A REVIEW OF IRRADIATION EFFECTS ON LWR CORE INTERNAL MATERIALS – NEUTRON EMBRITTLEMENT, VOID SWELLING, AND IRRADIATION CREEP

O. K. Chopra\textsuperscript{1} and A. S. Rao\textsuperscript{2}

\textsuperscript{1}Environmental Science Division
Argonne National Laboratory
Argonne, IL 60439

\textsuperscript{2}Division of Engineering
US Nuclear Regulatory Commission
Washington, DC 20555

Abstract

Austenitic stainless steels (SSs) are used extensively as structural alloys in the internal components of light water reactor (LWR) pressure vessels because of their relatively high strength, ductility, and fracture toughness. However, exposure to neutron irradiation for extended periods not only changes the microstructure and microchemistry of these steels, but also degrades their fracture properties. Other irradiation-related degradation issues are changes in material dimensions due to void swelling and stress relaxation due to radiation creep. The existing data on irradiated austenitic SSs are reviewed to determine the effects of key parameters such as material type and condition and irradiation temperature, dose, and dose rate on these processes. Differences in the radiation-induced degradation of fracture properties between LWR and fast-reactor irradiations are also discussed. The results are used to (a) define a threshold fluence above which irradiation effects on fracture toughness of the material are significant, (b) evaluate the potential of neutron embrittlement under LWR operating conditions, (c) assess the effects of void swelling, including its effect on fracture toughness, and (d) investigate the significance of irradiation creep relaxation on the functional integrity of reactor internal components.

1. Introduction

In light water reactors (LWRs), austenitic stainless steels (SSs) are used extensively as structural alloys in the internal components of reactor pressure vessels because of their relatively high strength, ductility, and fracture toughness. Fracture of these steels occurs by stable tearing at stresses well above the yield stress, and tearing instabilities require extensive plastic deformation. However, exposure to neutron irradiation for extended periods changes the microstructure (radiation hardening) and microchemistry (radiation-induced segregation or RIS) of these steels and degrades their fracture properties [1-11]. Loss of fracture toughness due to radiation embrittlement was not considered in the design of LWR core internal components constructed of austenitic SSs, but it has been considered in addressing nuclear plant aging and license renewal issues. In addition, dimensional changes due to void swelling [12-15] and stress relaxation due to radiation creep [16-20] are other aging degradation processes that affect LWR core internal components exposed to fast neutron radiation, and need to be considered in addressing plant aging issues.
The microstructural changes in austenitic SSs due to neutron irradiation vary with the irradiation temperature, neutron fluence, flux, and energy spectrum. At temperatures below 300°C (572°F), the material microstructure primarily consists of small (<5 nm) “black spot” defect clusters and faulted dislocation loops, whereas large faulted loops, network dislocations, cavities/voids (clusters of vacancies and/or gas bubbles), and precipitates are observed above 300°C [1-3]. Metal carbides are the primary precipitates in 300-series SSs under LWR conditions, although RIS of Ni and Si to sinks may lead to the formation of γ’ phase (Ni₃Si) and G phase (M₆Ni₁₆Si₇) [2,3].

The point defect clusters and precipitates act, to varying extent, as obstacles to dislocation motion, which leads to matrix strengthening, increase in tensile strength, and reduction in ductility and fracture properties of the material [1,5-8]. In general, cavities (or voids) are strong barriers, large faulted Frank loops are intermediate barriers, and small loops and bubbles are weak barriers to dislocation motion [1]. For SSs, a maximum yield strength occurs at irradiation temperatures near 300°C (572°F). Irradiation damage is characterized by either the neutron fluence in neutrons per square centimeter (n/cm²) or the average number of displacements experienced by each atom, i.e., displacements per atom (dpa).*

This review paper presents a critical assessment of the susceptibility of austenitic SSs to irradiation effects such as neutron embrittlement, void swelling, and stress relaxation in LWR environments. An assessment of the susceptibility of austenitic SSs to irradiation-assisted stress corrosion cracking (IASCC) is presented in a companion review paper published elsewhere in this journal. The existing data, in the open literature as well as Nuclear Regulatory Commission (NRC) and industry reports, have been evaluated to establish the effects of material parameters (such as composition, thermo-mechanical treatment, microstructure, and microchemistry) and environmental parameters (such as water chemistry, irradiation temperature, dose, and dose rate) on these processes. Differences in the radiation-induced degradation of fracture properties between LWR and fast-reactor irradiations are also discussed. The results are used to (a) provide a better understanding of the threshold fluence above which irradiation effects on fracture toughness of the material are significant, (b) evaluate the potential of neutron embrittlement under LWR operating conditions, including the synergistic effects of thermal and neutron embrittlement, (c) assess the effects of void swelling, including its effect on fracture toughness, and (d) investigate the significance of irradiation creep relaxation on the functional integrity of reactor internal components. The potential deficiencies or knowledge gaps in the existing experimental data on degradation of LWR core internal materials due to neutron irradiation are also discussed.

2. Neutron embrittlement and fracture toughness

Neutron irradiation can decrease the fracture toughness of austenitic SSs significantly, and failure may occur without general yielding. In such instances, a fracture mechanics methodology, such as elastic-plastic fracture mechanics (EPFM) or linear-elastic fracture mechanics (LEFM), is needed for analysis of structural integrity and development of inspection guidelines. The former involves the J integral-resistance (J-R) curve approach and is used where failure is caused by plastic deformation. The J integral is a mathematical expression used to characterize the local stress-strain field at the crack tip region (parameter J represents the driving force for crack propagation), and the J-R curve characterizes the resistance of the material to stable crack extension. The fracture toughness of such materials is represented by fracture mechanics parameters such as $J_{IC}$, the value of J near the onset of crack extension, and the tearing modulus, $T$, which characterizes the slope of the J-R curve:

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*In this study, the values of neutron fluence (n/cm²) were converted to dpa as follows: for LWRs, E>1 MeV, $10^{22}$ n/cm² ≈ 15 dpa; and for fast reactors, E>0.1 MeV, $10^{22}$ n/cm² ≈ 5 dpa.
where $E$ is the elastic modulus, $a$ is the crack length, and $\sigma_f$ is the flow stress defined as the average of the yield stress ($\sigma_y$) and ultimate stress ($\sigma_u$). The LEFM methodology is used where failure involves negligible plastic deformation. The fracture toughness of such materials is represented by the parameter $K_{Ic}$ (i.e., plane strain fracture toughness), which characterizes the resistance of the material to unstable crack extension. Under EPFM conditions, the equivalent critical stress intensity factor, $K_{Jc}$, can be determined from the saturation $J_{Ic}$ value using the relationship

$$K_{Jc} = \left( \frac{E'J_{Ic}}{1/2} \right),$$

where $E' = E/(1 - \nu^2)$, $E$ is the elastic modulus, and $\nu$ is the Poisson ratio.

The fracture toughness of austenitic SSs has been divided into three broad categories [6]. Category III corresponds to high toughness materials with $J_{Ic}$ above 150 kJ/cm$^2$ (857 in.-lb/in.$^2$). In these materials, fracture occurs after stable crack extension at stresses well above the yield stress. Category II corresponds to materials with intermediate toughness and with $J_{Ic}$ in the range of 30-150 kJ/cm$^2$ (171-857 in.-lb/in.$^2$). In Category II materials, fracture occurs by stable or unstable crack extension at stress levels close to the yield stress. Category I corresponds to low-toughness materials with $K_{Ic}$ less than 75 MPa m$^{1/2}$ (68.2 ksi in.$^{1/2}$) [$J_{Ic} < 30$ kJ/cm$^2$ (<171 in.-lb/in.$^2$)], and fracture occurs by unstable crack extension at stress levels well below the yield stress.

Nonirradiated wrought and cast austenitic SSs and their welds fall in Category III. The $J_{Ic}$ values for Type 304 and 316 SS at temperatures up to 125°C (257°F) vary between 169 and 1660 kJ/cm$^2$ (965 and 9479 in.-lb/in.$^2$), with a median value of 672 kJ/cm$^2$ (3837 in.-lb/in.$^2$) [6]. The $J_{Ic}$ values at 400-550°C (752-1022°F) are $\approx$35% lower, with a median value of 421 kJ/cm$^2$ (2404 in.-lb/in.$^2$). Fracture in such high-toughness materials is by the nucleation and coalescence of microvoids and is characterized by a dimpled fracture morphology.

Although cast austenitic SSs and SS welds also exhibit ductile fracture at temperatures up to 550°C (1022°F), their fracture toughness is lower than that of the wrought SSs. Also, a dimpled fracture morphology is observed in SS welds [21]. Because of a high density of inclusions in the weld, the dimples are relatively small and shallow, and often associated with inclusions. The overall fracture toughness of cast austenitic SSs and SS welds is controlled by the density and morphology of second-phase inclusions in these materials and varies with the cast or weld process. For example, static cast products have lower fracture toughness than centrifugally cast pipes. Gas tungsten arc (GTA) welds exhibit the highest toughness; shielded metal arc (SMA) welds have intermediate toughness; and submerged arc (SA) welds have the lowest toughness [6]. The median value of $J_{Ic}$ is 492 kJ/cm$^2$ (2809 in.-lb/in.$^2$) for GTA welds and 147 kJ/cm$^2$ (839 in.-lb/in.$^2$) for SA welds at temperatures up to 125°C (257°F).

Welding of austenitic SSs results in a heat affected zone (HAZ) adjacent to the fusion zone, where the material microstructure and microchemistry are greatly altered because of the precipitation of Cr-rich carbides at the grain boundaries. The formation of the carbides depletes Cr from the grain-boundary region, thereby creating a region that is susceptible to stress corrosion cracking (SCC). However, the fracture toughness of HAZ material is generally superior to that of the weld metal and may be comparable to that of the base metal.
2.1 Fracture toughness correlations

Neutron irradiation can degrade fracture toughness of austenitic SSs to the level of Category II or I. Until recently, most of the published data on neutron embrittlement of austenitic SSs had been obtained on materials irradiated in high-flux fast reactors, which use fast neutrons to burn more uranium-238 than conventional LWRs [22-38]. In these studies, the embrittlement of the materials has been characterized in terms of tensile properties, Charpy-impact properties, and fracture toughness. Typically, fracture toughness data for irradiated materials have been obtained from compact tension (CT) or single edge bend [SE(B)] specimens.

