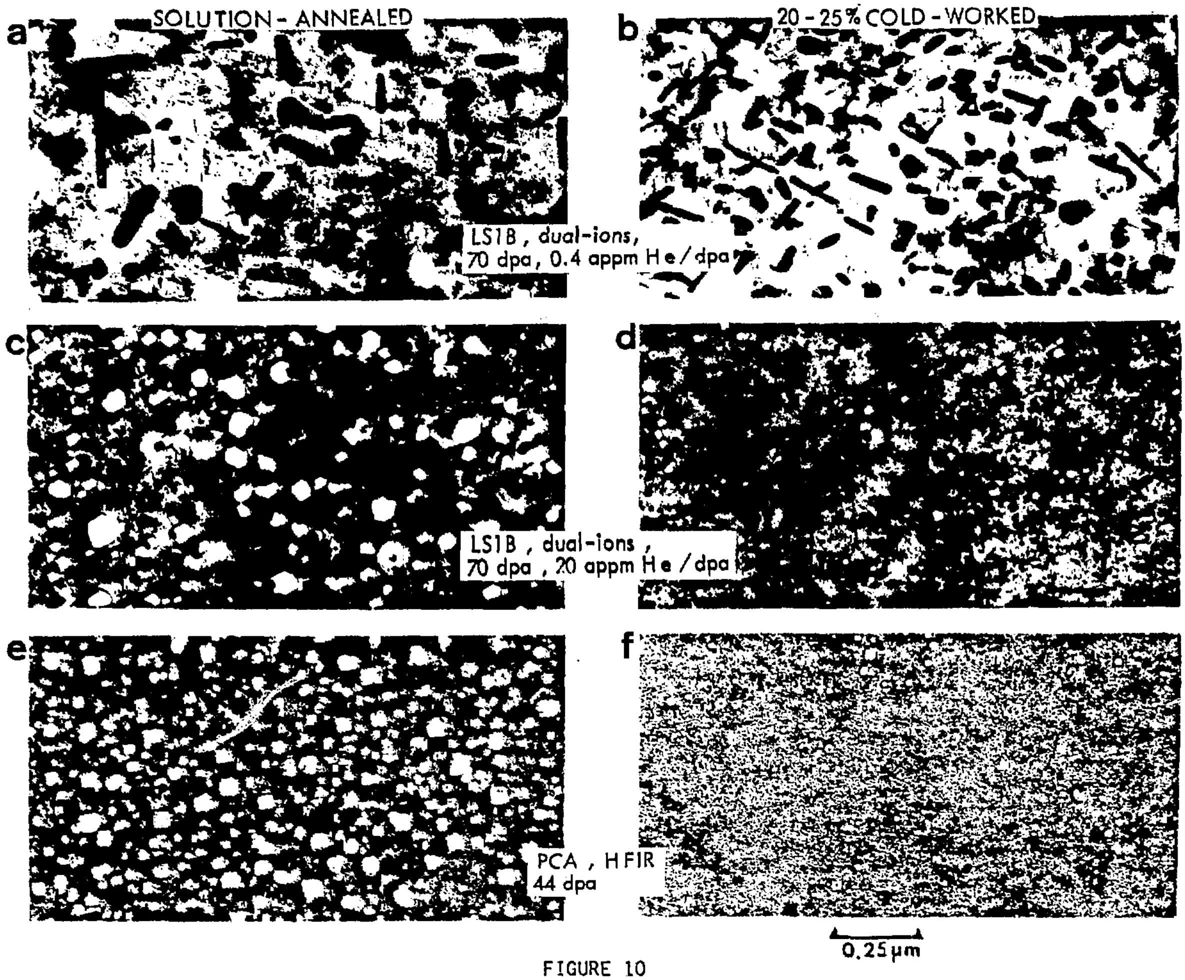

FIGURE 10
A comparison of microstructures to illustrate the different effects of increased helium generation on void formation between SA and CW material for the advanced titanium-modified austenitic stainless steels. Results of dual ion irradiations of LS1B at 675°C to 70 dpa for (a) SA and (b) CW material at 0.4 at. ppm He/dpa, (c) SA, and (d) CW material irradiated at 20 at. ppm He/dpa, from data of Lee and coworkers[46,47] (micrographs courtesy E. H. Lee, ORNL). Results of HFIR irradiations at 600°C to 44 dpa (e) SA and (f) CW PCA are taken from Maziasz and Braski.[57]

