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# Current status and future of IASCC research

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**Abstract**

Irradiation Assisted Stress Corrosion Cracking (IASCC) is a complex phenomenon occurring in structural components of nuclear reactor cores. It assumes a greater importance for the selection of suitable materials for fusion wall applications. This paper reviews the available data on the effect of irradiation damage on the environmental cracking behavior of austenitic and ferritic steels. It is shown that the changes in microstructures of the materials and the environmental changes due to irradiation have been widely investigated. However, further research is required to study the behavior of initiation and propagation of stress corrosion cracks, considering these as two separate events, for austenitic and ferritic steels under irradiation. 

## 1. Introduction

Irradiation Assisted Stress Corrosion Cracking (IASCC) has been drawing more attention over the years and has become potentially a critical phenomenon for core internals in light water reactors (LWRs). Alloys of iron and nickel base and oxygen free copper are the materials found to be affected by IASCC. It is widely recognized that IASCC is a result of the interaction of irradiation, material, environment, temperature and stress. The complexity of IASCC arises from the fact that irradiation has an impact on all the other variables listed above so that the knowledge available on SCC of materials in non-irradiated environmental conditions is not sufficient to solve the IASCC problem. Because irradiation can alter the microstructure and microchemistry of the material, can affect the aggressiveness of the environment by water radiolysis, can increase the temperature of the parts by gamma heating and can change the component stresses through relaxation of creep or by radiation hardening, interpretation of wide range of issues influencing IASCC requires specialized knowledge covering fracture mechanics, electrochemistry, physical metallurgy and core neutronics [1–10].

IASCC may have a higher potential to occur in fusion reactor components because of the higher dose rate of neutron irradiation than in LWRs. Components of the blanket and first wall cooling system, divertor cooling system and vacuum vessel cooling system are potential problem sites where IASCC could occur.

Though the mechanism of IASCC is not fully understood, factors affecting it are well documented, especially the effect of radiation on environment and on material properties.

Among the radiation effects, some are fluence dependent and some are flux dependent, while both fluence and flux cause joint effects. Radiation induced segregation (RIS), radiation induced microstructures and radiation creep relaxation are fluence dependent while radiolysis and to some extent RIS are flux dependent. The ratio of thermal to fast neutron flux affects transmutation.

This paper presents an overview of the current understanding of IASCC of the materials used in nuclear power generating environments and the need for future research on candidate materials of use in fusion reactors.

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## 2. Environmentally assisted cracking (EAC) and its implication to fusion technology

### *2.1. EAC of austenitic stainless steel in high temperature water*

EAC of stainless alloys in high temperature water occurs due to the synergistic interaction of stress, environment and material. Generally, active path corrosion cracking and hydrogen cracking are the mechanisms involved in EAC. Crack initiation and crack propagation are two distinct events, which are controlled by environmental, mechanical and material variables. Water chemistry and microchemistry of the material play a vital role in initiating the SCC. Dissolved oxygen and $CO_2$, presence of $Cl^-$ and $SO_4^{2-}$ are deleterious from an environmental point of view and inclusions such MnS, segregation of Si and P, sensitized microstructure and the presence of secondary phases such as sigma, laves, chi etc., are detrimental from a material point of view.

### *2.2. EAC of unirradiated ferritic/martensitic steels*

Ferritic stainless steels (>17% Cr) are considered to have better SCC resistance than austenitic stainless steels. This is true only when the Ni, Cu and Co contents are below certain levels [11]. However, 8–12% Cr steels are subjected to both SCC and hydrogen embrittlement. Apart from the environmental factors such as dissolved oxygen, presence of sulfate and chloride ions etc., microstructural condition of the material also control the cracking behavior. Untempered martensite and acicular bainite phases are found to be more prone to hydrogen cracking than tempered martensite and bainite + ferrite phases [12]. Generally it is observed that pitting is associated with the initiation of SCC or corrosion fatigue in this type of material [13–15]. Mostly intergranular cracking is observed along the prior austenite grain boundaries. However, though a large amount of literature is available on this subject, it is not very clear why only the prior austenite grain boundaries are the most preferred site for cracking and not other boundaries such as interlath boundaries or interfaces between two martensite packets. Probably certain solute elements segregated in the austenite grain boundaries may have more affinity to hydrogen, as discussed by Leslie [16]. But, Auger electron spectroscopy carried out on these fracture surfaces did not throw much light on this aspect.

Hydrogen cracking resistance of ferritic/martensitic steel is significant for fusion wall application because direct transmutation, water lithium interactions, radiolysis of water and corrosion could charge hydrogen into the steel. Hydrogen cracking could be enhanced by other irradiation damage mechanisms such as RIS, increased defect density etc.

