Radiation Tolerance of Neutron-Irradiated Model Fe-Cr-Al Alloys[†]

Kevin G. Field^{1*}, Xunxiang Hu², Kenneth C. Littrell³, Yukinori Yamamoto⁴, Lance L. Snead⁵

^{1,2,3,4,5} Oak Ridge National Laboratory, Oak Ridge, TN 37831, USA

¹ fieldkg@ornl.gov

² hux1@ornl.gov

³ littrellkc@ornl.gov

⁴ yamamotoy@ornl.gov

⁵ sneadll@ornl.gov

***Corresponding author:**

Kevin G. Field
Materials Science and Technology Division
PO Box 2008
Oak Ridge, TN 37831 (USA)
Phone: +1 865 241 5623
Email: fieldkg@ornl.gov

[†]*This manuscript has been authored by UT-Battelle, LLC under Contract No. DE-AC05-00OR22725 with the U.S. Department of Energy. The United States Government retains and the publisher, by accepting the article for publication, acknowledges that the United States Government retains a non-exclusive, paid-up, irrevocable, world-wide license to publish or reproduce the published form of this manuscript, or allow others to do so, for United States Government purposes. The Department of Energy will provide public access to these results of federally sponsored research in accordance with the DOE Public Access Plan (<http://energy.gov/downloads/doe-public-access-plan>).*

Radiation Tolerance of Neutron-Irradiated Model Fe-Cr-Al Alloys[†]

Kevin G. Field^{1*}, Xunxiang Hu², Kenneth C. Littrell³, Yukinori Yamamoto⁴, Lance L. Snead⁵

^{1,2,3,4,5} Oak Ridge National Laboratory, Oak Ridge, TN 37831, USA

¹ fieldkg@ornl.gov

² hux1@ornl.gov

³ littrellkc@ornl.gov

⁴ yamamotoy@ornl.gov

⁵ sneadll@ornl.gov

Abstract

The Fe-Cr-Al alloy system has the potential to form an important class of enhanced accident-tolerant cladding materials in the nuclear power industry owing to the alloy system's higher oxidation resistance in high-temperature steam environments compared with traditional zirconium-based alloys. However, radiation tolerance of Fe-Cr-Al alloys has not been fully established. In this study, a series of Fe-Cr-Al alloys with 10–18 wt % Cr and 2.9–4.9 wt % Al were neutron irradiated at 382°C to 1.8 dpa to investigate the irradiation-induced microstructural and mechanical property evolution as a function of alloy composition. Dislocation loops with Burgers vector of $a/2 \langle 111 \rangle$ and $a \langle 100 \rangle$ were detected and quantified. Results indicate precipitation of Cr-rich α' is primarily dependent on the bulk chromium composition. Mechanical testing of sub-size-irradiated tensile specimens indicates the hardening response seen after irradiation is dependent on the bulk chromium composition. A structure–property relationship was developed; it indicated that the change in yield strength after irradiation is caused by the formation of these radiation-induced defects and is dominated by the large number density of Cr-rich α' precipitates at sufficiently high chromium contents after irradiation.

Keywords: FeCrAl, Accident Tolerant, Phase Stability, Dislocation, Hardening

[†]This manuscript has been authored by UT-Battelle, LLC under Contract No. DE-AC05-00OR22725 with the U.S. Department of Energy. The United States Government retains and the publisher, by accepting the article for publication, acknowledges that the United States Government retains a non-exclusive, paid-up, irrevocable, world-wide license to publish or reproduce the published form of this manuscript, or allow others to do so, for United States Government purposes. The Department of Energy will provide public access to these results of federally sponsored research in accordance with the DOE Public Access Plan (<http://energy.gov/downloads/doe-public-access-plan>).