Odette and coworkers[21,22] first predicted that void swelling could be suppressed by sufficient bubble nucleation. Bubble-suppressed void growth is also suggested by recent cavity microstructural data for CW (DO-heat) 316 irradiated in HFIR compared to EBR-II or compared to SA (DO-heat) 316 also irradiated in HFIR (refs. 41,49,52,54,56). Microstructural data on SA and CW LS1B under dual-ion irradiation[46,47] [Fig. 10(c,d)], and for PCA under HFIR irradiation[57] [Fig. 10(e,f)] again indicate a similar interpretation for void suppression in the CW material. Conversely, early conversion of bubbles to voids leads to helium-enhanced void swelling,[38,58] as observed for SA (DO-heat) 316 and CW (N-lot) 316 irradiated in EBR-II and HFIR[41,52,57] (see Fig. 3). Theoretical modeling suggests several reasons for suppressed conversion of bubbles to voids;[16,17,36-38,58-60] (a) reduced bias,

(b) increased critical radius or decreased number of gas atoms per cavity, (c) bubble populations which become dominant sinks, (d) dislocation pinning by bubbles, and (e) matrix or interfacial bubble densities which hinder development of precipitate-associated voids. However, a surprising and important role of helium is its effect on RIS and precipitation under irradiation.

Inert gas atoms seem unlikely to directly affect solid-state chemical reactions. However, Maziasz[61] [SA (DO-heat) 316, HFIR] and Kenik[10] (SA LS1A, dual ion) presented data in 1978 to 1979 indicating effects of helium on precipitation. However, Maziasz observed helium to enhance precipitation kinetics whereas Kenik observed it to suppress precipitation under irradiation. Both Kenik[10] (for loops) and Odette[60] (for bubbles) proposed that increased sink strengths would dilute RIS and thereby delay radiation-induced precipitation (RIP). Further ion work confirmed these early data and ideas. The strongest effects are found for cold-preinjected helium or dual-beam irradiations with high helium/dpa ratios (refs. 11–13, 36,37,46–49,54) [cf. Fig. 10(a,b) with (c,d)].

Recent phase identity and compositional studies of precipitation in the DO-heat of AISI 316 (including EBR-II-irradiated, HFIR-irradiated, and thermally aged specimens) have provided further evidence for the influence of helium on precipitation behavior under irradiation. In the irradiation temperature range of 400 to 650°C, when helium suppresses the development of bias-driven voids, both RIS and RIP are suppressed. However, precipitation during irradiation is still enhanced because phases similar in character to thermal precipitation occur, but at an accelerated rate. Conversely, if an increase in the He/dpa ratio stimulates void formation, then RIS and/or RIP are also accelerated.

The situation of helium suppressed RIS/RIP is illustrated by the phases which develop in CW 316 (DO-heat) in three different environments, as shown in Fig. 11(a). The RIPs (γ' and G phases) which develop sluggishly in EBR-II do not develop at all during HFIR irradiation. However, the thermal phases ($M_{23}C_6$ and a low nickel-Laves phase) develop more rapidly and more abundantly in HFIR than during either thermal aging or EBR-II irradiation. Differences in the extent of RIS are also reflected in the composition of the Laves phases which develops in the two reactors [Fig. 11(b)]. In EBR-II, the substantial increase in nickel content with increasing fluence indicates a high level of RIS. Conversely, the low constant level of nickel in the Laves phase during HFIR irradiation indicates a suppression of RIS. This suppression is believed to be related to the presence of bubble-dominated microstructures and suppression of bias-driven void growth also found in the samples.[52] Recent work by Loomis et al.[63] also indicates such a relationship. They found RIS at voids to decrease when void swelling saturated in a cavity-dominated microstructure, as shown in Fig. 12, in a dual-ion irradiated Fe–20 Ni–15 Cr alloy.