Fracture mechanics is a correlative technology and does not attempt to describe the mechanisms that are occurring at the crack tip. It correlates the behavior of reactor components with that of test specimens through the use of the K parameter (stress intensity factor). If two cracks have the same K, then they have the same strains and stresses in a region near the crack tip. For this correlation between specimen and component to work, K has to characterize the stresses and strains at the crack tip in the process zone. Mathematically, it can be shown that this is true if the plastic zone size is “small enough.” The K/size criteria are combined theoretical and empirical results that have been found to ensure the plastic zone is small enough and K is controlling. The American Society of Testing and Materials (ASTM) specifications for specimen K/size criteria are intended to ensure the applicability and transferability of the cracking behavior of a component or specimen of a given thickness under a specific loading condition to a crack associated with a different geometry, thickness, and loading condition. For constant load tests, ASTM E 1681 requires that

\[ B_{\text{eff}} \text{ and } (W - a) \geq 2.5 \left( \frac{K}{\sigma_y} \right)^2, \]  

where \( K \) is the applied stress intensity factor, \( \sigma_y \) is the yield stress of the material, \( a \) is crack length, \( W \) is specimen width, and \( B_{\text{eff}} \) is the specimen effective thickness, defined as \( (B B_N)^{0.5} \) (B and \( B_N \) are thickness and net thickness of the specimen, respectively). The specimen thickness (B or \( B_N \)) or remaining ligament \((W-a)\) ahead of the advancing crack is at least a factor of 8 greater than the plastic zone size for tests conducted in accordance with the K/size criterion of Eq. 3. For high strain-hardening materials [i.e., ratio of ultimate stress to yield stress \((\sigma_u/\sigma_y) \geq 1.3\)], the flow stress defined as \( \sigma_f = (\sigma_u + \sigma_y)/2 \) may be used instead of yield stress.

Because the K/size criterion was developed for materials that show work hardening, it may not be applicable for materials irradiated to fluence levels where, on a local level, they do not strain harden. This lack of strain hardening, termed “strain softening,” is most dramatic when dislocation channeling occurs at high fluxes. For moderate to highly irradiated material, Andresen [39] has suggested the use of an effective yield stress, defined as the average of the nonirradiated and irradiated yield stresses \( [\sigma_{\text{eff}} = (\sigma_{\text{yirr}} + \sigma_{\text{ynonirr}})/2]; \) this discounts the irradiation-induced increase in yield stress by a factor of 2. For highly irradiated materials Jenssen et al. [40] have proposed use of an effective stress that discounts the irradiation-induced increase in yield stress by a factor of 3.

To reduce activity and facilitate handling, small specimens (e.g., \( \approx 8\)-mm thick \( \frac{1}{4}\)-T CT) have been used in several studies. For these specimens, J values above 150 kJ/m\(^{1/2}\) and crack extensions beyond about 1.2 mm are above the validity limits based on ASTM Specification E 1820-06. However, comparison of fracture toughness data obtained on 1-T CT and small specimens indicates that small specimens yield equivalent J-R curve data at least for materials with J\(_{\text{lc}}\) values up to about 300 kJ/m\(^{1/2}\).
Plots of $J_{ic}$ or $K_{ic}$ and $K_{jc}$ as a function of neutron dose are used for developing screening criteria for neutron embrittlement. In ASTM Specification E 1820-06, $J_{ic}$ is determined from the intersection of the best-fit power-law $J$-$R$ curve with the 0.2-offset line parallel to the blunting line, provided the specimen size criterion of Eq. 3 is satisfied. The blunting line is defined as

$$J = M\sigma_f \Delta a,$$

(4)

where $\sigma_f$ is the flow stress, $\Delta a$ is crack extension, and the constraint factor $M$ is 2 or a value determined from the best fit of the experimental data. However, the analysis procedures described in the ASTM specifications for $J_{ic}$ determination are not applicable to austenitic SSs because of their extremely high toughness, ductility, and strain hardening ability. The main difference concerns the expression for the crack tip blunting line. For austenitic SSs, a value of 2 for $M$ significantly overpredicts the crack extension due to crack tip blunting; therefore, it yields a non-conservative value of $J_{ic}$ [6,41]. For austenitic SSs, a value of 4 for $M$ better defines the blunting line. The constraint factor $M$, which relates $J$ to the crack tip opening displacement (CTOD), is given by the expression

$$J = M\sigma_y(CTOD).$$

(5)

The use of a higher value for $M$ in Eq. 4 is consistent with the expected variation of $M$ and $\sigma_f$ with strain hardening. The factor $M$ is 1 for materials with intermediate to high strength and low strain hardening, and 2 for materials with low strength and high strain hardening, such as austenitic SSs. For the latter, the yield strength is approximately two-thirds of the flow stress, and the crack extension associated with blunting is approximately one-third of CTOD [6]. Thus, for such materials, the crack tip blunting line is given by

$$J = M\sigma_y(CTOD) \approx 2(2\sigma_f/3)(3\Delta a) = 4\sigma_f \Delta a,$$

(6)

i.e., Eq. 4 with $M = 4$. This relationship has been used to determine $J_{ic}$ in most investigations on neutron embrittlement [11,42]. Some investigators have used a value of 2 for $M$ [43]. The latter typically yields a higher value of $J_{ic}$ for Category III materials (i.e., $J_{ic}$ above 150 kJ/cm). However, the difference in $J_{ic}$ values determined using values of $M$ of 2 or 4 is insignificant for Category II materials (i.e., $J_{ic}$ <100 kJ/cm²). Since it is primarily the cases in which the fracture toughness of irradiated austenitic SSs is reduced to Category II levels that are of interest, the effect of differences in the procedure to determine $J_{ic}$ is likely to be insignificant.

Another factor that may influence reported values of $J_{ic}$ is when an effective yield instead of the measured yield stress is used to calculate $J_{ic}$. An effective yield stress, in which the irradiation-induced increase in yield stress is discounted by a factor of 2 or 3 has been proposed to define K/size criterion for irradiated materials. However, because $J_{ic}$ is a measure of fracture toughness at instability without significant stable crack extension, the measured yield or flow stress of the irradiated materials seems more appropriate for $J_{ic}$ determinations. Nevertheless, the choice of measured or effective values of stress is likely to have an insignificant effect on the measured $J_{ic}$ of materials with poor fracture toughness.

2.2 Effect of neutron exposure and sample orientation

The effect of neutron exposure (in dpa) on the fracture toughness $J_{ic}$ at 25-427°C (77-842°F) of austenitic SSs irradiated at 90-450°C (194-842°F) up to 90 dpa in fast reactors is shown in Fig. 1 [22-38]. The fast reactor data show a rapid decrease in fracture toughness at a neutron dose of 1-2 dpa; the neutron dose at the onset of the rapid decrease varies with the chemical composition and heat treatment of the
steel. The effects of irradiation may be divided into three regimes: little or no loss of toughness below an exposure of ≈1 dpa, substantial decrease in toughness at exposures of 1-10 dpa, and no further reduction in toughness above a saturation exposure of 10 dpa. The degradation in fracture properties saturates at a \( J_{Ic} \) value of ≈30 kJ/m\(^2\) (171 in.-lb/in.\(^2\)) [i.e., \( K_{Ic} \) of 75 MPa m\(^{1/2}\) (68.2 ksi in.\(^{1/2}\))] . Also, the failure mode changes from dimple fracture to channel fracture.

Fig. 1. Fracture toughness \( J_{Ic} \) as a function of neutron exposure for austenitic SSs irradiated in fast reactors. Solid lines represent the scatter band for the fast reactor data on austenitic SSs (Refs. 23-38).

The fracture toughness trend shown in Fig. 2 for the LWR data [5,11,42-46] is similar to that observed for fast reactor data. Most of the fracture toughness \( J_{Ic} \) values for austenitic SSs irradiated in LWRs [288-316°C (550-601°F)] fall within the scatter band of the data obtained on materials irradiated in fast reactors, even though the LWR irradiations were at lower temperatures. There are only minor

Fig. 2. Fracture toughness \( J_{Ic} \) as a function of neutron exposure for austenitic SSs irradiated in LWRs. Dashed lines represent the scatter band for the fast reactor data on austenitic SSs irradiated at 350-450°C (662-843°F) (Refs. 5,11,42,45,46).
differences in the fracture toughness of the various wrought and cast austenitic SS materials. For the same irradiation conditions, the fracture toughness of thermally aged cast SS and weld metal is lower than that of HAZ material, which, in turn, is lower than that of solution-annealed materials. A similar behavior is also observed for the fast reactor data in Fig. 1. The $J_{IC}$ values of welds and HAZ materials are consistently lower than those for the solution-annealed and even cold-worked (CW) materials. In Fig. 1, the data for CF-8 cast SS were obtained at room temperature and, therefore, are relatively high; the $J_{IC}$ values are expected to be lower at LWR operating temperatures.

Some materials irradiated above 4 dpa at LWR temperatures show very poor fracture toughness; their $J_{IC}$ values are below the lower bound curve for the fast reactor data. For Type 304 SS irradiated to 4.5-5.3 dpa (shown as cross in Fig. 2), nine out of ten CT specimens showed no ductile crack extension, and the $K_{IC}$ values were 52.5-67.5 MPa m$^{1/2}$ (47.7-61.4 ksi in$^{1/2}$) [44]. The lowest fracture toughness, with $K_{IC}$ or $J_{IC}$ values in the range of 36.8-40.3 MPa m$^{1/2}$ (33.5-36.6 ksi in$^{1/2}$), was for a Type 347 SS irradiated to 16.5 dpa in a pressurized water reactor (PWR) [44] and for a Type 304 SS irradiated to 7.4-8.4 dpa in a boiling water reactor (BWR) [46].

Another significant result is a strong orientation effect on fracture toughness. Fracture toughness J-R tests have been conducted on Type 304 control-rod and Type 304L top guide materials irradiated to 4.7-12 dpa, and on Type 304 control-rod material irradiated to 7.4 and 8.4 dpa. The results show lower fracture toughness in the T-L orientation* than in the L-T orientation [43]. Some of the toughness values in T-L orientation are lower than the limiting fracture toughness $K_{IC}$ of 55 MPa m$^{1/2}$ (50 ksi in$^{1/2}$) that has been proposed by industry for flaw tolerance evaluation in austenitic SSs irradiated above 4.5 dpa (3 x 10$^{-21}$ n/cm$^2$) [43,47]. The lower fracture toughness along the T-L orientation has been attributed to the presence of stringers consisting of long, narrow particles oriented in the rolling direction, which result in a long and narrow quasi-cleavage structure parallel to the crack advance, thereby accelerating the crack advance [43]. In addition, the Type 304 SS irradiated to 7.4-8.4 dpa showed very low fracture toughness ($J_{IC}$ of 40 kJ/m$^2$ in L-T and 7.5 kJ/m$^2$ in T-L orientation). The low $J_{IC}$ of this material was considered a special case of materials containing a high density of particles aligned in the rolling direction. Nonetheless, these results show that very low fracture toughness values are possible for irradiated austenitic SSs. Microstructural characterization of the Type 304 control-rod material showed a fine distribution of $\gamma'$ phase with particle size in the range of 2-10 nm and an average size of 4.4 nm [43]. The $\gamma'$ phase has also been observed at dose levels above 4 dpa in CW Type 316 SS irradiated under the PWR condition [48]. The contribution of additional precipitate phases, voids, and cavities on fracture toughness needs to be investigated.