1. Introduction

Fe-Cr-Al alloys are currently under consideration as a replacement alloy for fuel cladding in light water reactors (LWRs) owing to their higher oxidation resistance in high-temperature steam environments compared with zirconium-based alloys [1, 2]. The increased oxidation resistance has the potential to enhance the overall safety margins and accident tolerance of the current LWR fleet [3, 4]. Current research efforts have focused on developing nuclear-grade Fe-Cr-Al alloys [5-7]. These Fe-Cr-Al alloys are being developed to balance the chromium and aluminum solute additions to optimize the aqueous corrosion and high-temperature oxidation resistance [1, 2, 5, 8] while maintaining acceptable properties for fuel-cladding applications such as formability [5], mechanical properties [5, 9], weldability [10-12], neutronics [13], and radiation tolerance [14-16].

Traditional Fe-Cr-Al alloys targeted for high-temperature oxidizing environments, such as PM2000™ or APMT™, use high (i.e., >15 wt %) chromium solute additions, which have been known to exhibit phase instability in typical LWR clad operating temperatures [9, 17-25]. This phase instability is the result of the miscibility gap in the body-centered-cubic (BCC) Fe-Cr system [26-28] and has been indicated to exist within the BCC Fe-Cr-Al ternary system as well [9, 17-25]. High-chromium-content Fe-Cr and Fe-Cr-Al decomposes into the Fe-rich α and Cr-rich α' phases. The formation of the Cr-rich α' phase during thermal aging has been known to degrade the mechanical performance of the BCC Fe-Cr and Fe-Cr-Al based alloys and has traditionally been called the “475°C embrittlement” [27, 29, 30].

Under irradiation, the formation of Cr-rich α' has been observed at and below the peak thermal aging temperature (475°C) and has been attributed to accelerated formation due to the production of mobile point defects after the damage cascade and quench [26, 31, 32]. For the Fe-Cr alloy systems, the formation of the Cr-rich α' phase and dislocation loops has been indicated as leading microstructural features causing significant radiation-induced hardening and embrittlement [33-39]. The severity of the radiation-induced hardening and embrittlement, under typical LWR conditions, is dependent on chromium composition. Above ~8 wt % Cr, the hardening and embrittlement response is due to the

formation of the Cr-rich α' phase with minor contributions from dislocation loop formation. Below ~8 wt % Cr, the hardening response is dominated by dislocation loop formation, as the alloys with lower chromium content lie outside the $\alpha - \alpha'$ miscibility gap (see Ref. [36] and those within).

Based on thermal aging studies of Fe-Cr-Al alloys, it is hypothesized the Fe-Cr-Al microstructural and mechanical responses after irradiation under typical LWR conditions are similar to those of Fe-Cr alloys. This work focuses on the validation of this hypothesis by investigating the microstructural evolution and mechanical performance of Fe-Cr-Al alloys after neutron irradiation in the High Flux Isotope Reactor (HFIR). The dislocation loop morphology and phase stability after irradiations near LWR operating temperatures at 1.8 dpa were studied using advanced characterization techniques including scanning transmission electron microscopy (STEM) and small-angle neutron scattering (SANS). In parallel to the microstructural characterization studies, the mechanical performance of the alloys was investigated using sub-size tensile specimens. Mechanical and microstructural information was used to develop a first-order structure–property relationship in the Fe-Cr-Al alloys based on the dispersed barrier-hardening model.

2. Materials and Methods

Simple ferritic model Fe-Cr-Al alloys with varying aluminum and chromium compositions were fabricated for this study. The chemical compositions were selected based on the kinetic boundary established for high-temperature oxidation resistance in the presence of steam for the Fe-Cr-Al alloy class [1, 8]. Yttrium was added to increase the oxidation resistance at elevated temperatures, as it is known to reduce the growth rate and increase the adherence of the protective alumina scale that forms on exposed surfaces of Fe-Cr-Al alloys [40, 41]. The chemical composition for each alloy can be found in Table 1. The model alloys were fabricated by arc-melting pure element feedstocks of Fe and Cr and pre-alloyed Al-Y specimens. After arc-melting, the model Fe-Cr-Al alloys were hot forged/rolled and heat treated based on the conditions developed by Yamamoto et al. [5] to form a fully ferritic microstructure with 20–30 μm grain size. After heat treatment, the bulk specimens were cold rolled with a 10% thickness reduction to simulate the potential cold-shaping process route for seamless fuel cladding tube production.