The converse situation of enhanced void formation and RIS due to increased helium is manifest by the phase evolution in SA 316 (DO-heat) in the same three exposure environments (Fig. 13). Precipitate development is enhanced in HFIR, but RIPs are absent compared to EBR-II [Fig. 13(a)]. However, the accelerated increase in Laves phase nickel content with fluence indicates that the extent of RIS is greater in HFIR than in EBR-II. This effect of helium on RIS is opposite to that found in the CW 316 (DO-heat) irradiated in the two reactors. However, it does further indicate a consistent relationship between void formation and RIS development through the avenue of the cavity evolution. Increased helium stimulates

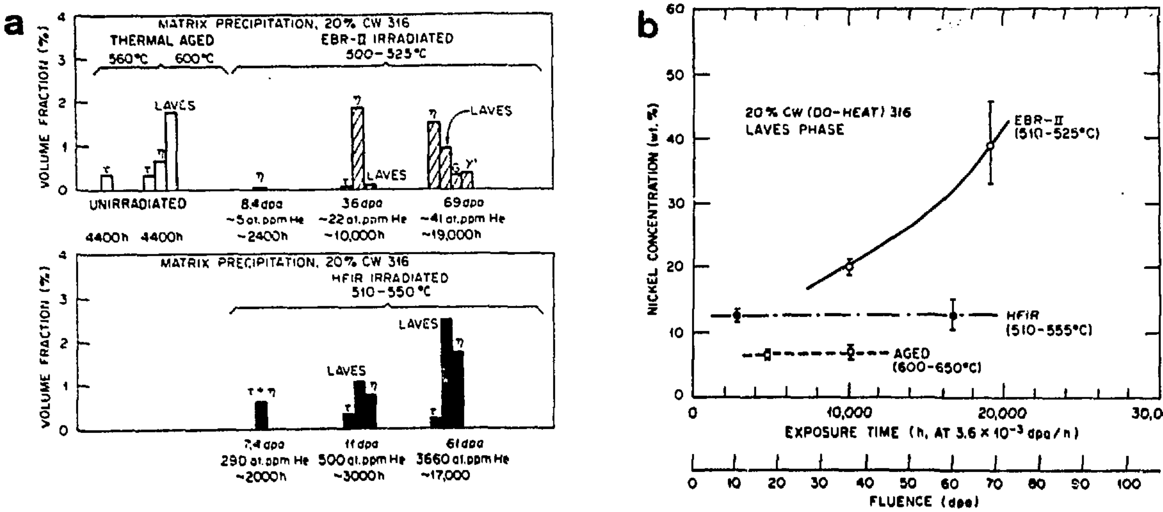

FIGURE 11
A comparison of (a) relative phase fractions of the total precipitation and (b) nickel concentration of the Laves phase as functions of fluence for 20%-CW (DO-heat) 316 irradiated at 500–550°C in EBR-II and HFIR. These are determined via quantitative analytical electron microscopy (AEM), and data from thermally aged specimens are also included.[32,41,62] Thermal precipitation develops early and irradiation induced phases (γ′ and G) are absent in HFIR. The low Laves phase nickel content in HFIR also indicates that RIS effects are suppressed.

