



Revealing the pitting corrosion mechanism of nuclear-grade stainless steel in the presence of irradiation damage and hydrogen

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ABSTRACT

In this study, 308 L stainless steel was irradiated with 2 MeV H⁺ ions to investigate the coupling effect of irradiation-induced defects and hydrogen on pitting corrosion. The results show that, along the depth direction, the peak damaged region (PDR) exhibits the highest pitting susceptibility and the most pronounced hydrogen enrichment. In the uniform damaged region (UDR), the pitting densities from the centre to the edge are $(1.04 \pm 0.36) \times 10^{-11} \text{ m}^{-2}$ in region A, $(0.79 \pm 0.12) \times 10^{-11} \text{ m}^{-2}$ in region B, and $(0.24 \pm 0.17) \times 10^{-11} \text{ m}^{-2}$ in region C. These values correlate well with the corresponding void number densities of $(1.20 \pm 0.22) \times 10^{-15} \text{ m}^{-2}$, $(0.43 \pm 0.03) \times 10^{-15} \text{ m}^{-2}$ and $(0.31 \pm 0.08) \times 10^{-15} \text{ m}^{-2}$, and with the hydrogen areal densities of $(4.05 \pm 0.68) \times 10^{-10} \text{ m}^{-2}$, $(3.77 \pm 1.05) \times 10^{-10} \text{ m}^{-2}$ and $(1.84 \pm 0.74) \times 10^{-10} \text{ m}^{-2}$. Furthermore, combined hydrogen microprinting and TEM observations reveal that irradiation-induced voids serve as the primary hydrogen traps, thereby accelerating pit initiation and propagation. This work provides direct experimental evidence that, under high-temperature conditions, irradiation-induced voids act as preferential hydrogen trapping sites, locally disrupt the passive film and promote pit initiation. These findings uncover a previously unknown synergistic interaction between irradiation damage and hydrogen in the localised corrosion of nuclear-grade stainless steel.

1. Introduction

Hydrogen is a key clean energy carrier essential for the global energy transition, and radiation-driven catalytic water splitting represents one of the most promising technological routes for its production. Studies have demonstrated that efficient hydrogen generation can be achieved through nanoparticle photocatalysis, surface catalysis on different Al₂O₃ crystal phases, and radiation-assisted water splitting [1–3]. The development of advanced materials has further enhanced the efficiency of radiation-induced hydrogen production: zirconium and its niobium alloys catalyse hydrogen generation from seawater under gamma irradiation, exhibiting notably high radiolytic yields [4]; both nanostructured

ZrO₂ and Al₂O₃ show excellent catalytic performance under gamma irradiation [5,6]; and the water–n-hexane–ZrO₂ system further improves hydrogen production efficiency under combined thermal and radiation conditions [7]. These findings provide a foundation for understanding the interactions between hydrogen and materials in radiation environments.

Nuclear power, as a form of clean energy, has attracted increasing attention. However, the effect of hydrogen on the corrosion of reactor materials remains insufficiently understood. In pressurised water reactors (PWRs), radiolysis of the primary coolant under neutron or gamma irradiation generates hydrogen alongside a variety of oxidising species [8]. To counteract the resulting highly oxidising environment, a

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substantial amount of hydrogen is deliberately injected to reduce the coolant corrosivity [8,9]. Prolonged exposure of reactor materials to such a hydrogen-rich atmosphere, combined with irradiation-induced internal defects that trap hydrogen, can severely compromise their corrosion resistance. The coupling between irradiation damage and hydrogen thus poses a significant threat to the integrity of reactor components. This study aims to systematically elucidate the mechanistic interplay between irradiation-induced defects and hydrogen in driving the corrosion of metallic materials for critical nuclear applications, thereby providing a theoretical basis for improving the resistance of future nuclear plant structural materials to both irradiation damage and hydrogen-induced degradation.

308 L austenitic stainless steel is widely employed as the inner surface protective cladding for the pressure vessel safety end – a critical component of pressurised water reactors (PWRs) – owing to its superior mechanical properties, favourable weldability and excellent corrosion resistance [9–11]. Given its proximity to the reactor core, the 308 L overlaying material is subjected to intense neutron irradiation damage during service [12]. Meanwhile, in the high-temperature, high-pressure water environment of the primary circuit, hydrogen is artificially injected to suppress the strongly oxidising conditions generated by radiolysis. Under the long-term synergistic action of irradiation damage and a hydrogen-rich atmosphere, the 308 L cladding is susceptible to hydrogen-promoted localised corrosion, particularly pitting corrosion, which is characterised by high concealment and marked detrimental effects [13–15]. Previous studies have shown that pitting corrosion can act as a primary crack initiation site [16], thereby facilitating stress corrosion cracking (SCC) in stainless steels and substantially increasing the risk of failure of in-service components [17,18]. Therefore, elucidating the pitting corrosion mechanism of the 308 L stainless steel overlayer under the combined influence of irradiation damage and hydrogen exposure is essential for mitigating the risk of subsequent corrosion-induced degradation during the crack incubation period of SCC.

Pitting is a highly localised form of corrosion, typically initiated by the local adsorption of aggressive anions on the material surface, followed by the breakdown of the passive film [19–21]. Abd El-Haleem *et al.* [20] demonstrated that chloride ions (Cl^-) gradually adsorb onto and penetrate the passive film, accumulating to a critical concentration that induces local rupture and initiates pitting. In the presence of hydrogen, interaction between hydrogen and the passive film degrades the protective performance of the latter, thereby increasing the pitting susceptibility of stainless steel [22–24]. Li *et al.* [25] reported that hydrogen initially alters the composition and characteristics of the passive film. During passivation, hydrogen elevates the $\text{OH}^-/\text{O}^{2-}$ ratio within the film, which substantially raises its electrical conductivity and consequently reduces its overall stability. Chloride ions in the electrolyte tend to replace OH^- , further compromising the local stability of the passive film and rendering it more susceptible to localised rupture, thus promoting pitting corrosion [26]. Guo [27] and Yang *et al.* [24] proposed that hydrogen facilitates the transformation of the passive film on duplex stainless steel from a p-type to an n-type semiconductor, the latter exhibiting higher pitting susceptibility. Guo *et al.* [28] performed immersion tests on stainless steel specimens with and without hydrogen pre-charging. Pronounced pits appeared on the hydrogen-precharged samples after only 6 days of immersion, whereas no significant pitting was observed on the non-charged specimens even after 30 days. These findings further confirm that hydrogen promotes pitting corrosion in stainless steel. Qiao's research team demonstrated that hydrogen capture ability varies with the activation energy of defect-related hydrogen traps, and this variation directly affects the occurrence of hydrogen-induced pitting corrosion [29]. In summary, nuclear-grade stainless steel subjected to extensive irradiation damage contains numerous defects, including dislocations and voids. Such defects are likely to act as hydrogen traps, thereby increasing the likelihood of hydrogen-induced pitting corrosion at elevated temperatures.

Studies have shown that defects such as grain boundaries, dislocations, voids and precipitates can act as hydrogen trapping sites [30–33], leading to locally elevated hydrogen concentrations. Hydrogen ingress significantly reduces the passivation of stainless steel, thereby inducing pitting corrosion and accelerating stress corrosion crack propagation [34]. Cui *et al.* [35] found that hydrogen accelerates the oxidation rate of the 308 L overlayer in deoxygenated high-temperature water. Moreover, phase boundaries and grain boundaries serve as trapping sites, where preferential oxidation occurs along grain boundaries and infiltration oxidation along phase boundaries. Li *et al.* [36] reported that cold work introduces a substantial number of dislocations in 2205 duplex stainless steel, which aggravates hydrogen-induced pitting corrosion. Chen *et al.* [37] directly observed significant hydrogen accumulation at dislocations, grain boundaries and precipitates using cryogenic transfer atom probe tomography (APT), confirming their role as effective traps and providing direct evidence for hydrogen trapping efficacy. Karpov *et al.* investigated hydrogen trapping in face-centred cubic metals and its relation to radiation-induced defects. They observed that hydrogen trapping occurs at 300–450 K via irradiation-induced dislocations; however, irradiation-induced He/Ar bubbles are more effective traps [38–41]. Furthermore, Karpov *et al.* [39] proposed that dislocation loops do not affect hydrogen transport, and that vacancies in the fcc structure are the primary hydrogen traps after irradiation. Therefore, to understand and control hydrogen-induced pitting corrosion in 308 L stainless steel, it is crucial to investigate hydrogen diffusion and its interaction with structural defects.

This study systematically elucidates the synergistic effects of irradiation-induced defects and hydrogen on the pitting susceptibility of austenitic stainless steel, and investigates how the spatial distributions of these defects and hydrogen jointly determine pit initiation sites and propagation paths. To achieve this, 308 L stainless steel was irradiated with 2 MeV protons to simultaneously introduce displacement damage and hydrogen, thus simulating the reactor service environment where irradiation and hydrogen enrichment coexist. A multimodal characterisation strategy was adopted, comprising atomic force microscopy (AFM) and scanning Kelvin probe force microscopy (SKPFM) to resolve surface topography and local electrochemical heterogeneity, hydrogen microprinting (HMT) to visualise hydrogen distribution, focused ion beam (FIB) milling and transmission electron microscopy (TEM) to examine the microstructures of irradiation-induced defects (i.e., hydrogen traps), and pitting immersion tests to identify preferential pit initiation sites. Based on these characterisations, the correlation between irradiation-induced defects and hydrogen distribution, along with their synergistic mechanism in pit initiation, is systematically discussed.

2. Materials and methods

2.1. Materials and sample preparation

The material selected for this study was 308 L austenitic stainless steel, a grade commonly used in nuclear power plants. Its nominal composition is (wt%): C 0.016, Si 0.32, Mn 1.33, P 0.015, Cr 19.80, Ni 10.30, and Fe balance. The microstructure exhibits a duplex phase of δ -ferrite dispersed in an austenite matrix (see Fig. 1), with a volume fraction of δ -ferrite in the range of 5–20% [42]. Specimens with dimensions of 20 mm $\times$ 3 mm $\times$ 2 mm were prepared by wire electrical discharge machining. The 20 mm $\times$ 3 mm face was designated for proton irradiation (see Fig. 2). The specimens were sequentially ground with SiC abrasive paper up to 5000 grit, then mechanically polished with 0.5 μm diamond paste. To remove the surface deformation layer, the specimens were further polished with a 40 nm silica suspension for 2 h. Subsequently, they were ultrasonically cleaned in ethanol for 15 min to eliminate residual silica particles. Finally, the specimens were dried with cool air and stored in a vacuum desiccator.