2.3 Effect of material and environmental parameters

Fracture toughness J-R curve data have been obtained for the following: Type 304, 304L, and 316L SSs and their welds, including weld HAZ materials, and CF-3, CF-8, and CF-8M cast austenitic SSs irradiated in LWRs up to about 14 dpa [11,42-47]. The change in the J-R curve with neutron dose is shown in Fig. 3. The decrease in fracture toughness is quite rapid up to about 6 dpa, and the toughness decreased moderately at higher dose levels. The effects of various parameters (such as material type and heat treatment; test and irradiation temperature; and neutron energy spectrum, flux, and dose) are discussed below.

**Irradiation Facility**: Fast reactor irradiations are at fluxes and temperatures higher than those typically observed in LWRs and have a different spectrum. Until recently, most of the high neutron

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*A The first alphabet represents the direction perpendicular to the plane of the crack and second alphabet represents the direction of crack advance. L = longitudinal or rolling direction and T = transverse direction.

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exposure data were from fast reactor irradiations at temperatures above 350°C (662°F) (Fig. 1). An accurate determination of the effects of neutron spectrum, flux, and temperature on the fracture properties of these materials requires data on the same heat of material irradiated in a fast reactor and an LWR to comparable neutron dose. Such information is not available. However, although the general data trends appear to be similar for fast reactor and LWR irradiations, the tensile property data indicate that tensile strength is higher and ductility is lower for BWR-irradiated materials than materials irradiated in fast reactors [44,49,50]. The existing data are inadequate to determine the individual contributions of irradiation temperature, flux, and energy spectrum to the degradation of fracture properties in irradiated austenitic SSs. Therefore, additional fracture toughness data should ideally be obtained on the same heat of material that has been irradiated in both fast and thermal reactors to comparable fluence levels at the same temperature.

Fig. 3. Fracture toughness JIC as a function of neutron exposure for austenitic SSs (Refs. 11,46).

Material Type: Most of the J-R curve data on LWR-irradiated austenitic SSs were obtained on Type 304 and 304L SS, and data on Type 316, 316L, 316CW, and 347 SSs are very limited. Also, the only data for SS welds are on Type 308L material irradiated to <1 dpa or 12 dpa. Similarly, there are only a few J-R curve tests reported for weld HAZ materials and CF-8M cast SS irradiated to 2.1-2.5 dpa. The fracture toughness data trends appear to show differences for the various grades of austenitic SSs. For the same irradiation conditions, the fracture toughness of the weld HAZ materials is lower than that of the solution-annealed materials, and the toughness of the thermally aged cast SS is lower than that of the HAZ material (Fig. 4). However, these differences may be artifacts of the limited data.

Although the fracture toughness of nonirradiated CW steels is lower than that of nonirradiated solution-annealed steels, the decrease in toughness of CW steels with neutron exposure is slower and the JIC value at saturation is higher than that of irradiated solution-annealed steels (Fig. 1). However, the data for CW steels are from fast reactor irradiations at relatively high temperatures, 400-427°C (752-800°F). The saturation JIC for CW SSs is likely to be lower for irradiations at LWR operating temperatures [i.e., 290-320°C (554-608°F)], and the differences may be small.

Nonirradiated weld metals and thermally aged cast SSs have lower fracture toughness than wrought austenitic SSs, and their fracture toughness generally decreases more rapidly with neutron exposure than that of solution-annealed material. However, the saturation toughness for the welds is not significantly different from that of solution-annealed SSs, and the same bounding curve for JIC appears to be applicable to both wrought materials and welds and cast austenitic SSs. Although LWR core internals are typically constructed of CF-8 or CF-3 steels, the only data for LWR irradiation of cast SS are for CF-8M steel. The data for thermally aged CF-8 cast SS shown in Fig. 1 are for materials that were irradiated in the
BOR-60 fast reactor, and may be non-conservative for LWR irradiation conditions. Furthermore, the data were obtained at room temperature; as discussed later in this section, fracture toughness at higher test temperatures is expected to be lower. For thermal embrittlement of cast SSs, the fracture toughness of CF-8M steel represents the worst-case scenario [21,51]. This material thus might represent a bounding case also for the synergistic effects of neutron and thermal embrittlement.

Fig. 4.
Fracture toughness J-R curves for sensitized Type 304 SS, weld HAZ materials of Type 304 and 304L SS, and CF-8M cast SS in high-purity water at 289°C (Ref. 8).

Test Environment: Nearly all of the existing fracture toughness data were obtained from tests in air and on specimens that were fatigue precracked at relatively low load ratios (typically 0.1-0.2) in room-temperature air. However, in reactor core components cracks are initiated primarily by SCC and have intergranular (IG) morphology, whereas the fatigue precracks in fracture toughness tests are always transgranular (TG). Also, the corrosion/oxidation reaction could influence fracture toughness. For example, hydrogen generated from the oxidation reaction could diffuse into the material and change the deformation behavior by changing the stacking-fault energy of the material.

Fig. 5. Fracture toughness J-R curves for irradiated specimens of (a) Type 304 SS SMA weld HAZ and (b) Type 304L SA weld HAZ in air and BWR water environments (Ref. 8).

To investigate the possible effects of the BWR coolant environment on fracture toughness (e.g., the effect of the corrosion/oxidation reaction during crack extension or use of specimens with an IG rather than TG fatigue crack), J-R curve tests have also been conducted in BWR normal water chemistry (NWC) environment [11]. The J-R curve data on irradiated SS weld HAZ materials (Fig. 5) indicate that an BWR NWC environment has little or no effect on the fracture toughness. The J-R curves for irradiated Type 304L SA weld HAZ in air and water environments are essentially identical (Fig. 5b), and, although
the complete J-R curve could not be obtained for Type 304 SMA weld HAZ in air, ductile crack extension occurred at approximately the same value of J in air and water environments (Fig. 5a).

The J-R curves for a sensitized Type 304 SS in air and water environments are shown in Fig. 6. The results indicate slightly lower fracture toughness in water. Also, the material tested in water was sensitized for a shorter time than the material tested in air. Therefore, for materials with identical sensitization treatment, the difference between the J-R curves in air and water environments may be greater than that indicated by Fig. 6.

![J-R curves for sensitized Type 304 SS in air and water environments](image)

Fig. 6. Fracture toughness J-R curves for sensitized Type 304 SS in air and BWR water at 289°C (Ref. 8).

Figure 7 shows J-R curves and the load vs. load-line displacement curves for two tests on thermally aged and irradiated CF-8M cast SS in BWR water. Companion tests in air have not been conducted on this material. In both tests, large load drops, accompanied by crack extensions up to 0.5 mm in Specimen 75-11TM and 1.0 mm in Specimen 75-11TT, were observed at the onset of crack extension. Such load drops are not typically observed during tests in air [51]. The fracture surfaces of these specimens have not been examined to establish the fracture morphology. Additional tests on irradiated cast SSs or SS welds in air and water environments should be conducted to determine the possible effect of LWR coolant environments on their fracture toughness.

![Fracture toughness J-R curves and load vs. load-line displacement curves for thermally aged and irradiated CF-8M cast SS](image)

Fig. 7. Fracture toughness J-R curves and load vs. loadline displacement curves for thermally aged and irradiated CF-8M cast SS (Ref. 8).
Irradiation Temperature: The available data are inadequate to establish accurately the effects of the irradiation temperature on the fracture toughness of austenitic SSs. However, tensile data for austenitic SSs indicate that irradiation hardening is highest, and ductility loss is maximum, at an irradiation temperature of \( \approx 300^\circ \text{C} \) (\( \approx 572^\circ \text{F} \)) \cite{52}. Thus, the \( J_{\text{lc}} \) values for all of the data at neutron exposures greater than 20 dpa (shown in Fig. 9 below) may overestimate the toughness of materials irradiated at temperatures of 290-320°C (554-608°F) because the irradiation temperatures for these data were above 350°C (662°F).

Test Temperature: The fracture toughness of nonirradiated austenitic SSs is known to decrease as the test temperature is increased. The change in the \( J_{\text{lc}} \) of irradiated SSs as a function of test temperature is plotted in Fig. 8 for several grades of SSs and welds irradiated in LWRs and fast reactors. The fracture toughness of steels irradiated to relatively low dose (less than 5 dpa) decreases with increasing test temperature. However, for steels irradiated to more than 12 dpa, test temperature has little effect on fracture toughness. The data on materials irradiated in LWRs or fast reactors exhibit the same behavior.

![Fig. 8. Fracture toughness \( J_{\text{lc}} \) of irradiated austenitic stainless steels and welds as a function of test temperature (Ref. 11).](image)

The effect of test temperature is also reflected in the fracture morphology of highly irradiated materials. At temperatures above 230°C (446°F) the failure mode is predominantly channel fracture characterized by a faceted fracture surface. It is associated with highly localized deformation along a narrow band of slip planes whereby the initial dislocation motion along the narrow band clears away the irradiation-induced defect structure, creating a defect-free channel that offers less resistance to subsequent dislocation motion. The localization of the deformation ultimately leads to channel failure.

2.4 Fracture toughness trends and data needs

The change in initiation toughness \( J_{\text{lc}} \) of wrought austenitic SSs (including weld HAZ materials and sensitized SSs) and cast SSs and weld metals is shown in Fig. 9 as a function of neutron dose. The fracture toughness data from both fast reactor and LWR irradiations are included. The irradiation temperatures range from 90 to 427°C (194-800°F) and test temperatures vary from 250 to 427°C (212-800°F). Only the data for CF-8 cast SS irradiated in a fast reactor to 10-11 dpa (inverted triangles in the figure) were obtained at room temperature. The data in Fig. 9 show little or no effect of test temperature for materials irradiated to 12 dpa or higher, although the toughness values are already quite low above 12 dpa. Overall, the results in Fig. 9 indicate little or no change in toughness below 0.5 dpa, a rapid decrease between 1 and 5 dpa, and no further change (saturation) beyond 10 dpa.
Fig. 9. Change in initiation toughness $J_{IC}$ of (a) wrought austenitic SSs and (b) cast austenitic SSs and weld metals as a function of neutron exposure. The data plotted at 0.005 dpa are for nonirradiated materials.