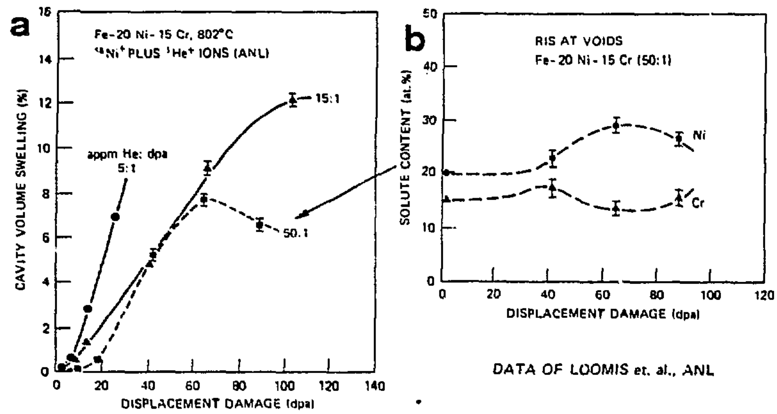

FIGURE 12
A correlation of (a) void swelling evolution with fluence at various helium-to-dpa ratios and (b) RIS evolution at voids, as judged from nickel enrichment and chromium depletion measured via AEM for SA Fe–20 Ni–15 Cr irradiated with dual ions at ~800°C and 50 at. ppm He/dpa by Loomis et al.[63] Note that as the void swelling saturates and begins to decline at 60 to 85 dpa, so does RIS at the voids.

both bubble development leading to enhanced void formation,[52] as well as enhanced RIS. This relationship is also supported by observations of enhanced codevelopment of the RIP γ′ and voids in SA PCA and CW 316 (N-lot) (10 dpa, 500–600°C)[57,64,65] and CW 316 (DO-heat) (~10 dpa, 425–450°C)[32] irradiated in HFIR.

Finally, helium enhances the stability of the MC phase when the effects of RIS are suppressed, apparently due to several superimposed and reinforcing mechanisms. The formation of MC itself under irradiation promotes much higher bubble concentrations than are found in unmodified austenites under the same irradiation and/or helium conditions (see Fig. 14). Fine MC particles trap helium and are known to form fine, stable bubbles at their interfaces.[13,18,32,34,35]

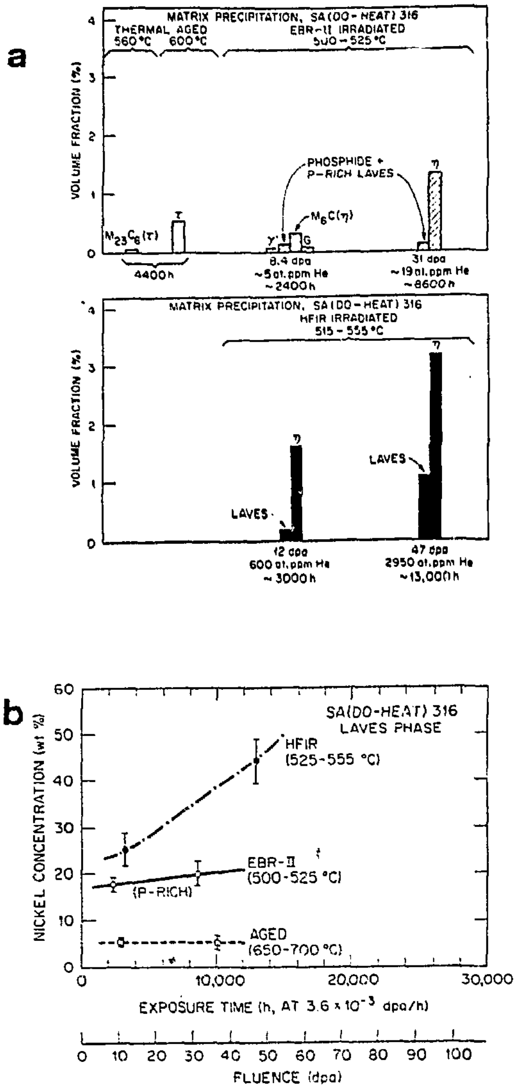

FIGURE 13
A comparison of (a) relative phase fractions of the total precipitation and (b) nickel content of the Laves phase as functions of fluence for SA (DO-heat) 316 irradiated at 500 to 550°C in EBR-II and HFIR.[41,62] Data are also included from thermally aged specimens.