Cold-worked feedstock was machined using electric discharge machining (EDM) to form SS-J2 dog-bone tensile specimens. The dimensions and configuration of the SS-J2 specimens are given in Figure 1. SS-J2 specimens were loaded into an irradiation target and irradiated in the target rod rabbit holder (TRRH) facility of the HFIR and irradiated for the entire duration of HFIR cycle 450b. The estimated neutron fluence was $2.0 \times 10^{25} \text{ n/m}^2$ ($E > 0.1 \text{ MeV}$), corresponding to a nominal dose level of 1.8 dpa and a dose rate of $8.1 \times 10^{-7} \text{ dpa/s}$.

The irradiation target was designed for an irradiation temperature of 320°C based on the position within the HFIR. Passive SiC thermometry was irradiated with the Fe-Cr-Al tensile specimens. Analysis of the passive SiC thermometry was used to validate the target irradiation temperature. Dilatometric analysis of the passive SiC thermometry was conducted up to 600°C at a constant ramp rate of 1°C/min using a Netzsch 402 CD dilatometer. Based on the dilatometric analysis proposed by Campbell et al. [42], the nominal irradiation temperature of the irradiation target in the THT facility was estimated at $381.9 \pm 5.4^\circ\text{C}$, where the error is calculated based on one standard deviation of the mean of three separate, randomly selected SiC thermometry samples. The non-instrumented capsules did not allow for real-time monitoring of the irradiation temperature, but based on the power cycle history for cycle 450b of the HFIR, large deviations in irradiation temperature are improbable but unconfirmed. The higher irradiation temperature of the irradiation target compared to the design temperature is most likely due to the higher nominal neutron fluxes during the experiment than the values used for the design of the irradiation target.

Uniaxial tensile tests were conducted using shoulder loading at a crosshead speed of 0.0055 mm/s corresponding to a strain rate of $\sim 10^{-3} \text{ s}^{-1}$ on a screw-driven machine. The engineering strain was calculated from the recorded crosshead separation using the initial gauge length of 5.0 mm. The engineering stress was calculated by dividing the applied load by the initial cross-sectional area. All tensile tests were performed at room temperature. One tensile test was performed for each sample and condition because limited samples were available for mechanical testing.

Single ends of fractured tensile specimens were mechanically polished on the flat sides to a mirror finish, using standard metallography techniques, before focused ion beam (FIB) preparation. Electron-transparent foils were prepared using an FEI Quanta 3D 200i FIB. FIB preparation was selected in lieu of electropolishing to reduce the magnetic aberrations observed during specimen tilting [43]. Samples were carefully fabricated to reduce FIB-induced damage that could create artifacts during analysis. All specimens were thinned to electron transparency and cleaned initially with a 5 kV Ga⁺ ion beam and finished with a 2 kV Ga⁺ ion beam polish.

At least two electron-transparent foils were fabricated from each alloy and condition; any foil showing significant FIB-induced damage artifacts under transmission electron microscopy (TEM) or STEM was excluded from analysis. A Philips CM200 FEG-(S)TEM operating at 200 kV housed at Oak Ridge National Laboratory (ORNL) was used. The dislocation loop type (i.e., dislocation loops with Burgers vector of either $a/2 \langle 111 \rangle$ or $a \langle 100 \rangle$) was determined by the observed morphologies during annular bright-field (ABF) STEM imaging on the [100] zone axis. Details on the conditions and principles of dislocation imaging in STEM mode can be found in Parish et al. [44]. Foil thickness was determined using convergent beam electron diffraction (CBED) analysis [45]. The defect densities were determined from the plots of slope of areal density versus foil thickness to correct for near-surface FIB-induced artifacts and/or loss of glissile defects to the foil surface [46-48].