## 3. Irradiation assisted stress corrosion cracking

### *3.1. Materials aspect*

Fig. 1 illustrates schematically the collision of an energetic particle (either a neutron, electron or proton) with a lattice atom generating radiation damage [6]. If the energy transfer of the elastic collision is greater than

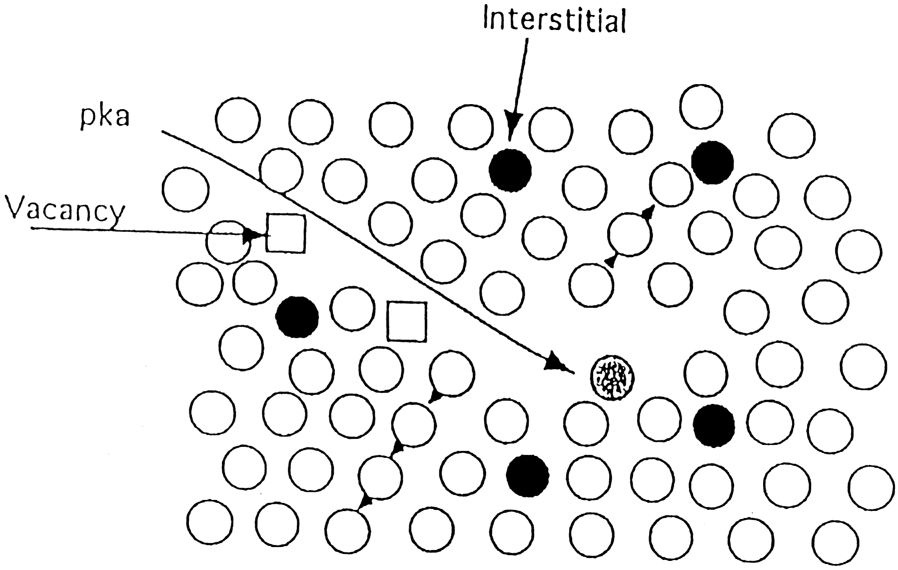


Fig. 1. Schematic illustration of generation of a primary knock-on atom (pka) [6].

the displacement threshold ($E_d$), a primary knock on atom (PKA) is generated. PKA can displace additional atoms through the lattice if it has sufficient energy until the energy of all the atoms has been reduced below $E_d$ [7]. The Frenkel pair, consisting of a vacancy and self interstitial atom (SIA) could be considered as the fundamental component of radiation damage [4].

The extent of radiation damage is a function of temperature. Several extensive reviews on microstructural evolution in irradiated austenitic stainless steels are available [5–9] which report a transition of microstructural damage approximately at 300°C. 50–300°C can be termed as the low temperature region and high temperature behavior is between 300–700°C. At low temperature, the vacancies are at equilibrium and they form clusters resulting in 'black spot' loops of <5 nm diameter.

At high temperatures, vacancy clusters in austenitic stainless steels become thermally unstable. The presence of voids and swelling are observed at higher temperatures. Under certain conditions small gas filled bubbles can grow to form voids, referred to as swelling, as the volume of material increases beyond the size limitation dictated by the thermodynamic equilibrium of gas. Both hydrogen and helium play an important role in swelling of a material [17,18].

A swelling rate of 1% per dpa is maintained at temperatures above 425°C. The lower limit of temperature for swelling is observed to be affected by displacement rate. At lower displacement rates (typical of PWRs) or at very low flux positions, cavity formation was observed even at 280–300°C.

### 3.1.1. Radiation induced microchemistry

Grain boundary segregation and radiation induced second phase precipitation are results of radiation damage, black spot. In austenitic stainless steels, depletion of Cr and Fe and enrichment of Ni have been reported. As Cr and Fe have higher diffusivity than Ni, they migrate away from the interface, enriching the boundary with Ni [19–30]. This could be attributed to the Inverse Kirkendall segregation. Frank interstitial loops produced by radiation damage interact with the undersized solutes like Si and P and cause segregation of these elements at grain boundaries by an uphill diffusion process [20]. Along with Cr and Fe, minor alloying elements such as Mn, Ti and Mo also are found to be depleted at grain boundaries. Mn levels drop to 0.5 at.% at grain boundaries in type 304 SS. In type 316 SS, more than 50% depletion of Mo after irradiation to 3 dpa has been reported [21]. For the same level of irradiation, enrichment of Si occurred to levels of about 6–8 at.%. Nickel-silicide precipitation also has often been reported to form at dislocation loops at higher temperatures (>380°C) and at higher doses (>20 dpa) [6]. Segregation of Sulfur at grain boundaries due to radiation damage is not well documented. However, at higher doses (PWR-relevant, >10 dpa) sulfur segregation can be expected due to the burn-up of Mn in MnS inclusions and subsequent release of S.

### 3.1.2. Radiation-induced precipitation

At lower temperatures (<400°C), radiation induced precipitation of second phases is not expected to occur [19]. However, radiation induced Cr depletion could retard carbide formation at grain boundaries. Radiation induced segregation of Ni and Si could lead to formation of $\gamma'$ or G phase at higher temperatures.

### 3.1.3. Transmutation

The presence of certain minor alloying elements in stainless steels (SS) such as Mn, B etc., impart beneficial effects to unirradiated material. For example small amounts of boron in the ppm level improves thermal creep resistance of SS. But under irradiation, formation of He and Li from boron occurs. Burn out of MnS through radiation releases sulfur which in turn transmutes to chlorine. If MnS precipitates were located on the grain boundary, this enhances the IASCC [17]. Formation of Li and He by transmutation increases the probability of IASCC. Existing as liquid at LWR-relevant temperatures, Lithium either reduces the cohesive strength of the grain boundaries or enhances the crack growth rate through exothermic chemical interactions with water in advancing cracks. Moreover, a lithium–water reaction would produce hydrogen at the crack tip [18]. The effect of He on IASCC is not clear at present.

## 3.2. Mechanical aspect

The consequences of irradiation damage on mechanical properties have been well researched with reference to the fast reactor field. But LWR irradiation effects are significantly different from those of a fast reactor because of different temperature and dose levels. So, the results from the extensively investigated fast reactor field cannot be used directly for correlating LWR conditions.