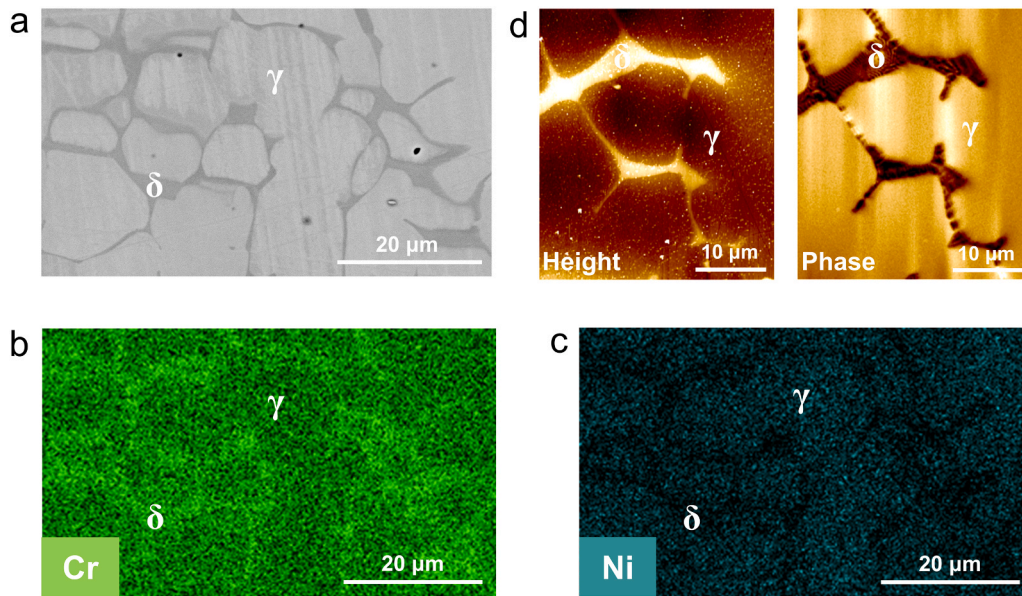


Fig. 1. (a) SEM image showing austenite and ferrite phases in 308 L; (b) Cr and (c) Ni EDS mapping results corresponding to (a); (d) AFM topography of ferrite and austenite phases and the corresponding MFM result.

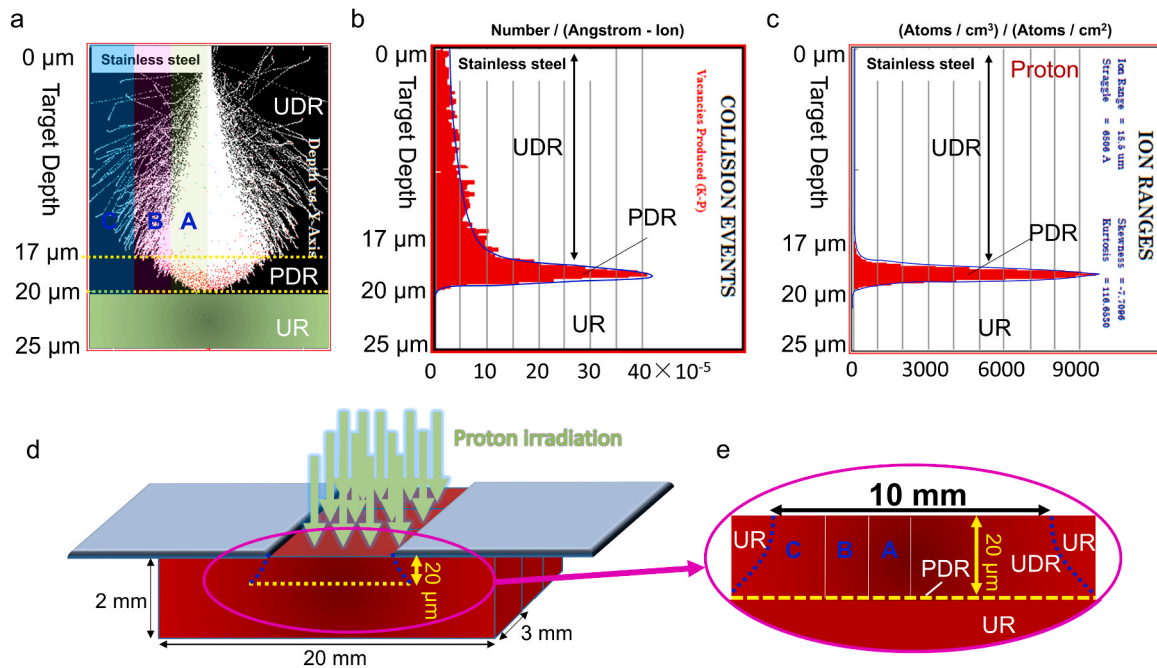


Fig. 2. (a) Ion trajectory profiles of 2 MeV hydrogen ions implanted into stainless steel. (b) Vacancy concentration as a function of depth. (c) Implanted hydrogen ion concentration as a function of depth. (d) Schematic diagram of the non-uniform distribution of hydrogen ions along the depth direction. (e) Enlarged image of the framed area in (d).

2.2. Proton irradiation experiment

Irradiation experiments were conducted at the Ion Beam Laboratory of the University of Michigan. A 20 mm × 3 mm area of the 308 L specimen was irradiated with 2 MeV protons generated by a 3 MeV accelerator at the Michigan Ion Beam Laboratory. Owing to fixation requirements at one end of the specimen, the effective irradiated area was centred at 10 mm × 3 mm. The displacement damage rate, calculated using the SRIM-2013 computer code [43], was 5.96×10^{-6} dpa/s. In the primary loop outside the reactor core, materials are typically subjected to neutron fluence ($E > 1$ MeV) corresponding to doses of

0.1–10 dpa over a 40-year service period. This damage range encompasses the majority of irradiation-assisted stress corrosion cracking (IASCC)-affected regions within the primary loop's multi-component system [44]. On the basis of the target total damage of 0.5 dpa, the required irradiation time at the given dose rate was approximately 23 h. During irradiation, the temperature was maintained at 360 ± 10 °C to simulate the ambient conditions of a nuclear reactor core. Under these irradiation conditions, a peak damage region is formed within the specimen, resulting in distinct irradiated and unirradiated zones in cross-section. Further details regarding the irradiation experiment are provided in reference [8].

A displacement threshold energy (E_d) of 40 eV, representing the energy required to form a stable Frenkel pair, was adopted for the SRIM simulations of stainless steel [45]. Based on the SRIM simulations (see Fig. 2a–c), the energy–depth relationship indicates that 2 MeV proton irradiation penetrates to a maximum depth of approximately 20 μm [46, 47]. The depth-dependent analysis of vacancy and injected proton profiles allows the entire damaged region to be divided into three zones: a uniformly damaged region (UDR, 0–17 μm) characterised by a relatively flat vacancy distribution (~ 0.5 dpa); a peak damaged region (PDR, 17–20 μm) where both vacancy and injected proton concentrations reach their maxima; and an unirradiated region (UR, beyond 20 μm) serving as a reference for evaluating changes in performance in the UDR and PDR (see Fig. 2b–e). As shown in Fig. 2a, upon proton injection into the stainless steel, the damage path is not vertical but rather forms a fan-shaped damaged region. Extending this point-wise irradiation across the entire irradiated surface (10 mm $\times$ 3 mm) introduces a transverse damage gradient in the irradiation cross-section—i.e., between the damaged and undamaged zones (see Fig. 2d and e). Theoretically, this region can be regarded as a transition zone from irradiated-damaged to unirradiated material. Although such a transition could be further subdivided into multiple zones (e.g., A, B, C, D, E, F...), the present study simplifies this gradient into three regions (A, B and C) for the convenience of subsequent characterisation of hydrogen distribution and corrosion behaviour (see Fig. 2e).

2.3. Electrochemical corrosion tests

A constant potential of 1.0 V was applied for 10 s to anodically corrode the sample in 30 vol% HNO_3 with a platinum sheet as the counter electrode, followed by observation of the corrosion morphology for SEM characterisation. To avoid surface oxidation after electrolytic corrosion, the corrosion morphology was immediately characterised by AFM and SEM. Additionally, to ensure that the local corrosion depth remained within the measurable range of AFM surface topography, the sample was subjected to electrochemical corrosion at 1.0 V for 3–5 s prior to AFM analysis of the irradiation-modified morphology.

2.4. SKPFM characterisation

Post-irradiation SKPFM measurements were performed using an atomic force microscope (Nanoscope V, Veeco Instruments Inc.). To predict hydrogen-induced irradiation damage and corrosion, SKPFM was employed to determine the local Volta potential, thereby enabling calculation of the electron work function. A Nanoprobe™ SCM-PIT conductive PtIr-coated silicon tip was used for SKPFM measurements, and the dual-scan mode allowed simultaneous acquisition of Volta potential and surface topography data. To measure the tip–sample voltage, an AC voltage was applied to the tip – which has a stable electron work function – rather than to the sample surface [48,49]. The potential was determined using the zero-potential method (also known as the Kelvin method). A direct measurement of the contact potential difference (V_{CPD}) between the tip and the sample was obtained via SKPFM. The conversion between the contact potential difference (V_{CPD}) and the electron work function is given by the following equation:

$$V_{\text{CPD}} = (\varphi_{\text{sample}} - \varphi_{\text{tip}}) / e \quad (1)$$

φ_{sample} and φ_{tip} indicate the work functions of the sample and the tip, and e is the electron charge. Given the tip work function (φ_{tip}), the sample's work function (φ_{sample}) can be calculated [50]. Moreover, calibration experiments were conducted on a pure Ni surface held at a stable potential.

2.5. Hydrogen-distribution characterisation

Hydrogen microprinting (HMT), which relies on conventional photographic emulsion development, enables effective detection of injected hydrogen distribution. The technique involves static reaction of metal specimens in an opaque silver bromide (AgBr) nitrite solution in the dark, where silver ions are reduced by hydrogen according to $\text{Ag}^+ + \text{H} \rightarrow \text{Ag} + \text{H}^+$ [51]. After formaldehyde treatment and rinsing, the silver particle distribution observed by scanning electron microscopy (SEM) indirectly characterises the hydrogen distribution within the specimen.

2.6. Irradiation defects characterisation

Focused ion beam (FIB, FEI Helios G4 CX) milling was used to prepare cross-sectional TEM specimens from irradiated regions, with the FIB targeted at hydrogen aggregates (silver particles). Subsequently, high-resolution characterisation of irradiation-induced defects associated with hydrogen accumulation was performed using a Spectra 300 S/TEM operated at 300 kV.