SANS was performed at ORNL on the CG-2 general-purpose beam line (GP-SANS) at the HFIR [48]. The data were measured at three detector settings—sample to detector distances of 1.079 m and 7.779 m using 0.472 nm neutrons and of 19.279 m using 1.2 nm neutrons—with the detector laterally offset by 0.4 m to maximize the accessible range of momentum transfer, Q , in each setting. The three settings together spanned the range $0.01 < Q < 10 \text{ nm}^{-1}$. The data were placed on the cm^{-1} absolute scale by normalizing the medium data to the attenuated direct beam after correction for thickness and transmission. Data in the other two settings were scaled to match the medium data to generate the combined curves. All measurements were performed at room temperature. Measurements were performed on a single end of each fractured tensile specimen with a nominal thickness of 0.5 mm. Multiple

scattering corrections were negligible. Analysis assumed nano-precipitates were Cr-rich α' with a nominal composition of Fe-81.3Cr-4.7Al (wt %) [25] and were spheres interacting with an exclusion volume.

3. Results

3.1 Unirradiated microstructures

The as-received microstructures demonstrated numerous dislocations organized in a pronounced cell structure in all alloys, as expected from the cold-worked condition of the unirradiated specimens. Example dislocation structures for each alloy are presented in Figure 2. The line dislocation density in the alloys varied from $6.3 \pm 1.0 \times 10^{13} \text{ m}^{-2}$ to $1.5 \pm 0.7 \times 10^{14} \text{ m}^{-2}$ depending on the alloy investigated. The Fe-10Cr-4.8Al alloy exhibited the lowest dislocation density of all alloys investigated. A limited number of grains were investigated per condition because of the small electron-transparent area ($\sim 10 \times 10 \times 0.1 \text{ }\mu\text{m}$) produced via FIB and because the grain sizes were on the order of 20–30 μm .

3.2 Irradiated microstructures

All irradiated samples exhibited a microstructure containing a high number density of dislocation loops within a sparse dislocation line network. The dislocation line density was generally reduced compared with those of the unirradiated microstructures, with the minimum observed density of $2.9 \pm 0.7 \times 10^{13} \text{ m}^{-2}$ in the Fe-10Cr-4.8Al irradiated alloy and the maximum observed in the Fe-18Cr-2.9Al irradiated alloy at $6.0 \pm 0.9 \times 10^{13} \text{ m}^{-2}$. The dislocation loops were characterized based on the morphology when the samples were imaged on the [100] zone axis using STEM. The individual dislocation loop types were determined by comparing their morphologies with those of the dislocation-loop maps provided by Yao et al. [50], taking into consideration the different diffracting conditions (g-vector excitation) between two-beam TEM and on-zone STEM [44]. Based on this analysis, all samples investigated had a mixed population of dislocation loops with Burgers vector of $a/2 \langle 111 \rangle$ or $a \langle 100 \rangle$. No determination of loop character (i.e., interstitial and/or vacancy type) was conducted, as it fell out of the scope of work.

The size, number density, and ratio of the dominant dislocation loop types ($a/2 \langle 111 \rangle$ to $a \langle 100 \rangle$) varied for each alloy investigated. Figure 3 presents typical STEM-based micrographs showing the

dislocation microstructures in the irradiated alloys. Table 2 summarizes the quantified number densities and sizes. The $a/2 \langle 111 \rangle$ and $a \langle 100 \rangle$ dislocation loops demonstrated a gamma distribution based on size, with a cut-off at the minimum size at ~ 10 nm. The number density of $a/2 \langle 111 \rangle$ dislocation loops was significantly higher in the Fe-12Cr-4.4Al specimen and lowest in the Fe-10Cr-4.8Al, and the average size was between 19.8 and 31.9 nm. No apparent trends in the $a/2 \langle 111 \rangle$ dislocation loop density were observed based on alloy composition. The areal density versus thickness plots for the $a/2 \langle 111 \rangle$ dislocation loops, which was used for density quantification, did not intercept at zero; this indicates that a population of $a/2 \langle 111 \rangle$ dislocation loops were lost to the foil surfaces before STEM analysis, which could affect the calculated size and number density measurements but is an unavoidable artifact of the measurement techniques.