In general, it is observed that with increases in irradiation dose, the yield strength of the material increases. The ultimate tensile strength also increases, but the increase is not as great as for the yield strength. Formation of higher densities of vacancies and interstitials is attributed as the cause for this increase. Suzuki et al. [21], reported increases in strength for various grades of austenitic stainless steels with increase in neutron fluence as shown in Fig. 2. However, a saturation level is reached at the $3 \times 10^{25}$ n/m$^2$ fluence level beyond which no further increase in strength could be observed. It was observed that type 304 SS was more prone to irradiation hardening than was type 316. Composition has two effects viz., (1) certain alloy elements help nucleate Frank

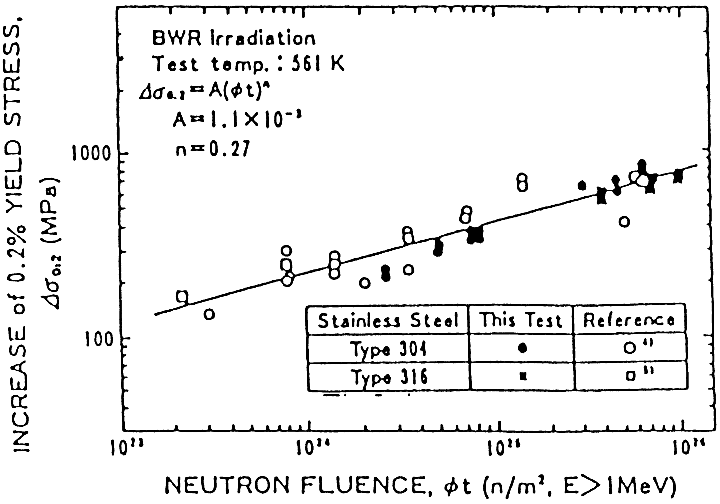


Fig. 2. Relation between the increase of the 0.2% yield stress and neutron fluence [21].

loops and (2) stacking fault energy (SFE) is altered. Low SFE results in more hardening. Also a low SFE can lead to nucleation of twins as an alternative deformation mechanism to dislocation glide. Alloying elements such as Ni, Mo and C increase the SFE in austenitic stainless steel and Cr, Si, Mn and N tend to decrease the SFE [25].

### 3.2.1. Work hardening and elongation

Loss of work hardening and uniform elongation is observed after irradiation. The elongation decreases significantly with increasing dose. This kind of loss in work hardening and hence uniform ductility could be attributed to the irradiated microstructure, where annihilation of barriers occurs due to their interaction with dislocation [27]. Interacting with obstacles, dislocations multiply in unirradiated material which results in development of back stresses and hence work hardening of the material. However, in irradiated conditions, the obstacles such as loops and voids can be destroyed when they interact with moving dislocations, resulting in work softening. This behavior causes flow localization, and hence the slip band spacing increases, ultimately reducing the macroscopic deformation [19–22]. At higher temperatures (above 600°C), the ductility is observed to be severely affected by He embrittlement. When a large void population develops near 400°C, the fracture mode is observed to be transgranular channel. The degradation in ductility and change in fracture mode result in decrease in fracture toughness for austenitic stainless steels.

The reduction of fracture toughness of irradiated SS can be attributed to the higher population of voids so that fracture occurs at an early stage by dislocation channeling or highly heterogeneous deformation–decohesion ahead of the crack tip [21].

RIS of Ni at voids also results in brittle behaviour of a material. This preferential segregation of Ni at voids results in matrix depleted of Ni and hence destabilizes the austenite. The strain induced martensite transformation, possible in the destabilized austenite, acts as low energy path for crack propagation [22]. This mechanism for cracking resulted in quasi-cleavage fracture with an overall fracture toughness of 80 MPa$\sqrt{m}$ after the austenitic material has been irradiated to high dose ($1.6 \times 10^{23}$ n/cm$^2$) at 425°C [23].

### 3.2.2. Irradiation hardening/softening of ferritic/martensitic steels

Irradiation hardening and softening are important factors in determining the fusion reactor life limits as EAC and creep properties are affected by these changes [31–40]. In ferritic steels, the irradiation hardening is attributed to the formation of small defect clusters and dislocation loops, with associated precipitation of small carbides such as $M_2C$, $M_6C$, etc. [35]. Whereas irradiation softening is considered as a consequence of the recovery of dislocation structure or coarsening of carbides etc. [36].

Kimura et al. [37] studied the irradiation hardening behavior of 9Cr–2W–V steel and reported saturation of irradiation hardening at a dose level of about 10–15 dpa. Irradiating at above 430°C, resulted in softening at dose levels of 40–60 dpa. Swelling was found to be associated only with hardening, in this study.

Shiba et al. [38] investigated the response of F82H steel to irradiation at low damage levels (<1 dpa) for 300–500°C. They observed hardening only at 300°C. However, no softening was observed when irradiated at 520°C. Interestingly for these test conditions, no change in DBTT was observed between irradiated and unirradiated charpy impact tested samples. However the upper shelf energies were lower for irradiated samples.

Klueh and Alexander [39] conducted a detailed study on Charpy impact toughness of different types of low activation steels irradiated at 0–24 dpa. They observed that 9Cr–2W type steels were least affected by irradiation. In terms of microstructure, they found that steels with 100% martensitic structure showed superior impact toughness properties after irradiation than steels with dual microstructures such as bainite + ferrite or martensite + ferrite. By changing the composition and microstructure, the effect of irradiation on toughness could be favorably modified.