2.7. Pitting corrosion immersion experiment

To remove the surface deformation layer induced by mechanical processing, the specimens were polished for 2 h with a 40 nm silica suspension, yielding a nanoscale surface finish. Pitting corrosion tests were then conducted by immersing the specimens in a 6 wt% FeCl_3 solution at 25.0 ± 0.5 °C. The corrosion morphology was examined using scanning electron microscopy (SEM, Axia) in secondary electron mode, operated at an accelerating voltage of 15 kV, a beam current of 1.1 nA, and a working distance of 10 mm. The pitting density (number of pits per unit area) in different regions was systematically quantified using ImageJ software.

3. Results

3.1. Division of the study regions

Fig. 3 shows cross-sectional SEM images of the irradiated samples after electrochemical corrosion. A clear boundary between the irradiated and unirradiated regions is visible in these images (see Fig. 3c–e), which corresponds to the peak damage position in the SRIM profile (see Fig. 2b and c). Within the peak damage region, the damage depth is greater in the centre and shallower at the edges (see Fig. 3c–e). Consequently, the sample is horizontally divided into three regions (A, B and C), consistent with the simulated damage distribution from SRIM (see Fig. 2a).

3.2. Electron work function of 308 L different regions

Fig. 4 shows AFM and SKPFM data for regions A, B and C of the irradiated sample. Fig. 4a, c and e present the AFM topography of regions A, B and C, respectively, while Fig. 4b, d and f show the corresponding Volta potential images. It is evident from Fig. 4 that the V_{CPD} increases progressively from region A to region C. To compare the electron work function variations across the UDR, PDR and UR within each region, areas of 7.5 μm in length and 3 μm in width were selected to measure the contact potential difference and calculate the work function.

Fig. 5 presents the electron work function (EWF) profiles of regions A, B and C. The EWF is considered an indicator of the ease with which electrons can transfer from the Fermi level to the vacuum above the metal surface; a lower EWF thus implies a higher electron emission propensity [52]. In the context of metal corrosion, EWF can serve as a predictor of corrosion susceptibility [48]. As shown in Fig. 5a, the EWF increases progressively from region A to region C. To enable a more detailed comparison of the EWF values in the UDR, PDR and UR across

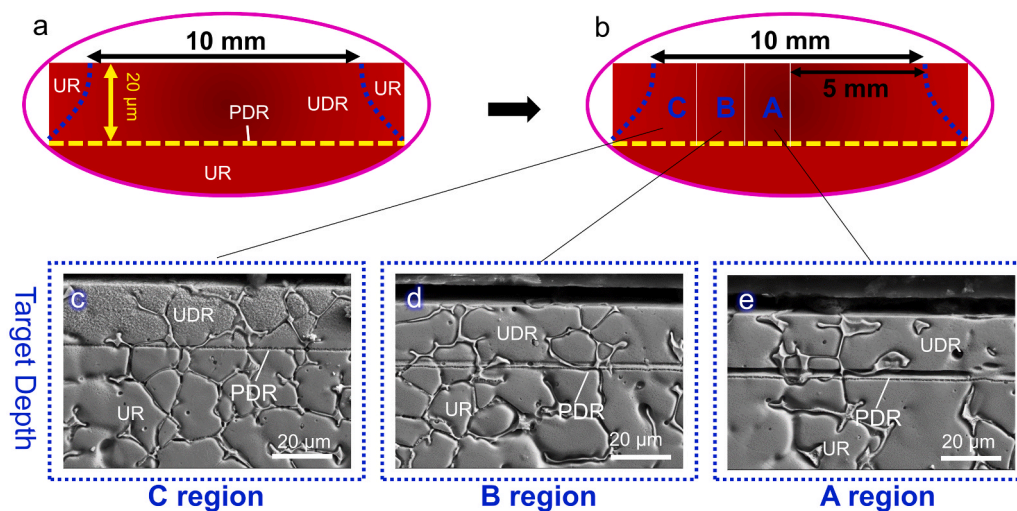


Fig. 3. (a, b) Schematic diagrams of irradiation-affected areas in the sample cross-section; (c–e) SEM images of electrochemical corrosion corresponding to regions C, B, A in (a).

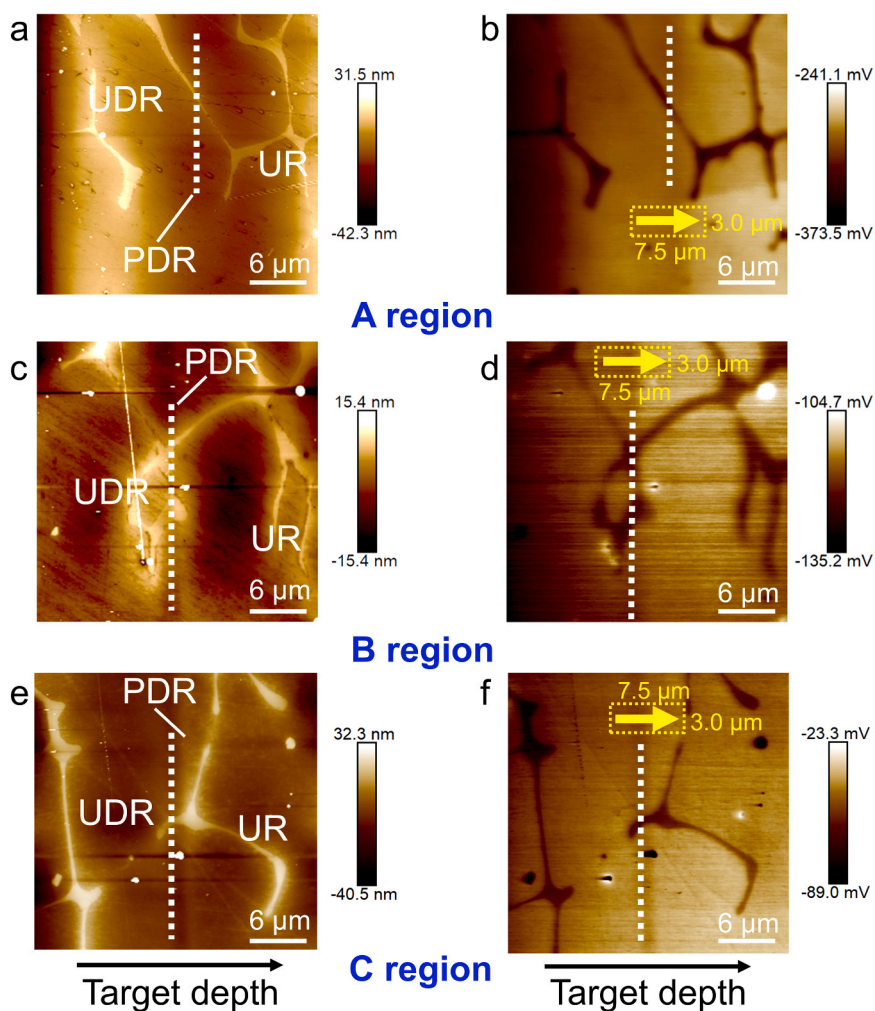


Fig. 4. 308 L AFM topography of each region and SKPFM potential diagram; (a,b) The morphology and potential diagram of A region; (c,d) The morphology and potential diagram of B region; (e,f) The morphology and potential diagram of C region.

the three regions, Fig. 5b–d present the depth-dependent EWF profiles for regions A, B and C, respectively. It is evident that, for all three regions, the EWF follows a consistent depth-dependent trend: PDR < UDR

< UR. Moreover, within the PDR, the EWF order is region A < region B < region C. Given that EWF can be used to evaluate corrosion tendency, the corrosion resistance in the PDR decreases progressively from region

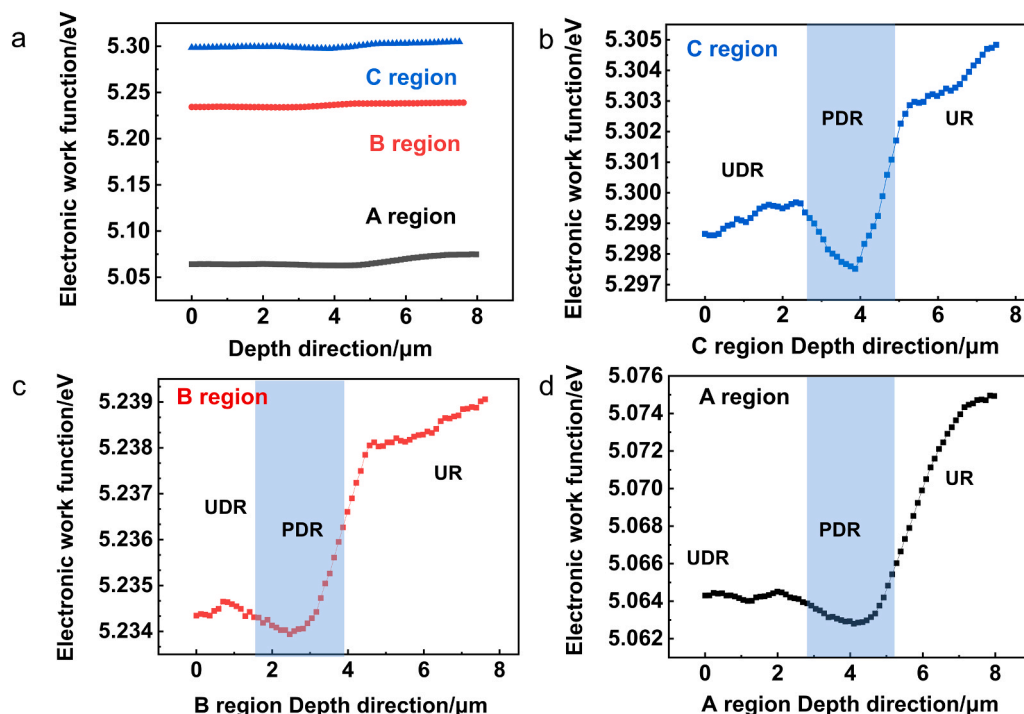


Fig. 5. The curves of electron work function in each region of 308 L; (a) Comparison of electron work functions in A, B, and C regions; (b) Details of the electron work function in region C of (a); (c) Details of the electron work function in region B of (a); (d) Details of the electron work function in region A of (a).

C to B to A. Furthermore, the SRIM simulation profiles (Fig. 2) indicate that the vacancy concentration reaches a maximum in the PDR, which correlates well with the corrosion resistance: a higher vacancy concentration corresponds to a lower EWF and poorer corrosion resistance. This correlation is further evidenced by the corrosion morphology of the PDR, as shown in Fig. 3.