The number density of $a \langle 100 \rangle$ dislocation loops was similar in the Fe-12Cr-4.4Al, Fe-15Cr-3.9Al, and Fe-18Cr-2.9Al alloys, with a value near $\sim 2 \times 10^{20} \text{ m}^{-3}$, but was significantly lower in the Fe-10Cr-4.8Al. The lower number density of $a \langle 100 \rangle$ dislocation loops in the Fe-10Cr-4.8Al specimen seemed to influence the loop size, as that sample showed the largest loop diameters of any alloy. Even though the Fe-10Cr-4.8Al alloy exhibited different behavior with regard to the $a \langle 100 \rangle$ dislocation loop population, it had similar ratios of both dislocation loop types to the Fe-12Cr-4.4Al and Fe-15Cr-3.9Al alloys; it had $a/2 \langle 111 \rangle$ to $a \langle 100 \rangle$ ratios between six and eight, whereas the Fe-18Cr-2.9Al ratio was nearly half of that. The large grain sizes (20–30 μm) promoted nearly uniform distributions of dislocation loops within the matrix. No influence of the grain boundaries on the dislocation loop behavior was observed. However, a limited number of grain boundaries were observed, and their character (e.g., high angle versus low angle) was not determined.

Figure 3 also shows a population of instances of black dot damage. Several samples exhibited a significantly higher fraction of black dots compared with the entire population of samples investigated; and, as noted, they were excluded from any analysis. They were excluded because black dot damage is known to occur not only from neutron irradiation but also from FIB sample preparation [51]. It is

assumed the samples with a significantly higher fraction of black dot damage represent a superposition of both effects. After these specimens were eliminated from analysis, the plots of the black dot areal density versus thickness resulted in a linear relationship that intercepted at zero within a 95% interval band, indicating limited artifacting from the FIB sample preparation. Furthermore, most black dots existed below the ~10 nm cutoff for identifiable dislocation loops, supporting the assumption that the STEM-based analysis technique using the system configuration in this study is incapable of uniquely identifying dislocation loops below 10 nm.

The black dots in Figure 3 are therefore considered to be neutron-induced dislocation loops or defect clusters in cases where their specific character cannot be directly determined because of the resolution limit of STEM. The sizes and number density measurements for neutron-induced black dots based on this assumption are also given in Table 2. The average size of the black dots changed insignificantly between alloys. The highest number density of black dots was observed in the Fe-12Cr-4.4Al irradiated alloy, which is similar to the observed trends in the $a/2 \langle 111 \rangle$ dislocation loop population. Note that the irradiated Fe-12Cr-4.4Al microstructure showed pronounced sub-grain boundaries and moderate dislocation cell networking where these microstructural features were absent from the other irradiated alloys.

No irradiated alloys except the Fe-12Cr-4.4Al alloy showed evidence of voids or bubbles at the investigated irradiation dose and temperature. One Fe-12Cr-4.4Al specimen showed limited evidence of void swelling and was considered atypical, as all other FIB foil specimens of the same alloy did not exhibit similar void distributions.

No evidence of phase instability was observed using conventional TEM- or STEM-based techniques. As discussed, the expected precipitation in the irradiated Fe-Cr-Al alloys is the Cr-rich α' phase within the host Fe-rich α phase. The Cr-rich α' phase has been notoriously difficult to image using conventional electron microscopy techniques because α' is coherent with the α matrix. Therefore, SANS was conducted to determine the size and number density of α' within the irradiated alloys, as it has been

successfully used previously to quantify α' precipitation in the irradiated Fe-Cr system [31, 52, 53] and has seen limited use in thermally aged Fe-Cr-Al alloys [54, 55].