Furthermore, high concentrations of bubbles would be expected to dilute RIS. Figure 15 shows that MC forms and remains stable in CW titanium-modified steels under HFIR irradiation

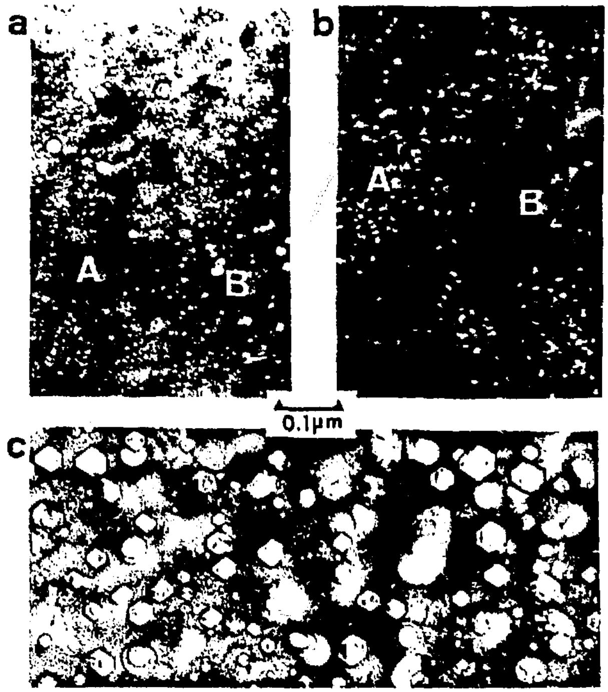

FIGURE 14
Correlation of (a) cavity and (b) MC precipitate (via dark field) microstructures for 25%-CW PCA irradiated in HFIR at 600°C to 44 dpa to illustrate the spatial correspondence of fine bubbles and fine MC particles. Similarly irradiated 20%-CW (N-lot) 316 is included (c) to illustrate the tremendous bubble refinement obtained with MC precipitation by comparison with (a). Data from Maziasz and Braski.[57]

at temperatures of 300 to 350°C and above,[31,32,65] whereas MC either does not form or dissolves below 500 to 525°C in EBR-II.[31] Consistently, this region of stable MC formation in HFIR coincides with extremely fine bubble populations, virtually no RIS/RIP effects, and no voids.[32,57,64-66] Void formation and RIS are similarly suppressed when MC is stabilized by increased helium, for CW LS1B under dual-ion irradiation[13,46,47] [Fig. 7(a) and 10(a,d)]. Helium-enhanced MC stability and suppressed RIS are essential for expecting helium to extend the low-swelling transient for fusion compared to FBR irradiation of the CW "D9" PCA type alloys.

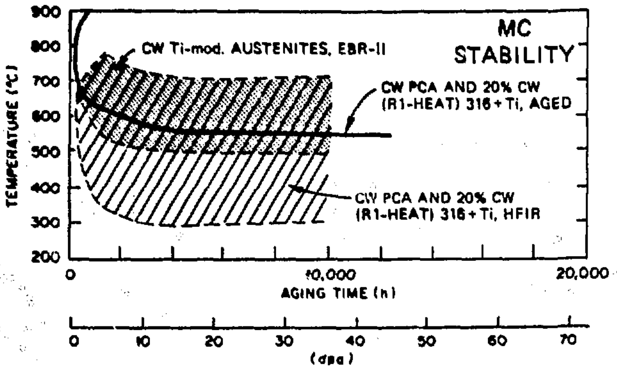

FIGURE 15
A time (fluence)-temperature-precipitation plot of the thermally stable titanium (MC) carbide phase under neutron irradiation and/or thermal aging. Data are combined from Maziasz and coworkers[32,57,62,64,65] and Lee et al.[12] For these CW titanium-modified steels, helium-suppressed void formation and RIS correlate with MC stability to much lower temperatures in HFIR than found in EBR-II.