3.3. Corrosion morphology and corresponding surface roughness in different regions

Given that the vertical range of AFM surface topography is limited to 1 μm, the electrochemical corrosion experiments were conducted at a reduced DC voltage (1.0 V) and a shorter duration (5 s). After corrosion, the surface morphology showed only minor changes, which are barely discernible in the SEM images (see Fig. 3) but detectable by AFM. Fig. 6 shows the corrosion morphologies of regions A, B and C obtained by

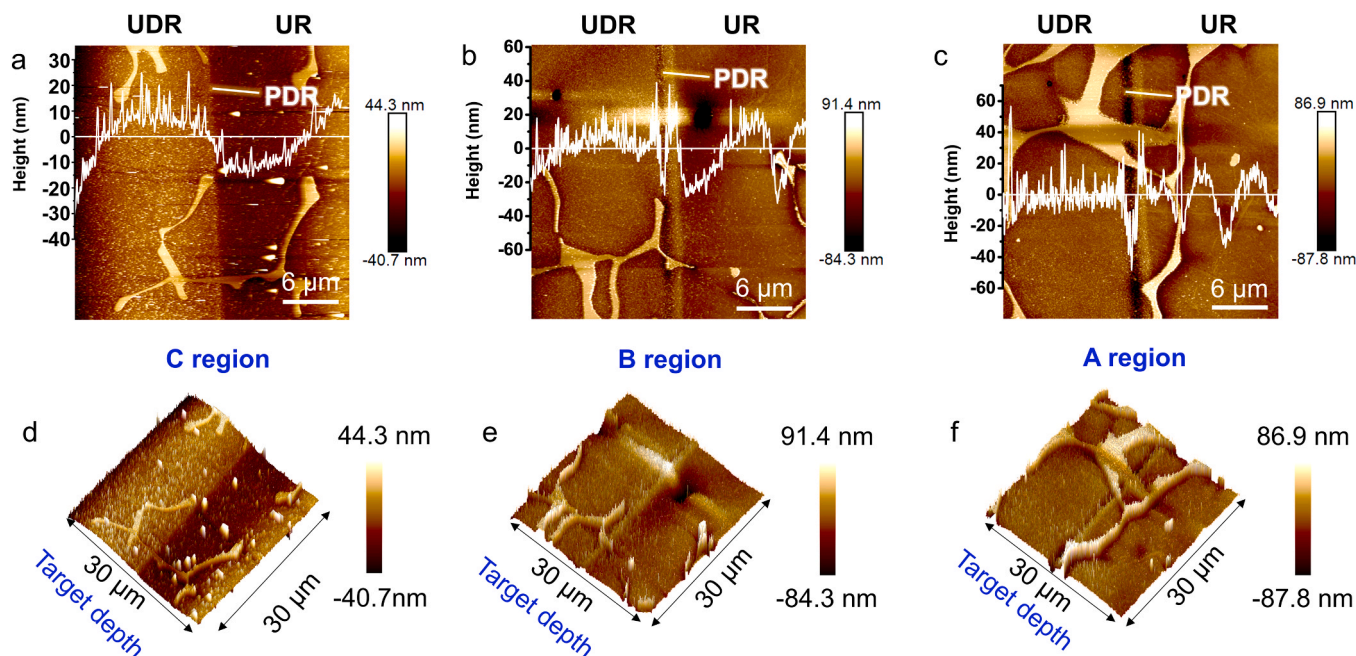


Fig. 6. AFM was used to characterize the 308 L etching morphology after electrochemical etching treatment; (a-c) C, B, A region two-dimensional topography and terrain change curve; (d-f) C, B, A region three-dimensional topography.

AFM. The corrosion depth was found to increase progressively from region C to B to A, indicating a decrease in corrosion resistance. This trend is consistent with the electron work function variations shown in Fig. 5b–d. The corrosion depth within the peak damaged region (PDR) reflects the extent of corrosion. In region C, a corrosion depression of approximately 20 nm in depth was observed in the PDR. In contrast, the depth profiles in the morphology maps reveal depression depths of about 30 nm and 50 nm for regions B and A, respectively.

Since the extent of corrosion in the uniform damaged region (UDR) and the unirradiated region (UR) within regions A, B and C (see Fig. 6) could not be determined from the morphology alone, the roughness (R_a) of the austenite phase in the UDR and UR was measured across the three regions. For each region, at least five areas of $3 \mu\text{m} \times 2 \mu\text{m}$ were selected for R_a calculation, and the results are presented as a histogram in Fig. 7. The data indicate that the R_a values of the austenite phase in regions A, B and C decrease progressively: first in the UDR, then further in the UR. Thus, based on the results in Fig. 7, the irradiated austenitic material exhibits a tendency towards corrosion. Moreover, the corrosion resistance from region C to region A shows a similar decreasing trend, which is fully consistent with the corrosion morphology observed in the peak damaged region (PDR).

3.4. Hydrogen distribution after proton irradiation

Hydrogen distribution was characterised using hydrogen microprinting (HMT) combined with scanning electron microscopy-energy dispersive spectroscopy (SEM-EDS). HMT reveals hydrogen-rich zones through the precipitation of silver particles, while SEM-EDS confirms the presence of silver via energy spectrum analysis. As shown in Fig. 8, along the depth direction, hydrogen enrichment is highest in the peak damaged region (PDR), followed by the uniform damaged region (UDR) and the unirradiated region (UR). In the transverse direction, the silver particle count decreases progressively from region A to B to C within the UDR, PDR and UR, with no particles detected in the UR of region C. To quantitatively evaluate the hydrogen content in different regions, three independent statistical analyses of silver particle density were performed on each of regions A, B and C in the 0.5 dpa specimen (see Fig. S1 in Supplementary materials). The statistical results are summarised in Fig. 9 and Supplementary Table S1. As shown in Fig. 9, the particle densities in the PDR from region A to C are $(15.18 \pm 1.26) \times 10^{10} \text{ m}^{-2}$, $(10.44 \pm 0.95) \times 10^{10} \text{ m}^{-2}$ and $(7.91 \pm 0.21) \times 10^{10} \text{ m}^{-2}$; in the UDR, the corresponding values are $(4.05 \pm 0.68) \times 10^{10} \text{ m}^{-2}$, $(3.77 \pm 1.05) \times 10^{10} \text{ m}^{-2}$ and $(1.84 \pm 0.74) \times 10^{10} \text{ m}^{-2}$.

3.5. SEM observation of pitting corrosion in different irradiation areas

Fig. 10 presents SEM micrographs of pitting initiation in regions A, B and C. After immersion for 1.5 h in 6 wt% FeCl₃ solution, the irradiated specimens exhibited pronounced regional selectivity in pitting

susceptibility. The peak damaged region (PDR) showed the highest pitting susceptibility and the highest density of pit nucleation, followed by the uniform damaged region (UDR), while the unirradiated region (UR) exhibited the best corrosion resistance (see Fig. 10). To quantitatively evaluate the pitting density across different regions, three independent statistical evaluations of pit density were performed on each of regions A, B and C in the 0.5 dpa specimen (see Fig. S2 in Supplementary materials). The statistical results are summarised in Fig. 11 and Supplementary Table S2. Notably, the pit density exhibited a lateral gradient (region A > region B > region C). In the PDR, the pit densities from region A to C were $(8.79 \pm 1.04) \times 10^{11} \text{ m}^{-2}$, $(6.17 \pm 0.29) \times 10^{11} \text{ m}^{-2}$ and $(3.91 \pm 0.44) \times 10^{11} \text{ m}^{-2}$; in the UDR, the corresponding values were $(1.04 \pm 0.36) \times 10^{11} \text{ m}^{-2}$, $(0.79 \pm 0.12) \times 10^{11} \text{ m}^{-2}$ and $(0.24 \pm 0.17) \times 10^{11} \text{ m}^{-2}$. These results are consistent with the hydrogen microprinting and SKPFM findings, suggesting a correlation between pitting and hydrogen distribution, with hydrogen-rich areas serving as preferential sites for pit initiation.

4. Discussion

Pitting corrosion mechanisms were investigated in the austenite, ferrite, and phase boundaries of 308 L stainless steel, with emphasis on irradiation damage arising from pressure vessel cladding and the combined influence of hydrogen in the primary coolant. Based on the results presented in Section 3, the discussion is structured into three parts: (1) the effect of irradiation defects on hydrogen trapping at elevated temperatures; (2) the synergistic effect of voids and hydrogen on pitting initiation and propagation; and (3) the influence of phase-boundary carbides on pitting corrosion.

4.1. Effect of irradiation defects on hydrogen capture under high temperature environment

The entrapment of hydrogen in metals is intrinsically governed by the presence of internal trapping sites. This trapping effect arises fundamentally from the interaction between the strain fields of crystal defects and those of interstitial hydrogen atoms [40]. Such defects include interstitial atoms, vacancies, dislocations, grain boundaries, secondary phases and voids [53,54]. Pressouyre *et al.* [55,56] classified hydrogen traps into reversible and irreversible types. According to current understanding, reversible traps include solute atoms, dislocations, low-angle grain boundaries and vacancies, whereas irreversible traps comprise TiC, MnS, Fe₃C, high-angle grain boundaries and phase boundaries [57–59]. However, this classification is not absolute, as the reversibility of traps is temperature-dependent. At elevated temperatures, hydrogen trapped in so-called irreversible traps at lower temperatures may be released, effectively rendering them reversible under such conditions [60]. Consequently, the binding energy of hydrogen traps is strongly influenced by temperature [61].

It is commonly assumed that deep traps at room temperature may release hydrogen upon exposure to elevated temperatures [60]. In the present work, however, transmission electron microscopy (TEM) and hydrogen microprinting (HMT) were combined to investigate hydrogen trapping by irradiation-induced voids in 308 L stainless steel under simulated pressurised water reactor (PWR) primary-loop conditions. The results demonstrate that irradiation-induced voids effectively act as hydrogen traps even at high temperatures (see Fig. 12). Lin *et al.* [58,59] reported TEM observations of numerous Frankel dislocation loops and large-sized voids in the austenitic phase of 308 L stainless steel after proton irradiation, with the sizes of both voids and loops increasing with irradiation dose (0.5, 1.5 and 3.0 dpa). In the current study, only the voids at doses of 0.5 and 3.0 dpa were characterised by TEM (see Fig. 13), and the observations are consistent with those of Lin *et al.* [62,63]. Therefore, based on the void data, we infer that the evolution of Frankel dislocation loops follows the same trend as reported by Lin *et al.* [62,63]. Furthermore, HMT on proton-irradiated 308 L austenite

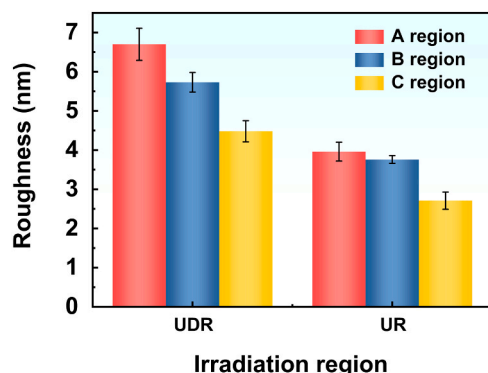


Fig. 7. Roughness histogram of UDR and UR in A, B, and C regions in Fig. 6.