Figure 4 shows an increase in the scattering intensity above $q \geq 0.5 \text{ nm}^{-1}$ indicative of a nanometer-size second phase; the intensity within this range increases with increasing chromium and diminishing aluminum content of the model Fe-Cr-Al alloys. The response in the scattering intensities in the Fe-15Cr-3.9Al and Fe-18Cr-2.9Al alloys is qualitatively similar to responses seen in the thermally aged Fe-15.7Cr-4Al alloy by Messoloras et al. [54] or in L4Ta irradiated to 2.9 dpa at 325°C by Mathon et al. [31]. Mathon et al. attributed the scattering intensity changes to precipitation of isolated Cr-rich α' surrounded by a spherical exclusion volume depleted in chromium. Here, the same approach was used for the measured scattered intensities in Figure 4 (open symbols) to which a spherical exclusion model (solid lines) was fit to extract quantitative data from the SANS data. The reasonable fit of the model to the measured values in the range of $0.5 < q < 2.0 \text{ nm}^{-1}$ is evidence that the scattering signal from the nanometer-size precipitates is indicative of Cr-rich α' in the irradiated Fe-Cr-Al alloys.

Table 3 summarizes the calculated volume fraction, size, and number density of the α' precipitates based on the best-fit model applied in Figure 4. The accuracy of the absolute scaling calibration procedure of SANS data is typically assumed to be nominally 10%. However, this uncertainty applies to the universal scaling factor used for all of the data and will thus change the scale of the data for the volume fraction without affecting the relative changes in value. There is a similar, but more difficult to estimate, uncertainty in the scaling associated with the nuclear scattering length density of the α' phase relative to the matrix. The uncertainty in the individual precipitate particle volume from modeling—the largest contributing factor to the relative uncertainty between quantitative data points—is determined from the fitting uncertainties to average less than 5% per sample. Here, it is assumed the uncertainty in the overall scale in the volume fraction data presented in Table 3 is approximately 20%, while the relative uncertainty between each composition relative to the others is $\leq 5\%$.

The minimal uncertainty ($\leq 5\%$) between the different composition points allows for reasonable assessments on the trends in Cr-rich α' formation using the SANS technique. Thus, Table 3 indicates the

volume fraction and number density of the α' precipitates had a nearly linear increase with a rise in the chromium content and minor changes in the aluminum content in the alloys. In contrast, the size of the precipitates was less affected by variations in alloy composition; the irradiated Fe-10Cr-4.8Al alloy was the only alloy exhibiting distinctly different average precipitate diameters compared with the other three irradiated alloys investigated.

3.3 Mechanical properties

Uniaxial tensile tests were performed on each alloy before and after irradiation. The corresponding room-temperature engineering stress-strain curves for the four model alloys before and after irradiation are shown in Figure 5. The stress-strain curves show radiation-induced hardening with an increase in the yield and ultimate tensile strength in the Fe-12Cr-4.4Al, Fe-15Cr-3.9Al, and Fe-18Cr-2.9Al specimens, although the uniform and total elongation increased or decreased depending on the specimen. The Fe-10Cr-4.9Al sample exhibited radiation-induced softening, due to significant recovery of cold work, compared with accumulated radiation-induced defects, as discussed in the following section.

Note that the values determined for Figure 5 and Table 4 are the result of a single tensile test; therefore, there was no opportunity to develop a statistical average. There was only a single test because of the limited number of specimens available for tensile testing. The results show a drastic change in the radiation-induced hardening with increasing chromium content in the irradiated Fe-Cr-Al alloys even if reasonable experimental error is assumed. As hypothesized, this behavior is inherently similar to that of Fe-Cr alloys, in which yield strengths increase monotonically with increasing chromium content [34, 35].

4. Discussion

The presence of both $a/2 \langle 111 \rangle$ and $a \langle 100 \rangle$ dislocation loops in the model Fe-Cr-Al alloy is consistent with previous reports on Fe-Cr and He-irradiated commercial Fe-Cr-Al alloys [14, 34, 35, 50, 56-58]. In this work, for all alloys investigated, the $a/2 \langle 111 \rangle$ dislocation loop population was higher but the average size was smaller than that of the $a \langle 100 \rangle$ dislocation loops. That finding indicates that although the $a/2 \langle 111 \rangle$ dislocation loops are postulated to be mobile in the vicinity of 400°C [39, 56],

they are not completely lost to defect sinks during irradiation at the conditions studied. The higher average size of $a\langle 100 \rangle$ dislocation loops supports the theory that $a\langle 100 \rangle$ dislocation loops are immobile but grow as a result of the absorption of point defects and defect clusters, including small $a/2 \langle 111 \rangle$ dislocation loops [56].