There appear to be some differences in behavior between subsets of the data in Fig. 9a. The threshold dose and the dependence of the decrease in fracture toughness $J_{IC}$ on neutron dose seem to vary for different grades of materials. The average $J_{IC}$ of the Type 304 SS drops from above 150 kJ/m$^2$ (857 in.-lb/in.$^2$) at 1 dpa to 12-24 kJ/m$^2$ 69-137 in.-lb/in.$^2$) at about 5 dpa. For Type 316L SS the decrease appears to occur at a somewhat higher fluence range (3 dpa to 10 dpa), and for Type 304L SS it appears to occur at a somewhat lower fluence. With increasing fluence, the decrease in toughness is the earliest for Type 304L SS, followed by Type 304 SS and then Type 316 SS.

The fracture toughness data in Fig. 9b for cast SSs and welds are lower than those of the wrought SSs for all dose levels less than the 10-dpa saturation level. However, the available data for irradiated SS welds and cast austenitic SSs are extremely limited. There are no data on any of these materials for fluences above 20 dpa, and little or no data on cast austenitic SSs for fluences below about 2 dpa. The existing data for welds suggest that $\approx 0.3$ dpa may be considered a threshold neutron dose below which irradiation has little or no effect on fracture toughness of SS welds. However, this threshold does not consider the possible synergistic effects of thermal and neutron embrittlement of welds.

The two curves shown in Fig. 9 represent a disposition curve proposed by EPRI [43] and a fracture toughness trend curve developed in the present study, which bounds the existing data. The trend curve takes into consideration: (a) a threshold neutron exposure for radiation embrittlement of austenitic SSs and a minimum fracture toughness for these materials irradiated to less than the threshold value, (b) a
saturation neutron exposure and a saturation fracture toughness for materials irradiated to greater than this value, and (c) a description of the change in fracture toughness between the threshold and saturation neutron exposures. As shown in Fig. 9, the fracture toughness $J_{lc}$ curve that bounds the existing data for $J_{lc}$ as a function of neutron dose (in dpa) may be represented by

$$J_{lc} = 7.5 + 110 \exp[-0.35(\text{dpa})^{1.4}].$$

This lower bound curve represents a threshold dose of 0.3-0.5 dpa for neutron embrittlement, a minimum fracture toughness $J_{lc}$ of $\approx 118$ kJ/m² below the threshold dose, a saturation threshold of 5 dpa beyond which the fracture toughness of these materials appears to saturate, a saturation fracture toughness $J_{lc}$ of 7.5 kJ/m² (or $K_{lc}$ or $K_{jc}$ of 38 MPa m¹/²), and a description of the change in toughness between 0.3 and 5 dpa. The value of $J_{lc}$ of $\approx 118$ kJ/m² below the threshold dose is appropriate for thermally aged and unaged cast SSs and SS flux welds. A value higher than $118$ kJ/m² may be considered for the minimum fracture toughness $J_{lc}$ for wrought austenitic SSs irradiated below the threshold dose for neutron embrittlement. The description of the change in fracture toughness below 1.5 dpa will change accordingly. The lower bound trend curve given by Eq. 7 is consistent with the Materials Reliability Program (MRP) model proposed for PWRs [45]. The MRP model is expressed in terms of a lower bound $K_{jc}$ (MPa m¹/²) curve. It bounds all the fracture toughness data from fast reactors, BWRs, and PWRs as a function of the neutron dose (in dpa) and is given by the expression,

$$K_{jc} = 180 - 142[1-\exp(-\text{dpa})].$$

Both Eqs. 7 and 8 predict a saturation fracture toughness $K_{lc}$ of 38 MPa m¹/². For materials irradiated below the threshold dose for irradiation embrittlement, Eq. 8 predicts a minimum $K_{jc}$ of 151 MPa m¹/², but the MRP expression predicts fracture toughness values that for some materials, such as SS welds or weld HAZ, may be higher than the minimum toughness of the materials in the nonirradiated condition.

The disposition curve proposed by EPRI for BWRs is also not bounding for the existing data for BWR-irradiated austenitic SSs. For example, at neutron doses $<0.7$ dpa the $J_{lc}$ values based on the EPRI curve are higher than the minimum $J_{lc}$ of nonirradiated SS welds (particularly flux welds), some heats of wrought SSs, and most thermally aged cast austenitic SSs with $>15\%$ ferrite [51]. The saturation $K_{lc}$ of 55 MPa m¹/² at 4.5 dpa for the EPRI curve is also higher than the value of 38 MPa m¹/² previously proposed by MRP for PWRs [45]. The saturation $K_{lc}$ for the EPRI curve was based on data for which the specimen orientation was unknown. Recent data indicate that fracture toughness in the transverse orientation is nearly half of that in the longitudinal orientation [43]. Therefore, the bounding $K_{lc}$ values above 4.5 dpa are likely to be lower than 55 MPa m¹/².

The existing fracture toughness database indicates that there are limited data for austenitic SSs, in particular for SS welds and cast austenitic SSs, irradiated in LWRs to neutron dose levels of 0.1-1.0 dpa or above 10 dpa. Such data are needed to confirm the threshold dose for significant decrease in the fracture toughness of austenitic SSs and the saturation fracture toughness of these materials inferred from the larger data set including materials from the fast reactors. Also, the potential synergistic interactions of thermal aging and neutron irradiation embrittlement of cast austenitic SSs and SS welds [11,51] have yet to be addressed.

The fracture toughness J-R curve may be used to analyze material behavior for loading beyond $J_{lc}$. The J-R curve is expressed in terms of the J integral and crack extension ($\Delta a$) by the power law $J = C(\Delta a)^p$. At dose levels below the threshold dose for saturation (i.e., at dose levels less than $\approx 5$ dpa), the effect of neutron irradiation on the fracture toughness of austenitic SSs can also be represented by a
A decrease in the coefficient $C$ of the power-law correlation for the J-R curve with neutron dose. The variation of $C$ for wrought and cast SSs and welds as a function of neutron dose is shown in Fig. 10. Except for the results for CF-3 (closed diamond in Fig. 10b) and CF-8 irradiated in BOR-60 reactor (open inverted triangle in Fig. 10b), the remaining data were obtained at temperatures above 250°C. Based on the data trends in Fig. 8, test temperature should have little or no effect on the data for CF-8 cast SS. However, the constant $C$ for CF-3 cast SS may be more than a factor of two lower at LWR operating temperatures. The two curves in Fig. 10 represent a disposition curve proposed by EPRI for BWRs [47] and a bounding curve for coefficient $C$ developed in the present study (Eq. 9 below).

![Graph](image)

**Fig. 10.** Coefficient $C$ of the power-law J-R curve for (a) wrought austenitic SSs and (b) cast austenitic SSs and weld metals as a function of neutron exposure.

Even for fluence levels above 10 dpa, most heats of wrought austenitic SSs (Fig. 9a) show ductile crack extension in the toughness tests. Under similar irradiation conditions, the coefficient $C$ for cast SSs and welds (Fig. 9b) is lower than that for wrought SSs. However, since most of the data are from irradiations in fast reactors and at temperatures of 370–427°C (698–800°F), the values of power-law coefficient $C$ may be lower for irradiations at LWR operating temperatures. As mentioned previously, fracture toughness data are very limited on materials irradiated in LWRs to neutron dose levels of 0.1-1.0 dpa or above 10 dpa. Therefore, it is not possible to define accurately the lower bound trend curve for the power-law coefficient $C$ as a function of neutron dose. For fluences less than 5 dpa, as shown in Fig. 10, the existing fracture toughness data can be bounded by the following expression for $C$:

$$C = 25 + 175 \exp[-0.35(dpa)^{1.4}],$$

(9)
an exponent n in the power law equal to 0.37 (the median value of the experimental data). The exponent n of the power-law curve typically ranges from 0.35 to 0.70 for nonirradiated materials and 0.16 to 0.65 for irradiated materials. Unlike the behavior for thermally aged cast austenitic SSs where exponent n typically decreases with a decrease in fracture toughness [51], no obvious trend of n with fluence is evident. For irradiated materials, the median value of n is 0.37. Equation 9 yields a C value of \( \approx 200 \text{ kJ/m}^2 \) (1285 in.-lb/in.\(^2\)) for materials irradiated to less than 0.1 dpa and \( \approx 31 \text{ kJ/m}^2 \) (\( \approx 160 \) in.-lb/in.\(^2\)) for materials irradiated to \( \approx 5 \) dpa. These values would yield \( J_{IC} \) of 125 and 17 kJ/m\(^2\), respectively, for materials irradiated to \(<0.1\) and \(5\) dpa. These values are consistent with the \( J_{IC} \) trend curve of Fig. 9. The \( J_{IC} \) at 5 dpa is also consistent with the data for the CT specimens of Type 304 SS irradiated to \( \approx 4.5 \) dpa in a BWR (closed triangles in Fig. 9a).

As noted previously, ductile crack extension was also not observed for some 20% CW Type 316 SS specimens irradiated to 74-88 dpa in a fast reactor at 410-425°C (770-797°F). The specimens failed by a quasi-cleavage fracture believed to be an indirect consequence of the onset of void swelling in the material. The \( K_{IC} \) values were 74-90 MPa m\(^{1/2}\) (67-82 ksi in.\(^{1/2}\)).

2.5 Synergistic effects of thermal and neutron irradiation

It is well known that thermal aging of cast austenitic SSs and SS welds at reactor operating temperatures leads to degradation of their fracture properties [21,51, 56-58]. An issue that has been a concern for reactor core internal components is the possibility of a synergistic interaction between irradiation embrittlement and thermal embrittlement of cast austenitic SSs and SS weld metals. Although wrought SSs are typically completely austenitic, welded and cast SSs have a duplex microstructure consisting of austenite and ferrite phases. The ferrite phase increases the tensile strength and improves resistance to SCC, but it is susceptible to thermal embrittlement after extended service at reactor operating temperatures. Thermal aging of cast SSs at 250-400°C (482-752°F) leads to precipitation of additional phases in the ferrite (e.g., formation of Cr-rich \( \alpha' \) phase by spinodal decomposition; nucleation and growth of \( \alpha' \); precipitation of a Ni- and Si-rich G phase, \( M_{23}C_6 \) carbide, and \( \gamma_2 \) austenite; and additional precipitation and/or growth of existing carbides at the ferrite/austenite phase boundaries) [53-55]. The formation of the Cr-rich \( \alpha' \) phase by spinodal decomposition of ferrite is the primary mechanism for thermal embrittlement; it strengthens the ferrite phase by increasing strain hardening and the local tensile stress. Thermal aging has little or no effect on the austenite phase. Thus, thermal aging of cast SSs and SS welds leads to the development of a material with a brittle phase dispersed in a ductile matrix.