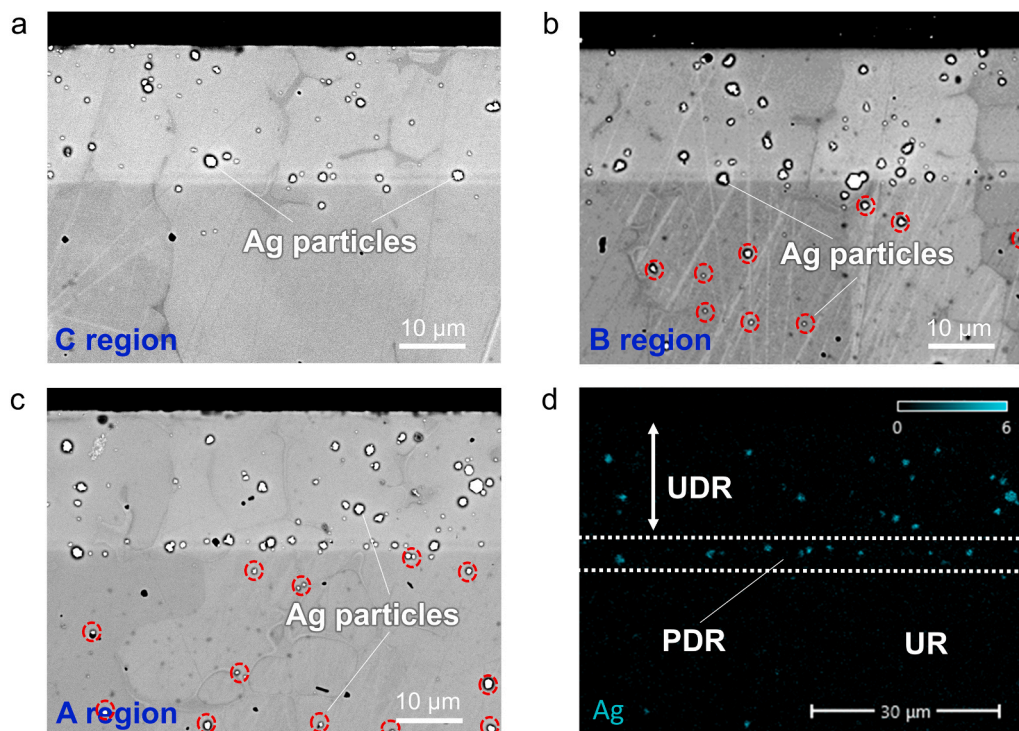


Fig. 8. (a–c) SEM images of Ag distribution in different regions (A, B, and C); (d) SEM-EDS mapping results of Ag elements corresponding to (c).

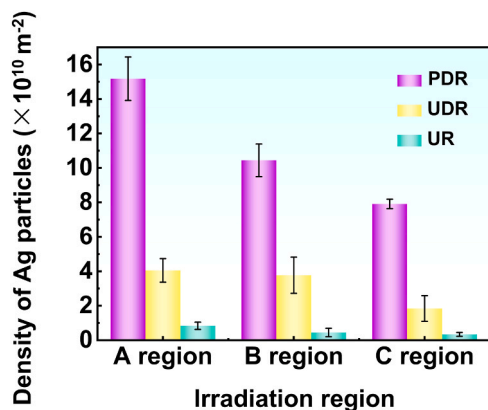


Fig. 9. Hydrogen micro-printing results of silver particle distribution histogram.

revealed that hydrogen introduced at high temperatures was captured by irradiation-induced defects and did not diffuse to the sample surface, despite the elevated temperature (see Fig. 12). In a related study, Anne-Cécile Bach *et al.* [64] combined experimental and modelling approaches to investigate hydrogen trapping by irradiation-induced defects in 316 L stainless steel. Their findings indicated that voids, rather than Frankel loops, constitute the principal trap type capable of capturing hydrogen at high temperatures. Thus, it is reasonably concluded that void defects generated by proton irradiation in 308 L stainless steel serve as the dominant hydrogen traps.

To further confirm that irradiation-induced voids act as effective hydrogen traps in austenitic stainless steel at elevated temperatures (290–320 °C), the void distribution in hydrogen-enriched regions within the peak damaged region of proton-irradiated specimens was examined (see Fig. 12). Specifically, a TEM lamella was extracted by focused ion beam (FIB) milling from the Ag-particle-decorated region of the HMT sample (see Fig. 12a, b). Subsequent HRTEM, EDS and morphological

analyses revealed a high density of irradiation-induced voids beneath the Ag particles (see Fig. 12c–f). Since the proton irradiation (i.e., hydrogen ion implantation) in this study was conducted at 360 ± 10 °C, these observations confirm that irradiation-induced voids serve as effective high-temperature hydrogen traps, consistent with the findings of Anne-Cécile Bach *et al.* [64].

4.2. Synergistic effect of voids and hydrogen on the initiation and propagation of pitting corrosion

The disparity in hydrogen trapping capacity among different regions is the fundamental cause of the non-uniform corrosion behaviour, consistent with previous reports [23]. Hydrogen enrichment promotes pitting, and as shown in Figs. 8 and 10, most pits are located in hydrogen-rich zones. The higher hydrogen density in the peak damaged region (PDR) results in a significantly higher pit density compared with that in the uniform damaged region (UDR). Moreover, from the centre to the edge in the horizontal direction, both hydrogen density (see Figs. 8 and 9) and pit density (see Figs. 10 and 11) decrease progressively. Thus, hydrogen enrichment not only induces pit initiation but also governs pit density. These observations clearly indicate that hydrogen increases the pitting susceptibility of 308 L stainless steel.

Qiao *et al.* demonstrated that in hydrogen-affected stainless steels, passive-film integrity governs pitting resistance, and Cl⁻-induced breakdown initiates pitting, which evolves through nucleation, metastable growth and stable propagation, with the metastable-to-stable transition controlled by local pH and cation enrichment to 70–80% saturation [65]. Hydrogen accelerates pitting via four concurrent mechanisms: chemical (interfacial acidification and Cl⁻ attraction to defects) [12], defect-chemical (increased OH⁻/O²⁻ ratio and oxygen vacancies, Cl⁻ substitution of hydroxyls, raising cation vacancy flux and causing film detachment) [66], electronic (proton-induced positive charging and semiconductivity reversal) [27], and mechanical (lattice distortion reducing Young's modulus and cohesion, thereby lowering the rupture threshold) [67]. In austenitic steels (304, 310, 316 L), hydrogen lowers the pitting potential and incubation time while increasing pit

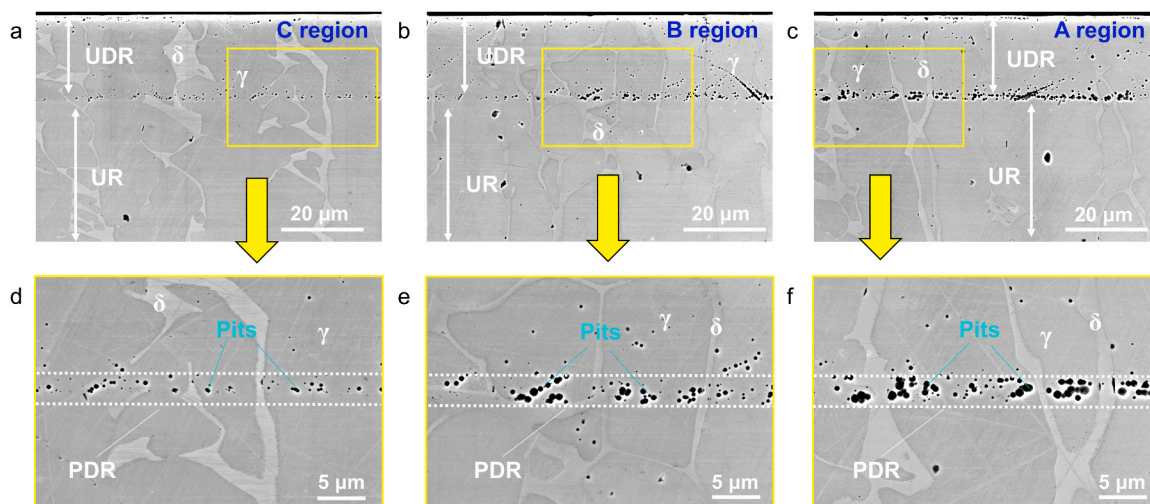


Fig. 10. SEM images of pitting corrosion after immersion tests; (a–c) Pitting morphologies in regions C, B and A, respectively; (d–f) Magnified views of the yellow-boxed areas in (a–c).

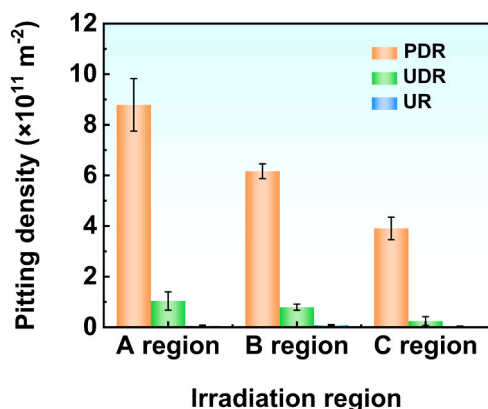


Fig. 11. Pitting density of different regions (PDR, UDR, and UR) in A, B, and C region.

density, although nitrogen alloying (~ 0.08 wt%) mitigates this effect [65,68]; in duplex steels (2205, 2507), hydrogen raises the passive current and decreases the potential, with preferential pitting at phase boundaries or ferrite – sites of low Volta potential and hydrogen enrichment, as confirmed in weld joints [33,69]. Overall, hydrogen synergises with Cl^- and microstructural heterogeneity to facilitate pit initiation and accelerate selective degradation.

In this study, the substantial number of irradiation voids formed within the austenite phase of 308 L stainless steel endows the material with a significant hydrogen trapping capacity for the implanted hydrogen. Accordingly, the void distribution is first characterised, followed by an examination of the influence of these irradiation-induced voids on the corrosion resistance of the passive film. Furthermore, on the basis of the hydrogen-assisted pitting mechanism proposed by Qiao *et al.*, the pitting corrosion behaviour of stainless steel under the coupled effect of irradiation voids and hydrogen is systematically discussed.