Khabarov et al. [40] analyzed the effect of neutron irradiation at low temperatures (280–350°C) to doses of 85 dpa on irradiation hardening of 13Cr2MoNbVB steel. They observed increase in yield strength with increase in fluence, which can be fitted by the equation $\Delta\sigma$ (MPa) $= 300 \times (\text{dose})^{0.2}$ for the dose level of 0.5–25 dpa. Beyond 40 dpa the yield strength increment decreased. Ductility and impact toughness values inversely followed the trend for yield strength. Dislocation loops, $\alpha'$phase precipitation, voids and $M_2X$ precipitation were attributed as the reasons for the change in mechanical behavior after irradiation. Typical 12% Cr steel shows very little void swelling (only 0.1% volume change for a dose of 90 dpa at 400°C [36]. It is generally observed that 9% Cr steel shows better impact toughness and DBTT values after irradiation than 12% Cr steel. However, by controlling the phase content (either 100% martensite or at least less than 20% delta ferrite) and with uniform distribution of carbide/carbonitride phases, much improved mechanical properties could be achieved.

### *3.2.3. Creep and creep rupture*

Two groups of mechanisms have been proposed for irradiation creep [6] viz., (1) irradiation induced creep and (2) irradiation enhanced creep.

*Irradiation induced creep*. Depending on the nucleation and growth of dislocation loops, two types of mechanisms are observed viz.,

1. Stress induced preferred nucleation (SIPN),
2. Stress induced preferential absorption (SIPA).

For the case of SIPN [5], interstitial loops are assumed to nucleate preferentially on planes perpendicular to the tensile stress and vacancy loops nucleate preferentially on planes parallel to the applied stress.

For the SIPA mechanism, interstitials are preferentially absorbed by the loops oriented perpendicular to the tensile stress so that there is an elongation in the direction of stress.

*Irradiation enhanced creep*. As the name suggests, in this case it is postulated that irradiation accelerates the thermal creep by producing excess vacancies and interstitials and thus facilitating the easier dislocation movements [6]. For example, jogs in the form of non-glissile edge dislocation segments on a screw dislocation can be moved by the point defects produced by irradiation, otherwise not possible.

*Significance of irradiation creep on IASCC*. Irradiation creep results in higher crack tip strain rate and enhances the crack growth. Irradiation creep relaxes the stresses in components with fixed deflection and may reduce the propensity for IASCC. However, the enhancement of the crack tip strain rate due to irradiation creep was calculated to be a few hundred times lower than the strain rate due to crack growth itself [5]. Hence, irradiation creep could predominantly relax the stress in the component.

### *3.2.4. Irradiation assisted creep and fatigue behavior in ferritic steels*

In the case of 9% Cr steel the volumetric swelling was around 0.1% at 100 dpa and 400°C which is similar to that for 12% Cr steel. After 90–100 dpa, the rate of swelling of 9% Cr steel was reported to be approximately 0.01%/dpa [41].

Bertsch et al. [42] studied the post irradiation fatigue properties of the ferritic–martensitic steel MANET at 250°C. The samples irradiated during fatigue testing showed increase in irradiation hardening with increase in the number of cycles. At higher strain ranges they observed strength recovery. Helium bubbles were observed in irradiated fatigue tested samples.

## *3.3. Environmental aspect*

### *3.3.1. Radiolysis*

Radiolysis is a complex issue effected by water chemistry, neutron flux (not fluence), flow rate, temperature etc. Radiation causes decomposition of water into many species which affect the corrosion potential. At high hydrogen levels (>1 ppm), radiolysis is sufficiently suppressed so that it has very little effect on changing the corrosion potential [5–7]. The interior of the cracks were not found to be polarized by radiation, as the corrosion potentials of cracks and tight crevices were not altered.

### *3.3.2. Flux dependence*

The structural materials are exposed to temperatures of 290–350°C in water reactors. In the case of a BWR the temperature is constant at 288°C, whereas in a PWR, the temperature varies with location to a maximum of 400°C in the baffle plates. The fast flux in a BWR is around $7 \times 10^{17}$ n/m$^2$ s ($E > 1$ MeV) and in a PWR, it is 20–30% higher than in a BWR. However, to express the level of radiation damage, the best method is reported to be use of displacements per atom (dpa) as calculated by approved methods [5]. Empirically 1.4 dpa per $10^{25}$ n/m$^2$ ($E > 1$ MeV) is used for LWRs. From this, the fast flux can be back calculated to be $10^{-7}$ dpa/s in the core of

LWRs and 1.5–4 × $10^{-7}$ dpa/s in test reactors. In fast reactors, the fast flux is given approximately as $10^{-6}$ dpa/s, and the temperature also is higher (>370°C) in fast reactors. So, the data generated in fast reactors cannot be compared with those of LWRs. The thermal-to-fast flux ratio also is an important issue. The thermal neutrons are those which are in thermal equilibrium with neighboring atoms and with energies below 0.5 eV.

*3.3.3. Radiation water chemistry and corrosion potential*

Radiation causes break down of water into primary species ($H^+$, $e^-_{aq}$) and molecules such as $H_2O_2$, $O_2$, $H_2$ etc. The concentration of species is proportional to the square root of the radiation flux. Fast neutron radiation has a stronger effect on water chemistry than other types of radiation such as thermal neutrons, beta particles and gamma radiation [21]. This feature is because of the higher linear energy transfer (LET) and the higher neutron flux of fast neutrons.