Embrittlement of the ferrite phase due to neutron irradiation occurs much faster than for austenitic SSs. At reactor operating temperatures of 288-343°C (550-650°F), a shift in the nil-ductility transition (NDT) temperature of up to 150°C (302°F) has been observed in pressure vessel steels after neutron exposures of 0.07-0.15 dpa (0.5-1.0 x 10\(^{20}\) n/cm\(^2\)) [59]. The irradiation temperature is an important factor in establishing the extent of embrittlement of ferritic steels. Although both the thermal aging embrittlement and the neutron irradiation embrittlement of ferrite are well characterized, the synergistic effect of thermal aging and neutron irradiation on the embrittlement of SS welds and cast SSs has not been investigated yet. The concurrent exposure to high neutron fluence levels could result in a synergistic effect wherein the service-degraded fracture toughness is reduced from the levels predicted independently for either of the two mechanisms.

The effect of neutron irradiation on the fracture toughness of wrought and cast SSs and welds has been discussed in the previous sections, and the change in fracture toughness \( J_{IC} \) and coefficient C of the power-law J-R curve with neutron dose is shown in Figs. 9 and 10, respectively.
Thermal aging increases the tensile strength, hardness, and Charpy-impact transition temperature, and it decreases the ductility, fracture toughness, and impact strength. For cast austenitic SSs, the extent of mechanical-property degradation is essentially determined by the chemical composition of the steel, the casting process used to construct the component, the ferrite content and ferrite morphology of the steel, and the time and temperature of service for the component [51]. Cast SSs with high levels of Mo (e.g., CF-8M) show greater susceptibility to thermal embrittlement than steels with low Mo content (e.g., CF-3 or CF-8). Static cast steels are more susceptible to thermal embrittlement than centrifugally cast components. The screening criteria to determine the susceptibility of cast SS components to thermal aging embrittlement are outlined in Table 1 [60].

Table 1. Screening criteria for thermal-aging susceptibility of cast austenitic stainless steels.

<table>
<thead>
<tr>
<th>Mo Content (wt.%)</th>
<th>Casting Method</th>
<th>Ferrite Content</th>
<th>Susceptibility Determination</th>
</tr>
</thead>
<tbody>
<tr>
<td>High (2.0-3.0)</td>
<td>Static</td>
<td>≤14%</td>
<td>Not susceptible</td>
</tr>
<tr>
<td></td>
<td></td>
<td>&gt;14%</td>
<td>Potentially susceptible</td>
</tr>
<tr>
<td></td>
<td>Centrifugal</td>
<td>≤20%</td>
<td>Not susceptible</td>
</tr>
<tr>
<td></td>
<td></td>
<td>&gt;20%</td>
<td>Potentially susceptible</td>
</tr>
<tr>
<td>Low (0.5 max.)</td>
<td>Static</td>
<td>≤20%</td>
<td>Not susceptible</td>
</tr>
<tr>
<td></td>
<td></td>
<td>&gt;20%</td>
<td>Potentially susceptible</td>
</tr>
<tr>
<td></td>
<td>Centrifugal</td>
<td>All</td>
<td>Not susceptible</td>
</tr>
</tbody>
</table>

For cast austenitic SSs, the minimum fracture toughness that can occur due to thermal embrittlement depends strongly on the ferrite content and morphology. A globular ferrite morphology in which the brittle ferrite phase is isolated in an austenitic matrix will have a higher fracture toughness than a lacy morphology in a material with greater than 9% ferrite, where a continuous fracture path through the brittle ferrite is possible. The minimum toughness due to thermal aging occurs when the ferrite is fully embrittled, and the remaining toughness depends on the toughness provided by the ductile matrix surrounding the embrittled phase. Based on a previous study [51], the predicted saturation J-R curves for the various cast SSs in the thermally aged condition (i.e., the lowest fracture toughness that could be achieved for the steel after thermal aging) are expressed as $J \approx 264 \Delta a^{0.35}$, $\approx 251 \Delta a^{0.34}$, and $\approx 167 \Delta a^{0.31}$, respectively, for CF-3, CF-8, and CF-8M steels at 290°C (554°F).

The results for SS welds indicate that the decrease in fracture toughness due to aging depends on the ferrite content and initial toughness of the weld [21]. Differences in the fracture toughness of SS welds arise from differences in the density and size of inclusions in the material. Failure occurs by the formation and growth of microvoids near hard inclusions. Welds with relatively high fracture toughness (e.g., gas tungsten arc or tungsten inert gas weld) show a significant decrease due to thermal aging, whereas welds with poor fracture toughness (e.g., submerged arc, shielded metal arc, or manual metal arc welds) show minimal change. In the latter case, failure primarily occurs by the formation and growth of microvoids. Such processes are relatively insensitive to thermal aging. The existing data indicate that at 280-350°C, the fracture toughness $J_{lc}$ of thermally aged welds can be as low as 40 kJ/m². A conservative estimate of the J-R curve for aged SS welds [21] is given by $J = 40 + 83.5 \Delta a^{0.643}$.

Reactor core internal components are subject to thermal and concurrent exposure to neutron irradiation. This could result in a synergistic effect wherein the service-degraded fracture toughness can be less than that predicted for either thermal aging embrittlement or neutron irradiation embrittlement independently.

For license renewal, to account for the effects of thermal aging and neutron irradiation embrittlement on the fracture toughness of reactor core internal components, the NRC staff has proposed that for cast austenitic SS components that have a fluence of greater than $1 \times 10^{17}$ n/cm² (0.00015 dpa) or
are determined to be susceptible to thermal aging embrittlement, an aging management program should be implemented consisting of either a supplemental examination of the affected components as part of the applicant’s 10-year in-service inspection program during the license renewal term, or a component-specific evaluation to determine the susceptibility to loss of fracture toughness [60]. Furthermore, the program should provide for the consideration of the synergistic loss of fracture toughness due to neutron irradiation embrittlement and thermal aging embrittlement.

An EPRI report on thermal aging embrittlement of cast SS components proposed the use of the J value at a crack extension of 2.5 mm (0.1 in.), \( J_{2.5} \), to differentiate between nonsignificant and potentially significant reductions in fracture toughness of cast austenitic SSs [61]. Flaw tolerance evaluations were presented in Appendices A and B of the EPRI report to support the choice of a threshold value of \( J_{2.5} = 255 \text{ kJ/m}^2 \) (1456 in.-lb/in.\(^2\)). The NRC staff has found that using \( J_{2.5} = 255 \text{ kJ/m}^2 \) is an acceptable screening approach for fracture toughness of cast SSs [60]. This concept can be extended to irradiated materials. However, the applicability of the flaw tolerance evaluations in Appendices A and B of the EPRI report would have to be demonstrated to support the use of the \( J_{2.5} \) parameter for evaluating the toughness of irradiated SSs.

For the wrought and cast austenitic SSs and welds listed in Fig. 10, the experimental J-integral values at a crack extension of 2.5 mm are plotted as a function of neutron dose in Fig. 11. The solid curve in Fig. 11 represents the predicted values of \( J \) at 2.5-mm crack extension that are expected to bound the

![Fig. 11. Experimental values of J-integral at a crack extension of 2.5 mm for (a) wrought austenitic SSs and (b) cast austenitic SSs and weld metals plotted as a function of neutron exposure.](image-url)
existing experimental data shown in Fig. 10. The curve was obtained using the power-law J-R curve relationship, coefficient C determined from Eq. 9, and exponent n of 0.37. The lower bound curve indicates that for cast SSs and welds irradiated up to 1.0 dpa, the predicted J at 2.5 mm is above the screening value of 255 kJ/m² (1456 in.-lb/in.²). However, additional fracture toughness data on irradiated SS welds and cast SSs, particularly at 0.1-2.0 dpa, are needed to better define the threshold dose when the fracture toughness of austenitic SSs begins to significantly decrease.

This evaluation does not consider the synergistic interaction of neutron irradiation and thermal aging embrittlement. Embrittlement of ferrite phase from neutron irradiation occurs at lower dose levels than does embrittlement of the austenite phase. A shift in the NDT temperature of up to 150°C (302°F) has been observed in pressure vessel steels irradiated to 0.07-0.15 dpa [59]. Thus, embrittlement of ferrite is expected to occur at 0.05-1.0 dpa, whereas any significant effect of neutron irradiation on embrittlement of the austenite phase occurs only above ~0.5 dpa (Fig. 9).

In addition to possibly altering the threshold dose for neutron embrittlement, synergistic effects of neutron irradiation and thermal aging embrittlement could decrease the saturation fracture toughness of irradiated welded and cast SSs and accelerate the change in fracture toughness between the threshold and saturation neutron exposures. Unfortunately, existing data are inadequate for an accurate assessment of such effects. Figure 11 shows the results of two tests on a CF-8M steel that was thermally aged for 10,000 h at 400°C and then irradiated to well above the threshold dose for neutron embrittlement (inverted triangles in Fig. 11). The resulting toughness is bounded by the curve for other SSs irradiated to a similar level, i.e., thermal aging does not seem to lower the toughness below that expected for irradiation alone at these neutron dose levels. Additional fracture toughness data are needed to better establish the potential for synergistic loss of toughness in these materials in the transition dose range from 0.05 to 2 dpa.

3. Void swelling

Void swelling refers to the volume change of materials under neutron irradiations. Void swelling was first observed in the late 1960s in austenitic SSs irradiated in fast reactors to neutron exposures above 1 x 10²³ n/cm² at temperatures of 370-650°C (698-1202°F) [12]. As discussed above, neutron irradiation damages materials by displacing atoms from their lattice position. Such displacements create vacancies and interstitials, most of which are annihilated by recombination. The surviving defects lead to microstructural changes as they rearrange into more stable configurations. Because of the relatively large strain field that surrounds interstitials, there is a strong interaction between interstitials and dislocations, which results in a preferential flux of interstitials toward dislocations. The remaining vacancies cannot annihilate by recombining with interstitials, and this condition leads to the nucleation of cavities or microvoids. Under certain conditions of temperature and dose rate, these cavities or microvoids can eventually grow to larger sizes. Fission products such as He and H also play an important role in void formation. By combining with these gas atoms, void nucleation is facilitated through reduction of the surface energy of vacancy clusters. The fundamental driving force of void formation, however, remains the excess vacancy flux toward voids.