Transmission electron microscopy (TEM) was used to examine the austenite phase in the uniform damaged region (UDR), peak damaged region (PDR) and unirradiated region (UR) along the depth direction of the specimen irradiated to 3 dpa (see Fig. 13a–c). The results show that the PDR contains the largest voids in terms of both size and number density, followed by the UDR, while no irradiation-induced voids are detected in the UR (see Fig. 13a–c). On the basis of this depth-dependent trend, a similar distribution pattern is inferred for regions A, B and C across the irradiated cross-section. To quantitatively assess the

variations in void number density and size among regions A, B and C, statistical analyses were performed on data from the 0.5 dpa and 3.0 dpa 308 L specimens (UDR) (see Fig. 13d–i and Fig. 14.). For each region, at least three areas of $150 \text{ nm} \times 150 \text{ nm}$ were selected to determine the void number density and void size (see Figs. S3 and S4 in Supplementary materials). The void number densities are presented in Fig. 14 and Supplementary Table S3, while the void sizes are shown in Supplementary Fig. S5. As illustrated in Supplementary Figs. S3–S5 and Fig. 14, the void size at 3.0 dpa is larger than that at 0.5 dpa. However, for both doses, no significant variation in void size is observed among regions A, B and C. Notably, the areal number density of voids decreases progressively from region A to C. In the uniform damaged region (UDR), the void number densities from region A to region C are $(1.20 \pm 0.22) \times 10^{15}$, $(0.43 \pm 0.03) \times 10^{15}$ and $(0.31 \pm 0.08) \times 10^{15} \text{ m}^{-2}$ for the 0.5 dpa specimen, and $(4.81 \pm 0.84) \times 10^{15}$, $(2.98 \pm 0.72) \times 10^{15}$ and $(1.26 \pm 0.20) \times 10^{15} \text{ m}^{-2}$ for the 3.0 dpa specimen.

As reported in our earlier work [8], proton irradiation of 308 L stainless steel not only generates a high density of voids capable of trapping implanted hydrogen ions, but also induces compositional segregation around these voids. Consistent with the findings of Lin *et al.* [70], irradiation causes Cr depletion accompanied by enrichment of Ni, Si and P at interfaces such as grain boundaries, phase boundaries, precipitate interfaces and dislocation loops. Our previous results further confirm that irradiation-induced voids produce heterogeneous Cr distribution, with pronounced Cr depletion at specific void interfaces [8]. From a corrosion perspective, such Cr-depleted zones compromise the protectiveness of the passive film, thereby favouring preferential attack by chloride ions and initiating pitting corrosion [71].

In addition to Cr depletion at void peripheries, void-induced local swelling also warrants consideration [72]. Larger voids contain more vacancies, leading to significant stress concentration in the surrounding matrix [73]. Such stress concentrations have been shown to induce dense dislocation tangles and localised plastic deformation [74]. These microstructurally altered regions attract aggressive ions, thereby accelerating pitting initiation [75]. Compared with Frankel dislocation loops in irradiated 308 L austenite, irradiation voids possess larger strain fields, making them more effective hydrogen traps at elevated temperatures and accounting for their dominant role in hydrogen capture [64].

Notably, Lin *et al.* [63,76] investigated oxide film formation on ion-irradiated nuclear-grade austenitic stainless steel in simulated high-temperature, high-pressure primary water. They observed that the number density of irradiation defects (voids and Frankel loops) increased with irradiation dose, accompanied by oxide film thickening.

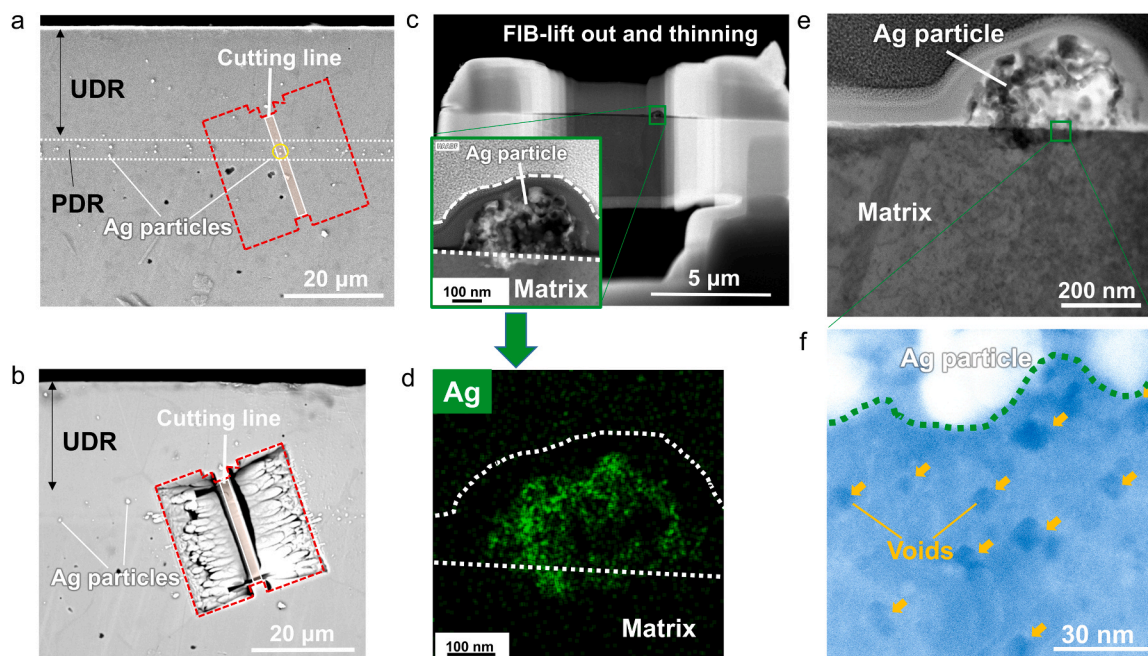


Fig. 12. FIB sample preparation and corresponding STEM analysis were performed on specific areas after the hydrogen micro-printing test. (a) The SEM morphology of the Ag particle position (hydrogen aggregation region) of the sample after the HMT test; (b) The SEM morphology of the silver particle position (hydrogen aggregation site) in the peak damage area after FIB sample preparation; (c) TEM observation of Ag particles on the surface of the sample (the green frame selected area enlarged) after FIB sample preparation; (d) The EDS mapping result of Ag particles in the green box in (c); (e) The dark field image of Ag particles in the FIB sample; (f) TEM results of irradiation void distribution under Ag particles, which was the area in the green box region shown in (e).

Lin *et al.* [63] attributed this to irradiation-induced defects enhancing inward oxygen transport and accelerating outward diffusion of metal atoms (Fe, Cr and Ni). In a related study, Jiang *et al.* [8] used current-sensing atomic force microscopy (CSAFM) to evaluate the corrosion resistance of oxide films formed in different irradiation damage zones under ambient conditions. Although irradiation voids and Frankel loops promoted oxide film growth, they ultimately impaired its corrosion resistance—owing to void-induced elemental segregation (Fe, Cr and Ni) and their role as hydrogen traps, both of which were linked to performance degradation by Jiang *et al.* [8].

Given that irradiation generates a high density of voids in the 308 L austenitic phase (see Fig. 13, and Fig. S6. in Supplementary materials) and that these voids act as high-temperature hydrogen traps, the synergistic effect of voids and trapped hydrogen on pitting corrosion warrants further discussion. The void and hydrogen distribution data in Fig. 12 confirm that irradiation voids immobilise hydrogen in specific regions, particularly in the PDR and parts of the UDR. Correlation with the pitting morphology (see Fig. 10a–f) reveals that pitting preferentially occurs in the austenitic phase, consistent with the hydrogen distribution shown in Fig. 8. Regardless of the sub-region (A, B or C), pits formed in the PDR are notably larger than those in the UDR (see Fig. 10). Moreover, pits in the PDR frequently exhibit an elongated, strip-like morphology, which is attributed to the high number density and large size of voids in this region (see Fig. 13a–c).

To elucidate the mechanism of pitting corrosion in 308 L stainless steel induced by the combined action of irradiation voids and hydrogen, a schematic diagram is presented in Fig. 15. Previous studies have shown that the synergistic effect between hydrogen and Cl^- ions can be interpreted within the framework of the Point Defect Model (PDM). According to this model, when the metal surface is exposed to a Cl^- -containing corrosive environment, hydrogen adsorbed at void sites can penetrate the passive film and exist as protons, creating a locally positively charged region. This region then attracts Cl^- ions from the surrounding environment via electrostatic interaction (see Fig. 15b), leading to localised Cl^- enrichment [68,77–79]. As illustrated in Fig. 15c, owing to the strong Coulombic interaction between protons

and oxygen ions, the incorporated protons can react with oxygen ions within the passive film to form OH^- species [80,81].



According to the model, the progressive accumulation of H^+ within the oxide film increases OH^- content, thereby raising the $\text{OH}^-/\text{O}^{2-}$ ratio and accelerating ion migration [66]. Accumulated Cl^- can readily displace OH^- or H_2O molecules in the passive film. Consequently, vacancies concentrate at the metal/film interface, ultimately forming holes that serve as nucleation sites for pitting corrosion [66]. Li *et al.* [82] further demonstrated that hydrogen alters the semiconductor properties of the passive film. Electrons generated by hydrogen ionisation act as donors, converting the passive film from p-type to n-type semiconductor [27,83]. This transition degrades the corrosion resistance of the oxide film, rendering it more susceptible to attack by aggressive ions.

As evidenced by the observed distributions of voids and hydrogen (see Fig. 15a), irradiation-induced voids promote the growth of the passive film in regions where both voids and trapped hydrogen coexist. However, the presence of hydrogen simultaneously increases the hydroxide content within the film, thereby reducing its overall corrosion resistance [66]. Consequently, when exposed to chloride-containing solutions, such regions become more susceptible to localised attack, leading to preferential pitting (see Fig. 15b). In the uniform damaged region (UDR), the lower number density and smaller size of voids, compared with those in the peak damaged region (PDR), result in a reduced hydrogen trapping capacity and a corresponding decrease in pitting density (see Figs. 8–14 and Fig. 15a, c, d). Moreover, the larger void size and higher void density in the PDR promote hydrogen enrichment and facilitate the coalescence of individual pits, forming elongated pitting morphologies, as shown in Fig. 10d–f and schematically illustrated in Fig. 15d.