6. SUMMARY OF ANTICIPATED FUSION SWELLING BEHAVIOR

Stoller and Odette[58] recently predicted greater swelling for 20%-CW AISI 316 in fusion reactor service than found in either EBR-II or HFIR. Bubble nucleation is predicted to be intermediate to that found in EBR-II and HFIR and leads to enhanced void formation. The behavior of SA (DO-heat) 316 irradiated in HFIR confirms such helium-enhanced void formation, particularly when compared to CW (DO-heat) 316 irradiated in EBR-II and HFIR. More importantly, helium also accelerates the effects of RIS in the SA (DO-heat) 316 in HFIR to allow accelerated development of phases which couple positively to RIS. These coarse, RIS-compatible particles then become the sites for precipitate-associated voids. Similar swelling and microstructural behavior are found in SA PCA irradiated in HFIR at 500 to 600°C. Mattas et al.,[4] recently assessed the impact of swelling on first wall lifetimes. They showed that such high swelling is unacceptable for current fusion designs. Table 1 summarizes expected fusion service swelling behavior and underlying reasons for the effects of helium on swelling in various alloys. The possibility of accelerated void swelling relative to FBR irradiation seems quite likely for SA austenitic stainless steels above about 400 to 450°C, making them unattractive under those conditions.

Projection of the possible swelling behavior of 20%-CW 316 based upon HFIR and FBR data is more uncertain. For heats with chemistries similar to heat DO, it is possible that the low-swelling regime will persist to fluences well beyond 65 dpa in HFIR. High bubble nucleation suppresses void formation both in the matrix and at precipitates. Furthermore, RIS effects are suppressed and thermal precipitation is enhanced. However, Stoller and Odette's analysis[58] indicates that for the He/dpa ratio typical of a fusion first wall, the low-swelling regime would be shorter than either the FBR or HFIR environments because bubble nucleation is intermediate to both. For heats with chemistries similar to N-lot, both RIP and swelling are accelerated in HFIR compared to EBR-II, suggesting helium-shortened transient swelling regimes for fusion.

A fusion design window, constructed simply by combining EBR-II and HFIR data (Figs. 2 and 8; refs. 15,66) for a worst-case limit of 10%

Table I. A summary of swelling expectations for fusion reactor service

ALLOY	EXPECTED BEHAVIOR	REASONS
SA 316 OR Ti-MOD. AUSTENITES (LIKE PCA)	SAME OR ACCELERATED *SWELLING* COMPARED TO FBR	• POSITIVE RIS-PHASE *COUPLING* • HELIUM ENHANCED VOID RATHER THAN BUBBLE SWELLING • NO HELIUM SUPPRESSION OF *RIS*
CW 316	SAME OR ACCELERATED OR SUPPRESSED *SWELLING* COMPARED TO FBR	UNCERTAIN HELIUM EFFECTS
CW Ti-MOD AUSTENITES (LIKE PCA)	ENHANCED *SWELLING RESISTANCE* COMPARED TO FBR	• NEGATIVE RIS-PHASE COUPLING • HELIUM ENHANCED BUBBLE RATHER THAN VOID SWELLING • HELIUM SUPPRESSION OF RIS

swelling, is shown in Fig. 16. Only by operating below 400°C could lifetimes longer than 50 dpa be considered for 20%-CW AISI 316.

Similar reasoning, however, leads to an optimistic projection of swelling resistance for the CW advanced titanium-modified alloys for fusion. As indicated in Table I, the combined effects of helium-stabilized MC (a negative RIS coupled phase), high bubble concentrations promoted by fine MC particles, and bubble dilution of RIS all cooperate to suggest better swelling resistance for fusion compared to EBR-II irradiation. The effect of MC-refined bubble nucleation would cause more nucleation per increment of generated helium in the CW titanium-modified alloys compared to 20%-CW AISI 316 (higher scaling factors in the model of Odette and Stoller).[58] This effect is the primary reason for anticipating the bubble dominated/low RIS microstructural development