It is generally believed that the corrosion potential has more influence than the concentration of oxidizing and reducing species in controlling SCC. Fig. 3 shows the effect of electrochemical potential (ECP) on SCC of unirradiated and irradiated type 304 SS in a BWR environment. The initial concentration of oxygen and hydrogen are found to be important in determining the final corrosion potential after irradiation. Though a large increase in concentration of some species occurs after irradiation, the change in corrosion potential is not drastic. When hydrogen was present at more than 200 ppb and at 0 ppb $O_2$, there was no radiation induced elevation of corrosion potential [22], whereas, the presence of $H_2O_2$ increased the corrosion potential.

## 4. Quantitative evaluation of IASCC susceptibility

### *4.1. Test method dependence of IASCC*

Under service conditions, the loading of components vary considerably depending upon their function and location with respect to the reactor core. For example, torque on bolts, residual stress on weldments with constraint, fit up stresses on sleeves, key ways etc., pressurization, internal swelling, differential thermal stresses and thermal fatigue etc., are different loading conditions in actual service, that are difficult to simulate in laboratory.

In turn, simpler loading conditions are used in a laboratory or during test reactor experiments, such as slow strain rate testing (SSRT), mandrel swelling tubes, constant load testing, constant displacement testing etc. These are relevant only for qualitative analysis and some extrapolation methodology is required for interpretation of these accelerated test data into meaningful design data for life limit estimations of actual components.

As the irradiation causes stress relaxation, it is not advisable to use either constant displacement test methods or constant load methods. Always, the use of active constant load, active constant stress intensity or active strain rate is desirable, as it will give better reproducible results. SSRT on notched specimen also will give a complex picture of strain rate distribution vs. time. Moreover, SSRT may not be a appropriate method to study the effect of radiation flux on creep, as the induced strain rate is many orders of magnitude higher than for radiation creep.

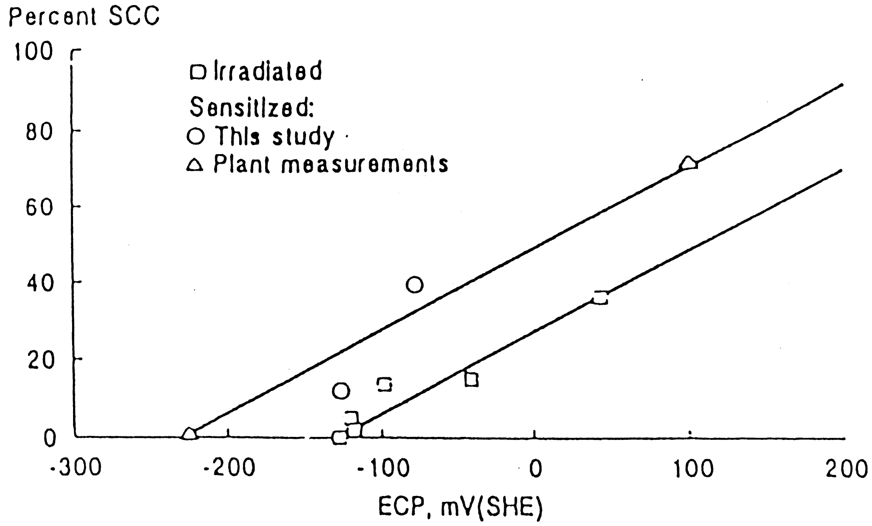


Fig. 3. ECP on cracking behavior of irradiated and sensitized type 304 SS [22].

### 4.2. Crack initiation and propagation

Not many studies have been done on the crack initiation due to IASCC. The definition of crack initiation is not very clear. However, a crack of one grain diameter deep can be assumed as an active crack whereas, a crack of 20 μm shows mature chemistry [7,22,23]. Crack initiation has more or less the same controlling parameters as crack propagation. Nevertheless, certain environmental and mechanical conditions affect only the crack initiation, such as surface preparation, pits and strain concentration around the crack etc.

#### 4.2.1. Crack initiation

It is generally observed that SCC initiation preferentially occurs at sites like pits and second phase particles. Preferential dissolution of secondary phases or inclusions creates a crevice where the local electrolyte chemistry and local strain level become more favourable for SCC initiation by a slip dissolution mechanism. In the case of IASCC, irradiated microstructural features (like Cr depletion, Si and P segregation etc.) and the presence of hard phases such as oxides make the crack initiation process much easier. Oxide particles effectively participate in IASCC initiation by two proposed mechanisms as followed [22]:

1. Oxides (10 μm in size) are hard to deform. So, under load, the shear stress at the interface of the oxide–matrix increases to very high levels as the ductile matrix around the particle deforms. This results in failure in the bonding, creating a crevice where the local chemistry of the electrolyte changes to more a conducive condition for promoting SCC.
2. Alternately, the oxide could fracture creating a microcrack which can either extend into the matrix or create a very high stress intensity for easy SCC initiation.

Strain at crack initiation (SCI) was proposed as the definition for IASCC initiation in SSRT by Tanaka et al. [23]. It was defined as the strain at which the stress-strain curve of SSRTs began to depart from that of tensile tests, when plotted using the same coordinates. An example is shown in Fig. 4. Higher SCI means SCC initiation starts at higher strain. Fig. 5 shows the dissolved oxygen (DO) dependence for IASCC susceptibility. Though the intergranular (IG) fracture ratio decreases with decreasing DO, it increases inversely below 10 ppb of DO. This phenomena may indicate the continuum of initiation of IASCC from BWR conditions to PWR conditions.