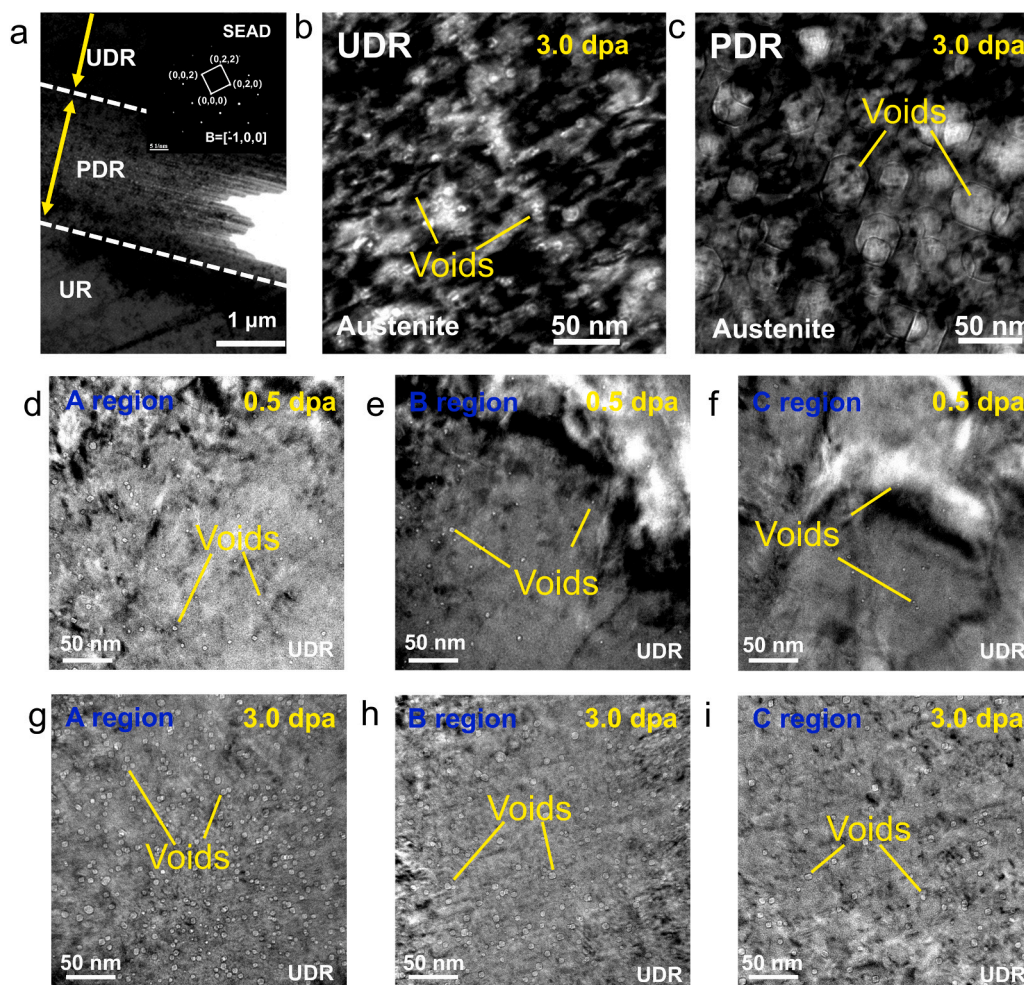


Fig. 13. (a) The UDR, PDR, and UR regions of FIB samples were observed by TEM for 3.0 dpa sample; (b, c) TEM characterization of irradiation-induced voids in the uniform damaged region (UDR) and peak damaged region (PDR) along the depth direction of the 3.0 dpa irradiated sample; (d–f) TEM images showing variations in void morphology within the UDR of 0.5 dpa 308 L samples from regions A, B, and C, respectively; (g–i) Corresponding TEM observations of irradiation voids in the UDR of 3.0 dpa 308 L samples from regions A, B, and C.

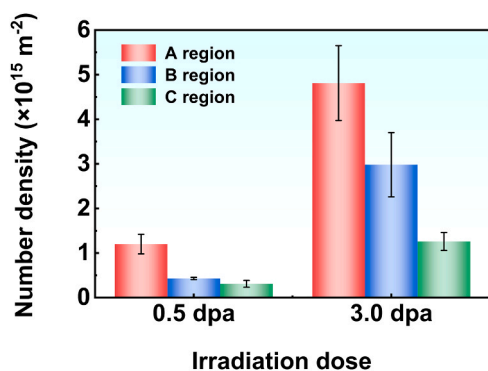


Fig. 14. Comparison of the number of irradiation-induced voids in regions A, B, and C (UDR) of the 308 L sample at 0.5 dpa and 3.0 dpa.

4.3. Effect of phase interface carbides on pitting corrosion of stainless steel in high temperature environment

In this study, pitting corrosion was predominantly observed in the austenite phase, although a few pits were also found at some phase boundaries (see Fig. 10b, d, f). Previous studies indicate that in duplex stainless steels, hydrogen tends to segregate at ferrite–austenite phase

boundaries, mainly due to differences in hydrogen solubility and diffusivity between the two phases [12]. However, under the high-temperature conditions ($360 \pm 10^\circ\text{C}$) in this work, the hydrogen distribution revealed no significant enrichment at these boundaries.

Li et al. [12] investigated the pitting susceptibility of 321 austenitic stainless steel weld joints after six years of service in a hydrocracking unit. They reported that in the as-received condition, hydrogen preferentially accumulated at the α/γ phase boundaries, where pitting initiated. However, after dehydrogenation heat treatment at 300°C for 10 h, hydrogen diffused more rapidly from the phase boundaries than from the austenite matrix. Consequently, pitting nucleation shifted from the phase boundaries to the γ -austenite phase, suggesting that under pressurised water reactor (PWR) operating conditions ($290\text{--}320^\circ\text{C}$), phase boundaries do not serve as effective hydrogen trapping sites. Consistent with this high-temperature behaviour, the pitting observed at certain phase boundaries (see Fig. 10) is likely not due to hydrogen enrichment, but rather to factors such as local chemical heterogeneity, interfacial energy or strain concentration at these interfaces [84–86].

AFM analysis of the corroded samples revealed that pitting corrosion preferentially occurred at phase boundaries decorated with 200–300 nm particles (see Fig. 16). As previously reported, these nanoparticles were identified as carbides precipitated at the phase boundaries [87]. To confirm that pitting initiation is associated with carbide-containing interfaces, a FIB lift-out technique was used to extract a TEM specimen from a pit site at a phase boundary (see Fig. 17). TEM characterisation

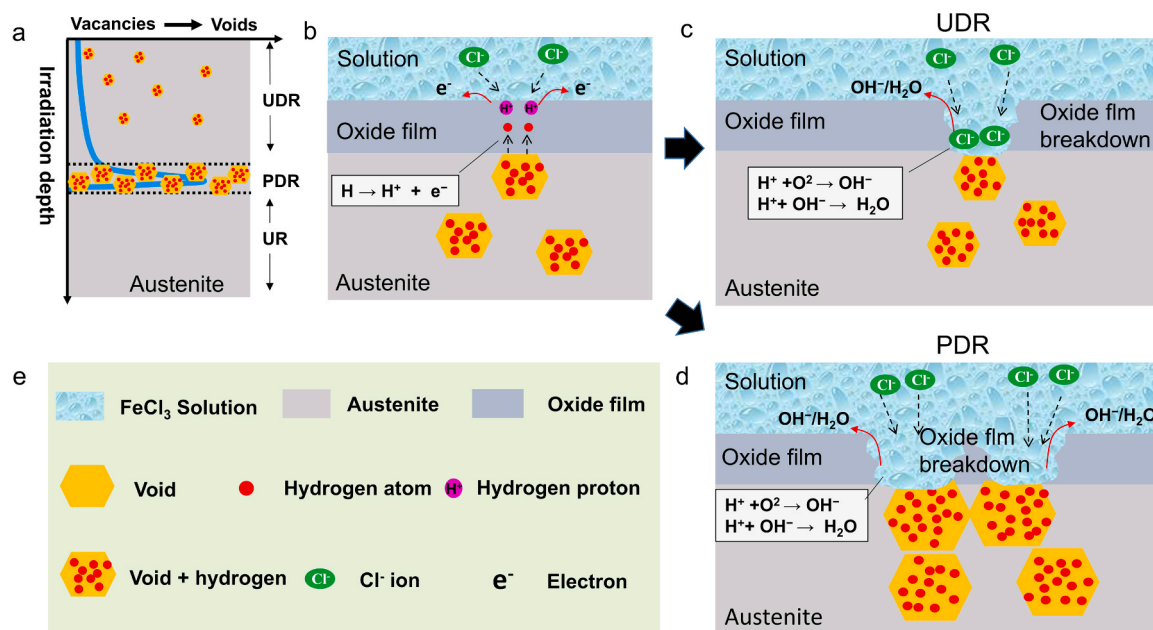


Fig. 15. The mechanism diagram of pitting corrosion under the coupling of irradiation voids and hydrogen. (a) The relationship between void distribution and SRIM simulation curve; (b) Initial nucleation stage of pitting corrosion; (c) The mechanism of irradiation voids and hydrogen-induced pitting corrosion in the UDR region; (d) The mechanism of irradiation voids and hydrogen-induced pitting corrosion in the PDR region; (e) The specific meaning corresponding to each icon in the mechanism diagram.

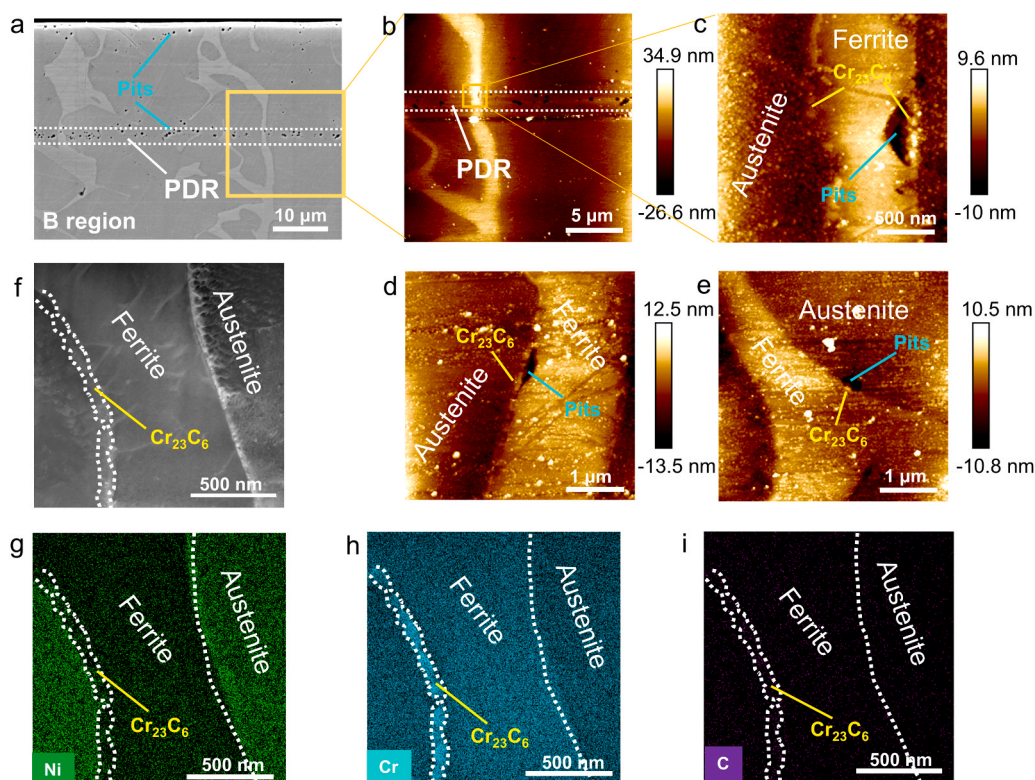


Fig. 16. (a–e) SEM and AFM morphologies of pitting corrosion at the two-phase interface (including carbides) in the PDR and UDR; (f) TEM image showing the distribution of carbides at the two-phase interface; (g–i) EDS mapping results of Ni (g), Cr (h), and C (i) corresponding to the TEM image in (f).

confirmed the presence of carbides directly beneath the corrosion pit (see Fig. 17). In contrast, no pitting was observed at carbide-free phase boundaries.