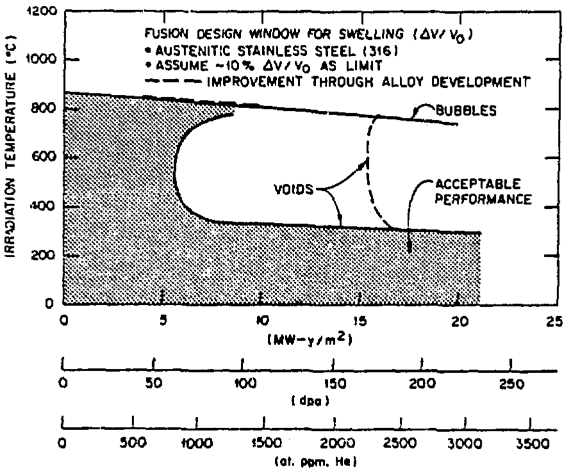

FIGURE 16
A schematic fusion first wall design window for swelling of 20%-CW AISI 316 in terms of irradiation temperature and exposure time. Void and bubble limits are drawn from combined EBR-II and HFIR data for several heats of steel.[15,23,27,57] The shaded region indicates swelling less than 10%, assumed acceptable for fusion.[3] The dashed lines indicate how development of more swelling resistant alloys widens the design window for fusion.

that correlates with void swelling resistance for the CW "D9" or PCA type alloys for fusion. If swelling for these alloys is simply confined to the lower end of the scatter band in Fig. 2, then the design window for swelling limitation at higher temperatures extends to significantly higher fluences, as indicated by the dashed curve in Fig. 16.

7. SUGGESTIONS FOR FUTURE WORK

Although current microstructural insight suggests extended low-swelling transients for 20 to 25% CW PCA type steels, this possibility needs better experimental confirmation. The CW "D9" and PCA alloys need to be irradiated to higher fluences in either FFTF or EBR-II (200 dpa or more) to establish the low-helium base line for FBR transients. Irradiations in HFIR should proceed to ~100 dpa, but higher fluences seem unattractive due to excessively high helium generation. Two experiments may confirm the hypothesis that helium extends the low-swelling transients and provide estimates of its duration for fusion. The first is simply to cold preinject 100 to 200 at. ppm He into the CW PCA and see if transients are longer than in uninjected FBR-irradiated material. Cold preinjection provides the finest bubble structures,[48,49] and if void swelling is not delayed by this helium treatment, it is unlikely to be delayed in fusion service. If void formation is suppressed, higher fluence irradiations should continue to test the duration of the helium-stabilized transient. A second experiment would be to take CW PCA type alloys irradiated in HFIR to ~10 to 20 dpa and in ORR to ~40 to 50 dpa, and then test their further swelling resistance in FFTF or EBR-II. These would provide more "natural" microstructures, evolved with high helium (500–1500 at. ppm) for comparison with the uninjected and helium preinjected FBR-irradiated materials. Together these FBR irradiations could then suggest reasonable estimates of low-swelling transients for fusion.

Aside from conceptual conjecture mentioned previously, there is no basis for expecting helium to enhance or suppress RIS through its effect on microstructural evolution formal RIS theory.[67,68] The theory on phase stability under irradiation emphasizes vacancy supersaturation and cascade (dissolution/reprecipitation) effects on the phases that form in steels,[69] but currently makes no provision for the observed effects of RIS on phase and compositional evolution. Modeling or theoretical work is needed to confirm these suspected effects and to better understand them.

Alloy development avenues to further extend the low-swelling transient to high fluences or temperatures appear to involve enhanced MC stability and suppressed RIS or RIP development. Aside from gross alterations to base alloy chemistry, these goals suggest adjusting carbon, silicon, titanium, and molybdenum concentrations for further optimization. Recent work by Lee et al.[70] also points out benefits of phosphorus addition on void swelling resistance. In summary, it appears essential to control phase evolution and RIS development in order to obtain the effects of helium that can lead to void swelling resistance for fusion.

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