The formation of a large number of voids results in an increase in the volume of the material, a process referred to as “void swelling.” The volume changes from void swelling are normally isotropic and occur in all directions. However, there are always constraints on swelling, both internally due to gradients of temperature and dose rate within the component, and externally from neighboring components. Such constraints result in anisotropic swelling and stress fields within the components that
activate irradiation creep in unconstrained directions, producing dimensional distortions and misfits that can compromise the functional integrity of the reactor core internal components.

Most void swelling data have been obtained from materials irradiated in fast breeder reactors at temperatures above 385°C (725°F) and at dose rates that are orders of magnitude higher than those in PWRs, and extrapolation of these results to estimate the void swelling behavior for PWR end-of-life or extended life conditions introduces substantial uncertainties.

3.1 Effect of material and environmental parameters on void swelling

The void swelling process is divided into two regimes: a transient regime followed by a steady-state swelling rate. All materials regardless of material composition and thermo-mechanical condition, stress state, neutron flux and spectrum, or irradiation temperature (above 300°C) are believed to reach a steady-state swelling rate of ~1%/dpa [12]. However, the duration of the transient regime varies with the material composition, stress, and irradiation parameters. In general, higher neutron flux and stress accelerate, and cold work in the material delays, the onset of the steady-state swelling rate. Significant effects of various material and environmental parameters on void swelling are as follows:

**Irradiation temperature:** In the temperature range for fast reactors, void swelling shows a strong dependence on irradiation temperature. For austenitic SSs a steady-state swelling rate of 1%/dpa is observed in fast reactor data at 427°C (800°F) or higher temperatures. Voids are typically observed in austenitic SSs at temperatures above 340°C. While PWR coolant temperatures do not exceed 345°C (653°F), gamma heating of thick section components near the reactor core could increase the local temperature within the component to 370°C (698°F) or even higher. The lower irradiation temperatures typical of LWRs extend the transient regime to much longer times and result in the formation of a higher density of smaller voids. However, depending on the irradiation temperature, dose, and dose rate, voids can form under PWR operating conditions within the reactor lifetime. Voids have been observed in austenitic SSs irradiated at temperatures as low as 300°C (572°F) [62].

**Material composition:** Reactor core internal components are primarily constructed of austenitic SSs. These materials have a face-centered cubic crystal structure and are more susceptible to void swelling (i.e., swelling occurs earlier and at a higher rate) than ferritic steels, which have a body-centered cubic crystal structure. The material composition can influence the transient regime for void swelling. The most important factor is the Ni content in the steel. In the 300-series SSs, swelling decreases rapidly with increasing Ni content up to about 30-40% [63]. In contrast, Cr has the opposite effect; it decreases the transient regime and increases swelling [63]. However, at low temperatures typical of LWRs, swelling is not expected to change significantly over the range of Cr found in the 300-series SSs used for nuclear service. Because of compositional differences, Type 304 SS typically swells more than Type 316 SS under similar thermal-mechanical conditions.

The effect of minor trace elements on void swelling is sensitive to material and environmental conditions and, therefore, is less predictable. Similar to the effect of Ni on swelling, P contents above 0.01 wt.% and Si contents above 0.2 wt.% decrease swelling in the 300-series SSs [64,65]. The effect of these solutes at small concentrations may be different at PWR temperatures. However, since most SSs used for the construction of LWR core internals typically contain 0.03-0.04 wt.% P and ~0.5 wt.% Si, these elements are expected to decrease void swelling. Both change the effective diffusivity of vacancies and thereby the vacancy supersaturation. Minor differences in the concentration of these minor elements and their distribution due to differences in thermal-mechanical history may explain the large differences in the swelling behavior sometimes observed in essentially identical heats of SSs [66].
Phase changes during irradiation: Irradiation at higher temperatures, particularly above 340°C, leads to the formation of second phase particles. The radiation-induced formation of γ' silicide (Ni$_3$Si), phosphides (M$_2$P and M$_3$P), and G phase (M$_6$Ni$_{16}$Si$_7$) can remove Ni, P, and Si from the alloy matrix, and thereby increase void swelling. However, radiation-induced precipitation of second phases is strongly dependent on both temperature and displacement rate [12].

Material condition: Cold work in the material prolongs the transient regime for void swelling in austenitic SSs. The high dislocation density produced by the cold work provides additional recombination sites for vacancies and interstitials, thereby decreasing supersaturation of vacancies in the material and delaying void nucleation. The high dislocation density also interferes with diffusion of minor elements and, therefore, delays or prevents the formation of second phase particles, which remove elements (such as Ni, Si, or P) that are known to suppress void swelling. However, the benefit of cold work is short lived because absorption of vacancies increases the mobility of the dislocations, permitting them to interact and decline in density to levels similar to those achieved in solution-annealed materials with comparable irradiation doses. In solution-annealed Type 304 SS, voids nucleate easily, and the transient regime is characterized by a slowly increasing rate of swelling with increasing neutron dose. By contrast, in CW Type 316 SS void nucleation is difficult or does not occur for some time, but the transient regime ends, and the steady-state swelling rate starts abruptly [13].

Stress and stress history: In general, stress accelerates void swelling and decreases the transient regime. The sign of the stress state, tensile or compressive, does not matter; it is the shear component and not the hydrostatic component that accelerates swelling [12]. As discussed above, constraints due to gradients in temperature and/or dose rate and from adjoining components lead to anisotropic swelling and build-up of stresses within the component. These stresses, however, are limited by irradiation-induced creep and will relax in time; the rate of relaxation is proportional to the swelling rate. Therefore, the swelling-induced stresses often vary with time, increasing due to swelling and component constraints, and decreasing with creep relaxation. The interaction of swelling and irradiation creep relaxation plays an important role in evaluation of swelling in PWR core internal components. The deformation due to irradiation creep is in the direction that relieves the shear component of the applied stress. In a bolt, the primary stress is tensile along the bolt axis. Thus, the bolt would creep to increase its length and decrease the diameter. However, if the plate in which the bolt is embedded swells at a greater rate, the stress in the bolt would be reestablished. In the case of a hot spot in a former plate behind a reentrant baffle plate corner, the local stresses that were generated because of external and internal constraints would redirect swelling in the vertical direction [66]. Another effect of stress is that a high swelling region within a component would produce stresses in an adjacent lower swelling region, thereby increasing the swelling rate. Therefore, estimates of swelling based on stress-free material would tend to overestimate swelling gradients in a component.

Displacement Rate: Void swelling data on annealed Type 304L SS irradiated to ~30 dpa at 390°C in fast reactors indicate that swelling increases as the displacement rate is decreased [67]. This suggests that estimates of void swelling based on high flux fast reactor data could yield nonconservative results for PWR core internal components. In other words, under comparable irradiation temperatures and neutron dose levels, the lower displacement rates typical of PWRs could result in higher swelling. However, the exact mechanism by which displacement rate affects swelling is not well understood. Initially, it was believed to be due to an upward shift in the temperature range for void swelling with increases in displacement rate [68]. Such an argument explains the change in swelling in terms of a change in void growth rate rather than a change in void nucleation. However, several studies on austenitic SSs irradiated in EBR-II at 373-444°C (703-831°F) indicate that the effect of displacement rate on swelling is due to its
effect on the transient regime and not void growth [69]. The data also indicate that the displacement rate effect in solution-annealed materials is similar to that in cold-worked materials.

**Helium production:** Under neutron irradiation, He can be produced by the $^{10}$B reaction to form $^4$He and $^7$Li, and also from the two-step production of $^{58}$Ni from $^{58}$Ni by thermal neutrons [i.e., $^{58}$Ni($n,\gamma$)$^{59}$Ni($n,\alpha$)$^{56}$Fe]. In the latter, the He generation rate is proportional to the $^{59}$Ni content and the thermal neutron flux. In PWRs, transmutations can produce both He and H. The nucleation of microvoids due to migration and condensation of vacancies produced by neutron displacements can be stabilized by the transmuted gas atoms (He or H), thereby increasing void swelling. A fine dispersion of He bubbles has been observed in SSs irradiated in LWRs at 300-340°C, especially in reactors with high thermal flux [70,71]. However, the effect of He on swelling is not as strong as that of Ni or P contents in the steel. Most of the earlier studies have focused on the effects of He on swelling. Studies on the effects of H are limited.

### 3.2 Assessment of void swelling in PWR core internal components

The void swelling measured in PWR internal components, such as the flux thimble tubes and baffle bolts [15,72,73], is plotted as a function of neutron dose in Fig. 12. Except for two data points that show 0.2-0.25% swelling, void swelling in these SSs irradiated up to 80 dpa is insignificant. However, the irradiation temperature for the low-swelling materials was less than 330°C (626°F). A review of the existing void swelling data on austenitic SSs irradiated in fast reactors and the limited data available on PWR-irradiated materials indicated that they all can reach the steady-state swelling rate of ~1%/s. The duration of the transient regime to reach this steady-state rate depends on material composition and thermal-mechanical condition, stress state, irradiation temperature, and neutron displacement rate. The state of knowledge on the effects of these variables is discussed next.

Fig. 12. Void swelling of PWR materials plotted as a function of neutron dose (Refs. 15,72,73).

**Temperature:** Although PWR coolant temperatures do not exceed 345°C (653°F), gamma heating can increase the local temperature in thick sections of reactor core components to values as high as 420°C (788°F). However, most of the void swelling data on PWR-irradiated materials are at temperatures below 330°C, where swelling is minimal. There are no swelling data at 370-420°C (698-788°F), where swelling is expected to be the highest. Estimates of void swelling in PWR components are based on materials irradiated in fast reactors at temperatures above 370°C (698°F).
**Displacement rate:** Typical displacement rates in PWRs are 0.3-9.0 x 10^{-8} dpa/s, whereas rates used in the void swelling studies are 0.3-10.0 x10^{-7} dpa/s. Void swelling data on annealed and CW austenitic SSs irradiated in fast reactors at 373-444°C (703-831°F) indicate that swelling increases as the displacement rate is decreased. The displacement rate effect is primarily due to its effect on the transient regime. In other words, the dose required to achieve the 1%/dpa steady-state swelling rate is shorter for austenitic SSs irradiated at lower dose rate. These results suggest that under comparable irradiation temperatures and neutron dose levels, the lower displacement rates typical of PWRs would result in higher swelling. Limited void swelling data on PWR-irradiated materials are consistent with this prediction, which is based on fast reactor data at temperatures relevant for PWR internals and at displacement rates below 5.0 x 10^{-8} dpa/s.

**Stress State:** In general, applied stress or swelling-induced stress accelerates void swelling. Secondary stresses due to thermal gradients and geometric constraints, however, are limited by irradiation creep and relax rather quickly. Therefore, the interaction of swelling and irradiation creep relaxation plays an important role in evaluation of swelling in PWR core internal components. Also, another stress-related effect is that stresses produced by a high swelling region in an adjacent lower swelling region, force them to try to catch up with the faster growing regions. Thus, estimates of swelling based on stress-free material would tend to overestimate swelling gradients in a component.