#### 4.2.2. Crack propagation

Nakata et al. [25] studied the IGSCC growth rates of type 304 SS both in unirradiated and irradiated condition using gamma rays (from Co-60). They also studied the effects of DO, hydrogen and addition of sulfate ($SO_4^{2-}$) and nitrate ($NO_3^-$) ions on IGSCC. Gamma ray irradiation is not expected to affect the microstructure or

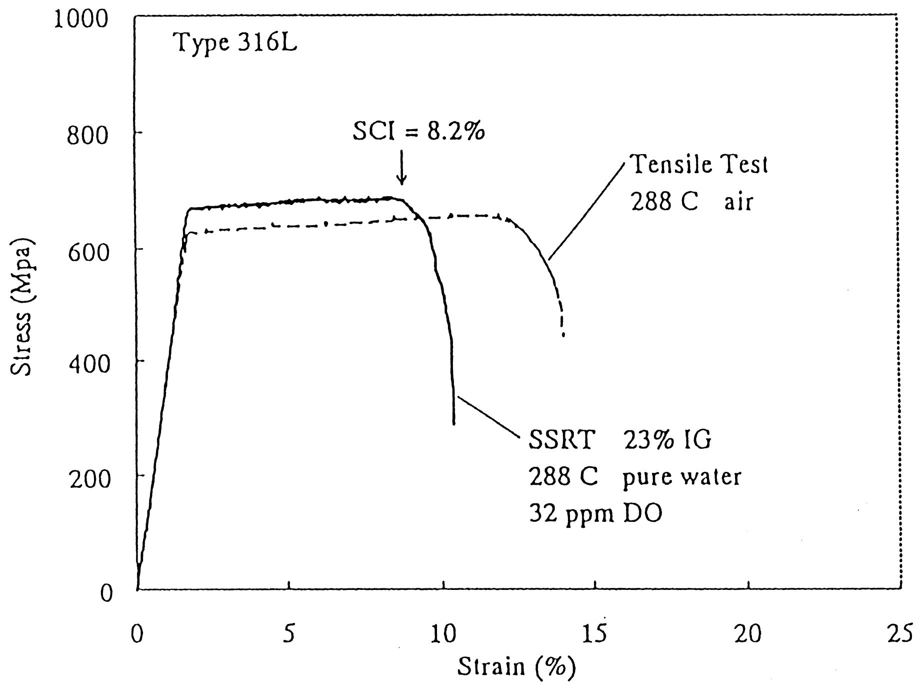


Fig. 4. Stress–strain curves from SSRT and a standard tensile test for 316L [23].

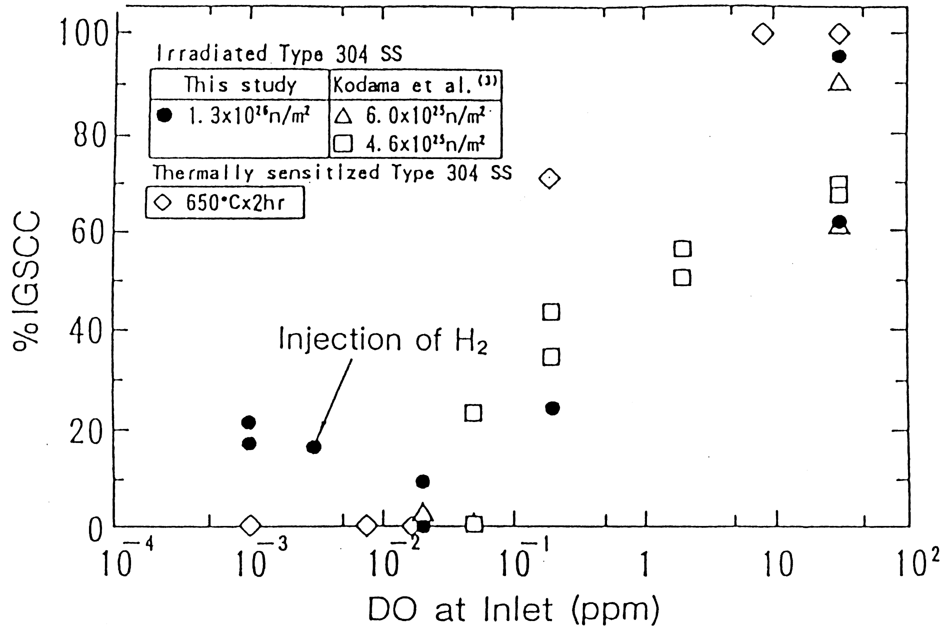


Fig. 5. DO dependence of percentage of IGSCC in type 304 SS. Reference data obtained from type 304 SS irradiated in BWRs [24].

microchemistry of the material. However, it decomposes water into many kinds of radiolytic products of which hydrogen peroxide ($H_2O_2$) is very important to IASCC. In the 288°C BWR environment gamma irradiation accelerated the crack growth to varying degrees depending on the water chemistry, flux etc. For example, the average crack growth rates in unirradiated, irradiated with gamma ray for fluxes of $5 \times 10^6$, $9 \times 10^6$ R/h were $7.2 \times 10^{-10}$, $1 \times 10^{-9}$ and $1.3 \times 10^{-9}$ m/s, respectively. From these values, the crack growth rates in low conductivity pure water could be observed to be marginally affected by gamma ray irradiation. Fig. 6 shows the effect of DO on crack velocity with additions of $Na_2SO_4$. Both irradiated and unirradiated test conditions showed similar trends. It could be observed that addition of $SO_4^{2-}$ ions showed more effect in accelerating the crack growth than did irradiation. DO also had a similar effect. Suppressing the DO content decreased the crack growth rate [24].