Extensive studies have shown that second-phase particles significantly influence the pitting resistance of stainless steels [88,89]. The

precipitation of Cr-rich carbides at phase and grain boundaries induces the formation of adjacent Cr-depleted zones, which in turn leads to a less protective passive film and thereby facilitates pitting initiation [90–93]. Consistent with this mechanism, energy-dispersive X-ray spectroscopy (EDS) analysis in Fig. 17 reveals a discernible Cr-depleted zone along the

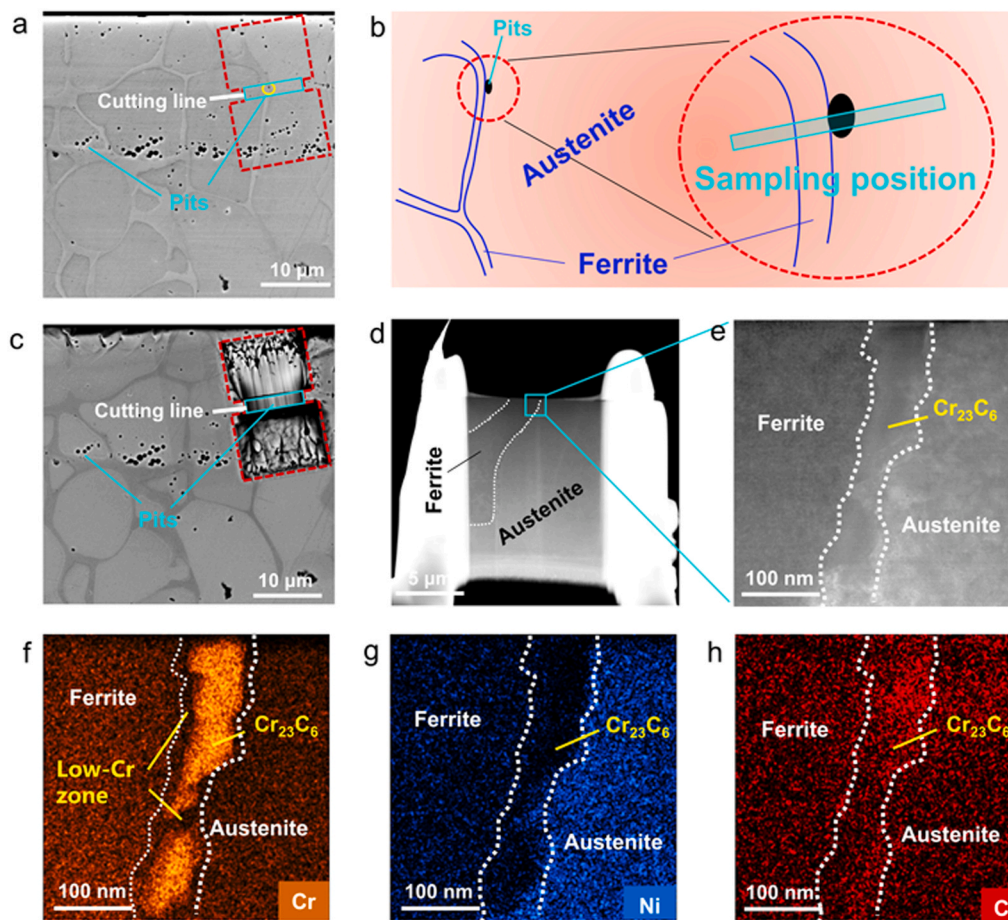


Fig. 17. (a) SEM morphology of pitting corrosion pits at the austenite/two-phase interface in the UDR; (b) Schematic diagram of FIB preparation of TEM samples at the two-phase interface containing corrosion pits; (c) SEM morphology after FIB sampling (corresponding to (a)); (d) TEM morphology of the FIB-prepared sample; (e) Enlarged TEM image of the framed region in (d); (f–h) EDS mapping results of Cr (f), Ni (g), and C (h) corresponding to (e).

carbide-containing phase boundary, in agreement with the findings of Cui *et al.* [35].

Lin *et al.* conducted constant extension rate tensile tests on proton-irradiated 308 L stainless steel weld metal in simulated PWR primary water and observed severe corrosion along phase boundaries containing a high density of carbides, with no deformed microstructures [94], suggesting that localised corrosion at carbide-decorated phase interfaces may play a critical role in irradiation-assisted stress corrosion cracking (IASCC) propagating along such boundaries. Moreover, several studies have emphasised the importance of preferential oxidation at internal interfaces in stainless steels for crack initiation, propagation and final failure [95,96].

Notably, Lin *et al.* also reported higher local strain at carbide-containing phase boundaries than at carbide-free interfaces, providing an additional factor for preferential pitting at these sites [94]. Such strain is known to facilitate irradiation-accelerated stress corrosion cracking, as demonstrated by McMurtrey *et al.* [97]. Furthermore, Bing *et al.* [98] established a close correlation between pitting and crack growth, capturing the entire evolution from pit nucleation to crack propagation via phase-field modelling.

On the basis of the above findings, in the uniform damaged region (UDR) of irradiated 308 L, the pitting probability in the austenite matrix is comparable to that at phase boundaries. In contrast, in the peak damaged region (PDR), pitting occurs predominantly within the austenite phase, with substantially fewer pits at phase boundaries (see Fig. 10). Accordingly, under external stress, crack nucleation in the UDR is expected to initiate at carbide-containing phase boundaries—consistent with the observations of Lin *et al.* [94]. In the

PDR, however, cracks are more likely to nucleate within the austenite phase itself. This difference is attributed to the higher irradiation dose and greater hydrogen enrichment in the PDR (see Figs. 8–9). Further investigation, such as in-situ stress tests on proton-irradiated samples at various doses, is needed to validate this hypothesis [99,100].

In summary, the synergy between irradiation-induced defects and hydrogen enrichment promotes selective pitting, with pits preferentially initiating in hydrogen-rich regions. The early-stage breakdown of the passive film is attributed to a coupled mechanism: rapid hydrogen diffusion through irradiation-induced voids and subsequent film weakening by accumulated hydrogen, which alters the film properties and attracts aggressive Cl^- ions. Consequently, hydrogen-enriched zones serve as preferred nucleation sites. Overall, this study demonstrates that irradiation damage and hydrogen enrichment act synergistically to promote pitting initiation.

5. Conclusion

308 L stainless steel was implanted with 2 MeV H^+ ions at 360 ± 10 °C to simulate neutron-irradiation-induced defects and investigate their interaction with hydrogen in pitting corrosion. TEM characterisation revealed the irradiation-induced voids and their depth distribution. Hydrogen microprinting (HMT) and immersion tests were subsequently combined to analyse the synergistic effect of voids and localised hydrogen enrichment on pit initiation. The main conclusions are as follows:

(1) Under high-temperature irradiation at 360 ± 10 °C, hydrogen is preferentially trapped by irradiation-induced voids rather than being

uniformly distributed. HMT analysis indicates that the trapping efficiency increases with both void size and number density.

(2) Post-irradiation, hydrogen accumulates at irradiation-induced voids, locally destabilising the passive film and thus creating preferential sites for pit initiation. This provides direct experimental evidence that irradiation-induced defects promote hydrogen-assisted localised corrosion.

(3) Pitting susceptibility varies with the spatial distribution of irradiation damage. In the depth direction, the extent of pitting decreases in the order: peak damaged region (PDR) > uniform damaged region (UDR) > unirradiated region (UR). Laterally, corrosion severity decreases progressively from the centre of the irradiated zone towards the periphery. This heterogeneous corrosion behaviour correlates with the distributions of irradiation-induced voids and hydrogen.

(4) Irradiation-induced voids serve both as hydrogen traps and as promoters of passive-film breakdown. Under the combined action of irradiation and hydrogen, the coupling of defects and hydrogen accelerates pit initiation – a critical consideration for the integrity assessment of nuclear structural materials.

CRediT authorship contribution statement

Yang He: Visualization, Methodology, Formal analysis. **Lijie Qiao:** Writing – review & editing, Supervision, Methodology, Formal analysis. **Weiting Yang:** Writing – review & editing, Validation, Supervision, Investigation, Funding acquisition. **Xiaohuan Du:** Writing – original draft, Investigation, Data curation. **Zhijie Jiao:** Visualization, Supervision, Formal analysis, Data curation. **Hua Yang:** Writing – review & editing, Validation, Formal analysis, Data curation. **Dongjiu Zhang:** Visualization, Methodology, Data curation. **Aiying Wang:** Writing – review & editing, Supervision, Investigation, Formal analysis. **Zhenyu Wang:** Writing – review & editing, Methodology, Investigation, Data curation. **Weichen Zhang:** Validation, Investigation, Data curation. **Binglu Zhang:** Supervision, Investigation, Data curation. **Huafeng Liu:** Supervision, Methodology, Data curation. **Yida Deng:** Writing – review & editing, Visualization, Supervision, Investigation, Funding acquisition, Formal analysis. **Baolong Jiang:** Writing – review & editing, Writing – original draft, Supervision, Investigation, Funding acquisition, Conceptualization.

Declaration of Competing Interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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Appendix A. Supporting information

Supplementary data associated with this article can be found in the online version at [doi:10.1016/j.corsci.2026.114115](https://doi.org/10.1016/j.corsci.2026.114115).

Data availability

Data will be made available on request.

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