**Material composition and condition:** The data on LWR-irradiated materials are inadequate to establish the differences in the swelling behavior of austenitic SSs under LWR and fast reactor irradiation conditions. The material composition can influence the transient regime for void swelling. In general, an increase in Ni, P, or Si content decreases swelling by increasing the transient regime. In contrast, Cr has an opposite effect. However, under PWR irradiation conditions, swelling is not expected to change significantly with minor differences in the Cr content. Cold work also increases the transient regime for swelling, but does not affect the steady-state swelling rate. Such differences between LWR and fast reactor irradiations need to be investigated.

**Phase change:** Radiation-induced formation of second phases can remove elements such as Ni, P, and Si, which are known to decrease swelling, from the alloy matrix. This effect increases void swelling. The precipitation of second phases is strongly dependent on material composition and thermal-mechanical condition as well as temperature, dose, and dose rate. However, another aspect of precipitation of second phase particles may tend to decrease swelling. Typically, irradiation-induced precipitation at PWR operating temperatures is extremely small (i.e., a few to several tens of nanometers), and the precipitate density is large. Therefore, at low temperatures, precipitation creates very large internal surfaces (i.e., interfaces between the precipitates and the metal matrix), which would act as sinks for vacancies, thereby suppressing the accumulation and condensation of vacancies to form voids. This behavior has been well known to occur in vanadium alloys [15]. The role of precipitation on void swelling in austenitic SSs under PWR irradiation conditions needs to be investigated.

Most of the void swelling data on LWR-irradiated materials have been obtained on CW Type 316 SS, and data on solution-annealed Type 304 SS are limited. Because of the higher Ni content, void swelling in Type 316 SS is less than that in Type 304 SS. Also, cold work further reduces swelling in CW Type 316 SS. Additional data are needed on PWR-irradiated annealed Type 304 SS. Also, it is important to develop a technical basis for extrapolating the fast reactor data and the limited PWR data to predict void swelling in PWR core internals. Although extrapolation of these results suggests that void swelling will not be a significant problem during the first license extension period, investigations are still needed on the effects of irradiation temperature (particularly above 370°C), displacement rate, and stress on void swelling, including the interaction of swelling and irradiation creep relaxation.
Stress-free swelling equations have been developed by the MRP for annealed Type 304 and CW Type 316 SSs used in PWR core internal components [74]. The swelling rate SR (%/dpa) is expressed in terms of the temperature T (°C), neutron dose \( \phi \) (dpa), and dose rate \( \dot{\phi} \) (10\(^{-7}\) dpa/s). The swelling rate for annealed Type 304 SS is given by

\[
SR_{SA304} = 2\phi^{0.731} \exp \left( 22.106 - \frac{18558}{T + 273.15} \right),
\]

and the swelling rate for CW Type 316 SS by

\[
SR_{CW316} = 0.9 \left[ 10 + 2\phi \left( 1 - \exp\left( -0.01\phi \right) \right) \right] \phi^{0.731} \exp \left( 22.106 - \frac{18558}{T + 273.15} \right).
\]

Equation 11 predicts a slightly higher swelling rate for Type 316 SS than that for annealed Type 304 SS up to 10 dpa and a lower swelling rate for higher dose levels. The predicted void swelling for Type 304 SS core barrel and former plate and for CW Type 316 SS core barrel-to-former bolt for a 60-y life at 90% effective full power years (EFPY) is shown in Fig. 13. Note that the effect of temperature seems to be much stronger than that of dose rate. For the core barrel, although the dose rate is low, the 60-y end-of-life void swelling is quite low because the total dose is only 12.4 dpa.

An expression to correlate stress-enhanced swelling has also been developed [74]. The incremental stress-enhanced swelling, \( \Delta S \) (%), is given by

\[
\Delta S = S'\Delta\phi \left( 1 + 0.005\bar{\sigma} \right),
\]

where \( S' \) is stress-free swelling rate \( SR_{SA304} \) or \( SR_{CW316} \), \( \Delta\phi \) is the increment of neutron dose, and \( \bar{\sigma} \) is the von Mises effective stress (MPa). As discussed previously, any stress state, negative or positive, will accelerate the swelling rate. These expressions are based on isothermal irradiation, and the maximum swelling rate is limited to 1%/s.

![Fig. 13. Void swelling of PWR materials plotted as a function of effective full power years (Refs. 15,72,73).](image)
3.3 Embrittlement due to void swelling

As the volume of voids in the material continues to increase beyond 3-5%, the size and distribution of voids essentially control the fracture properties of the material. Several studies have shown that austenitic SSs irradiated at ≤400°C to neutron dose levels that produce ≥10% void swelling suffer from severe embrittlement [73-77]. Furthermore, the critical swelling level required to induce an extremely brittle state decreases with decreasing irradiation temperature [75]. A Type 316 SS irradiated in the EBR-II fast reactor at 400°C to 130 dpa produced 14% void swelling, and fractured during handling at room temperature [76]. Also, a Russian SS, EI-847, irradiated in the BN-350 fast reactor in the annealed condition to 73 dpa at 335°C and in the CW condition to 82 dpa at 365°C, had more than 10% void swelling and was found to be exceptionally brittle [77].

In austenitic SSs with ≥10% void swelling, the high embrittlement has been attributed to a process that involves stress concentration between voids, Ni segregation at void surfaces, and a tendency toward martensite formation when the steel is deformed at room temperature [75]. As discussed previously, the removal of Ni and Si from the metal matrix to form γ' or G phase promotes void nucleation and swelling. Once voids form, Ni segregates to void surfaces due to the inverse Kirkendall effect, in which the slowest diffusing elements segregate. This further depletes Ni from the matrix. The loss of Ni and relative increase of Cr in the matrix in the region between the voids decrease the stacking fault energy (SFE) substantially, which promotes planar slip and formation of martensite during room-temperature deformation. The austenite/martensite boundaries provide a low-energy crack propagation path, resulting in a very brittle, quasi-cleavage fracture at room temperature [76,77]. At high temperatures, the effect of Ni segregation on SFE is much less pronounced, and large levels of flow localization are required for failure, which occurs along the flow localization path and is referred to as “channel fracture” [34].

There are few quantitative data correlating the void swelling with fracture toughness of the material. A Fe-14Cr-16Ni-1.5Mo SS (D9) specimen irradiated in the Fast Flux Test Facility (FFTF) at 390°C (734°F) to 114 dpa showed a $K_{Ic}$ of about 30 MPa m$^{1/2}$ at 410°C and zero tearing modulus [78]. Although the void swelling was not measured for this material, it is expected to be above 10%. Swelling in the same material irradiated at 410°C to 175 dpa showed about 30% void swelling, and 20% CW Type 316 SS irradiated at 410°C to 170 dpa showed 18-23% swelling.

The fracture toughness of highly irradiated SSs with >10% swelling seems to be only slightly lower than that observed in Type 304 or 316 SS irradiated to 4.5-8.5 dpa in BWRs (at 290°C) with little or no void swelling. As discussed in Section 2.2, $K_{Ic}$ values of 53 and 37 MPa m$^{1/2}$ have been reported for BWR control blade material irradiated to 4.5 and 8.0 dpa, respectively. The latter value is currently defined as the saturation fracture toughness for irradiated austenitic SSs. The results for high swelling suggest that the fracture toughness of austenitic SS PWR core internal components could be lower than the saturation value ($K_{Ic}$=38 MPa m$^{1/2}$) proposed by MRP [45]. The embrittlement of austenitic SSs due to void swelling under PWR irradiation conditions should be further investigated.

4. Stress relaxation and creep

Loss of preload for bolted joints and redistribution of stresses in components due to stress relaxation are another aging degradation concern that needs to be addressed to assure the functional integrity of the reactor core internal components. Stress relaxation represents plastic deformation that occurs with time under constant strain below the yield point of the material. In other words, stress relaxation may be considered as creep that occurs under constant strain instead of constant load or stress. It can occur either by thermal creep [16] or neutron-irradiation-assisted creep [18,19].
Thermal creep is represented by primary (or transient), secondary (or steady-state), and tertiary creep stages. For most metals and alloys, at temperatures below 0.4 or 0.5 of their melting temperature (in degrees kelvin), creep under constant load is characterized by only primary creep with decreasing creep rates, and steady-state creep is insignificant. For the 300-series austenitic SSs and Ni alloys used in LWR core internal components, this temperature is about 394°C (740°F), which is higher than the typical temperature range for PWR core components. Only a few locations close to the core may be above this temperature due to gamma heating. Thus, for PWR internals, the contribution to total strain due to steady-state thermal creep under constant load is likely to be relatively small.

Stress relaxation data are generally obtained from tests performed in accordance with ASTM Specification E 328-86 under tension, compression, torsion, or bending to reflect the stress state imposed on the component. Most of the available data are for stress relaxation tests that have been conducted under tension (for bolted joints) and bending (for leaf springs). The stress state behavior of coil springs is more complex, and there are no stress relaxation data for coil springs in the open literature. In general, the thermal stress relaxation of coil springs appears to be much larger and depends on the coil design geometry [66].

The existing data on thermal stress relaxation have been obtained from short-term tests (less than 1000 h) at 260-705°C (500-1300°F) on Type 304 SS as a function of cold work, thermo-mechanical condition, and stress level [17]. The results indicate that the percentage of stress relaxation increases with increasing initial stress and levels off at about 207 MPa (30 ksi). At PWR operating temperatures, stress relaxation saturates within 100 h with a maximum decrease of 10-20% of the initial preload stress depending on the thermal-mechanical treatment of the material [66]. At temperatures below 315°C (600°F), for the same initial stress, the extent of stress relaxation decreases with increasing amount of cold work in the material. The effect of cold work is less significant at temperatures above 371°C (700°F). Stress relaxation reaches saturation in the short-term tests because only an elastic strain and microplastic strain contribute to the total time-dependent strain (or creep strain). Secondary or steady-state creep is insignificant in these tests at relatively low temperatures (i.e., less than 394°C). However, over the reactor lifetime, even a rather low steady-state creep rate could lead to appreciable stress relaxation than anticipated from the short-term tests. At 350°C (662°F), the steady-state creep rate for Type 304 SS is estimated to be $5.2 \times 10^{-13}$ s, which corresponds to 0.049% strain over 30 y; the total stress relaxation is 76.5 MPa (11.1 ksi) [16].