Though crack velocity increased with sulfate ions as in the case of the unirradiated condition, DO had a major effect in controlling the crack behavior in the irradiated condition also. Nitrate additions were found to be less aggressive than sulfate additions in a BWR environment for 304 SS. Dissolved hydrogen showed greater beneficial effect in suppressing crack growth. Fig. 7 exhibits the effect of hydrogen injection into the BWR environment on IASCC in 304 SS [26]. The mechanism of crack growth mitigation by hydrogen injection could be explained by analyzing the corrosion potential of the system. The presence of molecules like $H_2O_2$ and $O_2$ increases the free corrosion potential which falls into the cracking range and hence the crack velocity is enhanced following the slip dissolution model and Faraday's law. Whereas, when hydrogen is introduced into the environment it helps the recombination of species and thus reduces the corrosion potential well below the cracking range.

Jenssen and Ljungberg [27] carried out IASCC tests on irradiated SS under BWR condition using the SSRT method. They presented average crack growth data by dividing the maximum crack depth by total test duration. As shown in Fig. 8 maximum crack growth rate divided by the test time was suppressed by hydrogen water chemistry (HWC) below $3 \times 10^{21}$ n/cm$^2$, but not above $3 \times 10^{21}$ n/cm$^2$. It could be observed from this data that variations in either fluence level ($3 \times 10^{20}$–$9 \times 10^{21}$ n/cm$^2$, $E > 1$ MeV) or flux level ($1.5 \times 10^{13}$–$7.6 \times 10^{13}$ n/cm$^2$ s) did not affect the crack velocity drastically (a maximum of a factor of two).

Fig. 9 shows the $K$ dependency on IASCC crack growth rate of irradiated SUS304 as a function of $m$ based on a model proposed by Shoji et al. [43], where $m$ is the repassivation rate of the bare surface of a material. Radiation induced segregation can alter the values of $m$. The model calculations predict an effect of $m$, such that there is an effect of HWC even at higher fluence, which is in disagreement with the experimental data above $3 \times 10^{21}$ n/cm$^2$. Because the model calculation uses the same parameter of a film rupture strain and numerical constants given by current decay curve, it might be explained by the difference of these param-

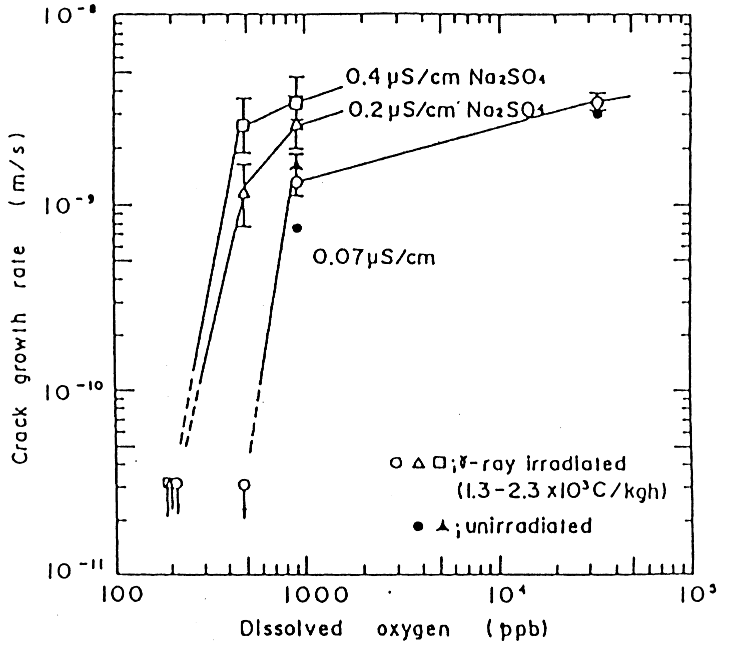


Fig. 6. DO dependence of crack growth rate under γ-ray irradiation with fluxes of 1.3–2.3 × $10^3$ C/kgh in high purity water and containing $Na_2SO_4$ [25].

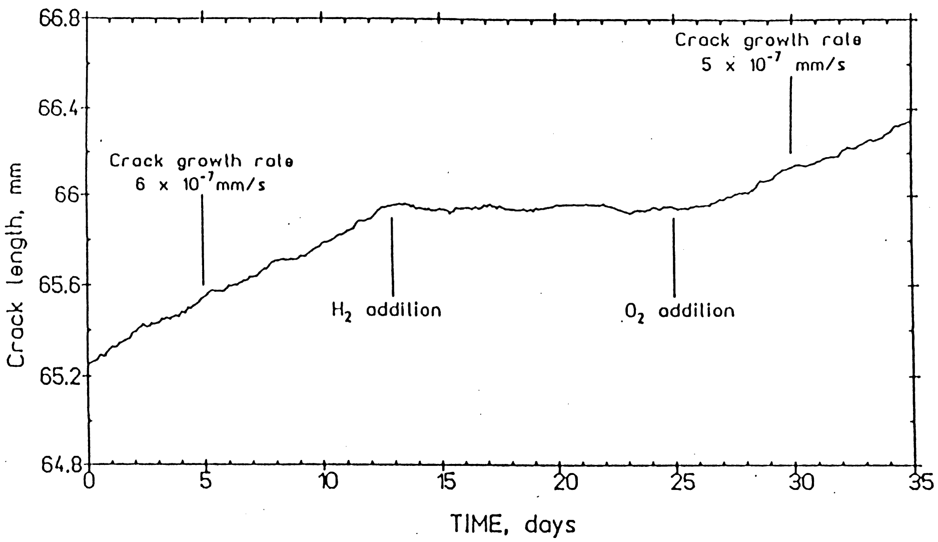


Fig. 7. Typical crack growth rates for solution annealed type 304 SS in the presence of 400 ppb $O_2$ and 400 ppb $H_2$ [26].

eters between unirradiated materials and irradiated materials.

## 5. Conclusion

From the literature overview, it is shown that irradiation damage in austenitic stainless steels is well documented with reference to changes in microstructural and environmental conditions. The intergranular cracking of austenitic stainless steels because of IASCC in the absence of any sensitization heat treatments could be explained based on the observed RIS phenomenon. The effect of fluence level on IASCC susceptibility has been studied widely. However, the issue of initiation of IASCC requires further research.

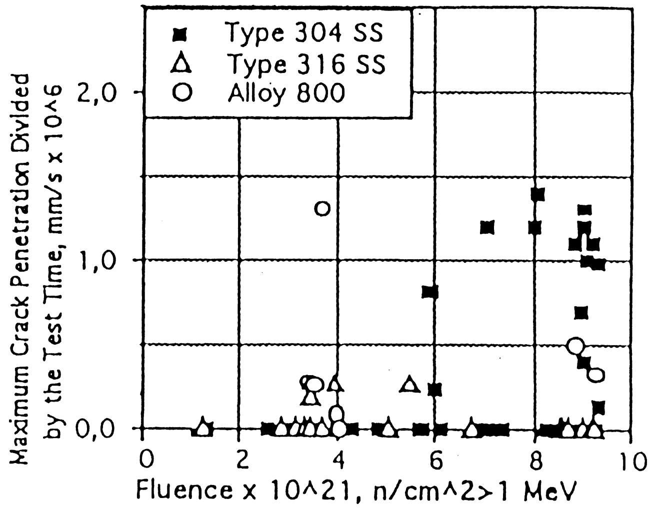


Fig. 8. IGSCC of various types of materials in HWC (ECP < −230 mV (SHE) [27].

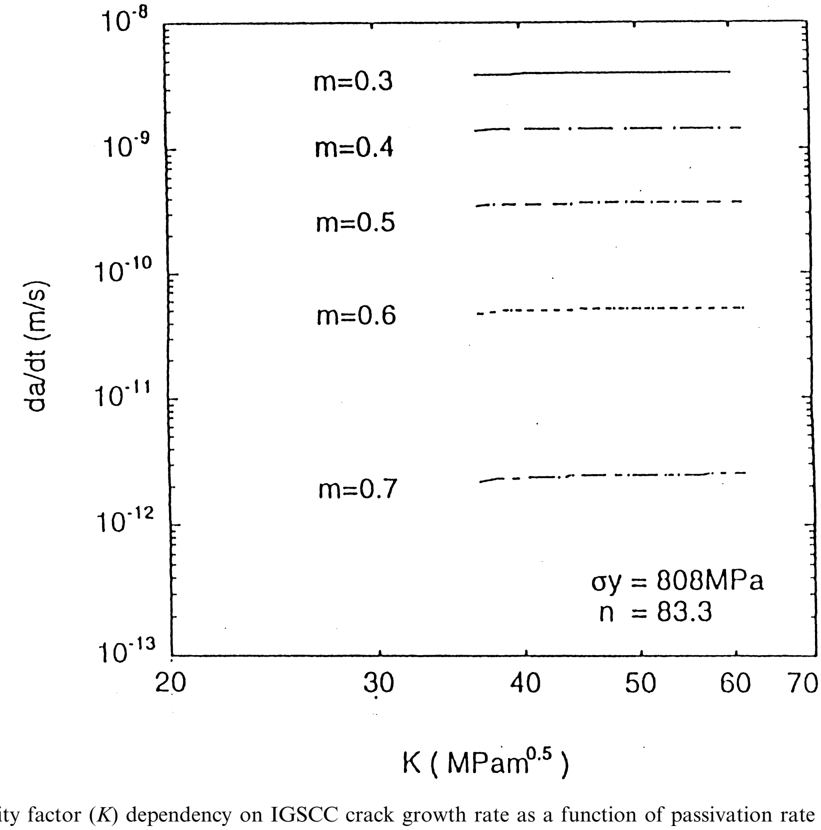


Fig. 9. Stress intensity factor ($K$) dependency on IGSCC crack growth rate as a function of passivation rate $m$. (for type 304 SS at $1.3 \times 10^{25}$ n/m$^2$, $E > 1$ MeV [43].

Regarding the use of ferritic steels for possible fusion wall applications where very high dose levels of radiation damage are expected, it is shown from the available reports that they are better suited than austenitic stainless steels. However, no information is available on SCC and hydrogen assisted cracking behavior in irradiated ferritic/martensitic steels. Hydrogen cracking in 8–12% Cr steel could be a major problem in light of its enhancement due to radiation damage.

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