Neutron-irradiation-enhanced creep can greatly increase the plastic strain by increasing both the primary creep and steady-state creep rates. Unlike thermal creep, neutron irradiation can result in significant creep strain at PWR temperatures. Irradiation creep can be divided into three stages: transient, steady state, and void-swelling-induced creep. The magnitude of both transient creep and steady-state creep is essentially proportional to the applied stress level. Transient creep is typically short and, although its magnitude depends on material condition and structure and irradiation temperature, the contribution to total creep strain is relatively small (PWR fuel rods show <0.05% transient creep) [12]; the steady-state creep rate is independent of these parameters. Figure 14 plots the thermal creep and irradiation-enhanced creep of 20% CW Type 316 SS in the EBR-II fast reactor at 454°C (850°F) and under an uniaxial stress of 138 MPa (20 ksi).

Empirical models have been developed for irradiation creep by using data obtained from fast reactors. Because of the relatively weak dependence of steady-state creep on dose rate or temperature, the fast reactor data have been used to estimate irradiation creep at PWR operating temperatures [66]. The existing data are inadequate to determine the effects of cold work or material composition on irradiation-enhanced steady-state creep.
Initially, void-swelling-induced creep was not considered to be significant for PWR core internal components. However, void swelling could be significant for PWR core internals because of the low dose rates, which could lead to considerable creep strain. A detailed evaluation based on typical temperature/dose rate information for core components is needed to estimate the contribution of void-swelling-induced creep.

Stress relaxation is creep that occurs at constant strain. It follows the trends observed in thermal and irradiation creep. Initially, there is a sharp decrease in stress representing transient creep, followed by a gradual decrease in stress due to essentially steady-state irradiation creep. Empirical equations have been developed from the irradiation creep data to estimate stress relaxation as a function of initial stress and neutron dose, and the constants in the equations were derived by regression fitting to the available stress relaxation data. One such curve representing stress relaxation (%) as a function of neutron dose is given by [66],

\[
\frac{\sigma}{\sigma_0} = \exp\left\{-0.774\left(1 - \exp\left(-23f\right)\right) - 0.688f\right\},
\]

where \( f \) is the accumulated neutron fluence in dpa, and \( \sigma \) and \( \sigma_0 \) are the remaining and initial stress. The first term [i.e., 0.774(1 - exp(-23f))] in the equation represents stress relaxation due to irradiation-enhanced transient creep, and the second term [i.e., 0.688f] represents stress relaxation due to steady-state creep. The screening trend curve obtained with the available stress relaxation data is shown in Fig. 15. The curve bounds all available test data for Type 304 SS, CW Type 316 SS, and Alloy X-750 [18-20]. The results indicate approximately 60% stress relaxation at a neutron dose of 0.2 dpa (or \( 1.3 \times 10^{20} \) n/cm\(^2\) E \( >1 \) MeV). Based on the screening trend curve in Fig. 15, all bolted PWR internals or spring component items that require preload for functionality and reach 0.2 dpa or higher are candidates for further evaluation [66].

However, most of the existing data on irradiation-enhanced stress relaxation/irradiation creep have been from fast reactor irradiations, and data obtained under PWR irradiation conditions is limited. Therefore, the effects of neutron spectrum and He production rate are not known.

Fig. 14. Thermal creep and irradiation-enhanced creep of a 20% CW Type 316 stainless steel irradiated in EBR-II (Ref. 66).
5. Summary

The fracture toughness of austenitic SSs has been divided into three broad categories. The fracture toughness (J_{lc}) is above 150 kJ/cm² for Category III materials, and 30-150 kJ/cm² for Category II materials. These materials fracture after stable crack extension at stresses well above or close to the yield stress. Category I materials fracture at stress levels well below the yield stress by unstable crack extension, and their fracture toughness K_{lc} is less than 75 MPa m^{1/2}. Nonirradiated wrought and cast austenitic SSs and their welds fall in Category III. However, neutron irradiation degrades the fracture toughness of austenitic SSs to the level of Category II or even I at high dose levels.

Fracture toughness data on irradiated wrought and cast austenitic SSs and their welds have been compiled and evaluated to define the threshold neutron dose above which fracture toughness of these materials is reduced significantly. The validity of the fracture toughness J-R curve data and the different methods for determining J_{lc} are discussed. Comparison of small specimen data with valid fracture toughness data indicate that small specimens yield equivalent fracture toughness J-R curve data, at least for materials with J_{lc} below 300 kJ/m^{1/2}.

The existing fracture toughness data on austenitic SSs irradiated in LWRs indicate little or no loss of fracture toughness below an exposure of about 0.5 dpa and a substantial and rapid decrease at exposures of 1-5 dpa. Also, fracture toughness appears to saturate at approximately 8 dpa. A similar trend was observed earlier for austenitic SSs irradiated in high-flux fast reactors at 350-427°C and tested at 300-427°C. However, for the LWR-irradiated materials, the saturation fracture toughness K_{lc} or K_{Jc} values are in the range of 36.8-40.3 MPa m^{1/2} (33.5-36.6 ksi in^{1/2}), which are lower than the K_{lc} of 75 MPa m^{1/2} (68.2 ksi in^{1/2}) observed for the fast-reactor-irradiated materials. Several fracture toughness J-R curve tests on Type 304 SS irradiated to 4.5-5.5 dpa showed no ductile crack extension.

Most of the fracture toughness data on LWR-irradiated austenitic SSs have been obtained on Type 304 and 304L SSs, and similar data on SS welds, cast SSs, or weld HAZ materials are very limited. The available data indicate that for the same irradiation conditions, the fracture toughness of thermally aged cast SS and weld metal is lower than that of HAZ material, which, in turn, is lower than that of
solution-annealed materials. Available data indicate a strong orientation effect on fracture toughness; fracture toughness in the transverse orientation is significantly lower than that in the longitudinal orientation. Potential effects of the coolant environment and crack morphology on fracture toughness of irradiated SSs are discussed. The available data are inadequate to establish accurately the effects of the irradiation temperature on the fracture toughness of austenitic SSs.

The existing fracture toughness data have been evaluated to develop a fracture toughness trend curve that includes (a) a threshold neutron exposure for radiation embrittlement of austenitic SSs and a minimum fracture toughness for these materials irradiated to less than the threshold value, (b) a saturation neutron exposure and a saturation fracture toughness for materials irradiated to greater than this value, and (c) a description of the change in fracture toughness between the threshold and saturation neutron exposures. However, a review of the existing data indicated very limited data on materials irradiated in LWRs to neutron dose levels of 0.1-1.0 dpa or above 10 dpa, and on LWR-irradiated cast austenitic SSs and welds. The contribution of additional precipitate phases, voids, and cavities on the fracture toughness of these materials needs to be investigated.

Cast and welded austenitic SSs have a duplex structure consisting of austenite and ferrite phases, and are susceptible to thermal embrittlement during service at LWR operating temperatures. Formation of the Cr-rich $\alpha'$ phase in the ferrite is the primary mechanism for thermal embrittlement of these materials; thermal aging has no effect on the austenite phase. Embrittlement of the austenite phase from neutron irradiation under LWR operating conditions occurs at dose levels above 0.5 dpa, and that of the ferrite phase occurs at lower dose levels (above 0.07 dpa). However, for reactor core internal components, concurrent exposure to neutron irradiation can result in a synergistic effect wherein the service-degraded fracture toughness can be less than that predicted for either thermal embrittlement or neutron irradiation embrittlement independently. The available fracture toughness data are inadequate to evaluate the synergistic effects of thermal and neutron embrittlement on the threshold dose for embrittlement. Additional data are needed to better establish the potential for synergistic loss of toughness in these materials in the transition dose range from 0.05 to 2 dpa.

Void swelling refers to the volume change of materials under neutron irradiation. Volume changes can produce dimensional distortions and misfits in reactor internal components, particularly when coupled with irradiation creep, which can compromise the functional integrity of the components. Microscopic voids, developed from vacancy coalescence, give rise to this geometry instability, with important consequences on the strength and resistance to failure of these materials. All materials are believed to reach a steady-state swelling rate of ~1%/dpa. The time to reach the steady-state rate (i.e., the transient regime) depends on the material composition and thermo-mechanical condition, stress state, neutron flux and spectrum, or irradiation temperature (above 300°C).

The existing data have been reviewed and analyzed to assess void swelling in PWR core internal components and to evaluate the effects of key material and environmental parameters. The results indicate that in the 300-series SSs, void swelling decreases with increasing Ni, P, or Si contents. However, radiation-induced precipitation of second phase particles can remove Ni, P, and Si from the alloy matrix and thereby increase void swelling. In addition, cold work prolongs the transient regime for void swelling in austenitic SSs. Void swelling in austenitic SSs is typically observed at irradiation temperatures above 340°C (644°F), and shows a strong dependence on temperature. The low temperatures, typical of LWRs, extend the transient regime to much longer times. Nonetheless, depending on the irradiation temperature, dose, and dose rate, voids can form under PWR operating conditions within the reactor lifetime. Available data obtained in fast reactors suggest that under comparable irradiation temperatures and neutron dose levels, the lower displacement rates typical of
PWRs would result in higher swelling. The effects of applied or swelling-induced stress on void swelling, including the interaction of swelling and irradiation creep relaxation, are discussed. The stress-free swelling equations, including stress-enhanced swelling, developed by the MRP for solution-annealed Type 304 and cold-worked Type 316 SSs are also described.

Embrittlement of the material due to voids is also discussed. Typically, austenitic SSs with ≥10% void swelling suffer from severe embrittlement, particularly at room temperature. This behavior has been attributed to a second order process, which involves stress concentration between voids, Ni segregation at void surfaces, and a resultant tendency toward martensite formation when the steel is deformed at room temperature. However, there are few quantitative data correlating void swelling with fracture toughness of the material. The embrittlement of austenitic SSs due to void swelling under PWR irradiation conditions should be further investigated.

Loss of preload for bolting and springs due to stress relaxation is another aging degradation effect that needs to be addressed to assure the functional integrity of the reactor core internal components. Stress relaxation represents plastic deformation that occurs under constant strain below the yield point of the material. Thermal stress relaxation data have been obtained from short-term tests. At PWR operating temperatures, stress relaxation saturates within 100 h with a maximum decrease of 10-20% of the initial preload stress. The extent of relaxation depends on thermal-mechanical treatment of the material. For PWRs, plastic strains resulting from thermal stress relaxation are considered to be insignificant and not a concern for the dimensional change of core internal components. However, neutron-irradiation-enhanced creep at PWR temperatures can greatly increase the plastic strain by increasing both the transient creep and steady-state creep rates. Empirical models have been developed for irradiation creep based on data obtained from fast reactors. The results have been used to develop expressions for estimating stress relaxation as a function of initial stress and neutron dose.

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