No clear trends were observed for chromium content in the range of 12–18 wt % Cr and aluminum content in the range of 2.9–4.8 wt. %, as the mean dislocation loop diameters did not significantly change within the means of the measurement error, and higher number densities of $a/2 \langle 111 \rangle$ dislocation loops were consistently reported irrespective of alloy content. These results are qualitatively similar to those for neutron-irradiated Fe-Cr alloys [35, 57, 58]. The lower dislocation density (both types) in the Fe-10Cr-4.8Al result is similar to the findings of Katoh et al. [58], who indicated the lowest dislocation density was at 9 wt % Cr in Fe-Cr alloys irradiated in the Fast Flux Test Facility at a temperature of 425°C. The significant decrease in the line dislocation density for all alloys indicates a recovery of cold-work competing/interacting with the radiation-induced defect structure.

The absence of voids in the defect structure for the Fe-Cr-Al system is expected, as Little and Stow [15] showed no void swelling in a Fe-14Cr-4Al alloy neutron-irradiated up to 23 dpa between 380 and 615°C. The limited or no void swelling observed in this study and in the study by Little and Stow indicates that, within the limited composition and irradiation space studied, aluminum does not detrimentally impact the historically good swelling resistance of BCC Fe-Cr-based alloys. More detailed studies of the void swelling resistance of Fe-Cr-Al alloys are needed to confirm this finding.

The presence of α' after irradiation in this study is consistent with previous work on irradiated Fe-Cr and thermally aged Fe-Cr-Al alloys. Bachhav et al. [32, 59, 60] investigated a series of model Fe-Cr alloys irradiated to 1.82 dpa at 290°C using atom probe tomography. That study showed precipitation of α' to be linearly dependent on chromium composition, with an upper limit in number density near $5.3 \times 10^{24} \text{ m}^{-3}$ in the Fe-18Cr specimen, a result remarkably similar to the result here for the Fe-Cr-Al systems studied (see Table 3). Furthermore, the size change as a function of chromium content was nearly

identical; only the 10Cr composition alloys varied from the population, and they were the largest in average size for both experiments and alloy systems. Average size and number density data for the irradiated model Fe-Cr-Al alloys are also consistent with SANS data for low-dose neutron-irradiated commercial ferritic/martensitic steels [31]. The number densities reported here for irradiated Fe-15Cr-3.9Al are within the same order of magnitude (10^{24}) as those of Fe-15.5Cr-5Al specimens aged at up to 1000 h between 300 and 475°C and analyzed using SANS [54]. However, the average size of the irradiated specimens is larger than the average size reported in the thermally aged study for similar times at temperature. Variances between thermally aged and irradiated specimens are expected because of radiation-enhanced diffusion leading to possible accelerated coarsening.

The precipitation of α' can be rationalized in terms of the Fe-Cr and Fe-Cr-Al ternary phase diagrams. Bonny et al. [26] proposed an estimated phase diagram in which precipitation of α' is expected above ~8.5 wt % Cr at the irradiation temperature investigated in Fe-Cr alloys. Kobayashi and Takasugi [17] indicated that at 475°C, $\alpha - \alpha'$ phase separation is expected in Fe-Cr-Al alloys and significant additions of aluminum can effectively shift the $\alpha - \alpha'$ phase boundary toward higher chromium contents. Similar variants of the Fe-Cr-Al ternary phase diagram have been proposed by Ejenstam et al. [61] and supported by the modeling work of Li et al. [18].

The irradiated Fe-Cr-Al alloys in this study reside within the proposed dual phase $\alpha - \alpha'$ composition space, given the experimental uncertainty in the techniques employed and difference in the temperature studied by Kobayashi and Takasugi. The presence of α' revealed via SANS on the neutron-irradiated model Fe-Cr-Al alloys supports the proposed phase instability in the alloy class. The precipitation observed in the neutron-irradiated Fe-Cr-Al model alloys is probably due to an irradiation-accelerated mechanism, similar to that proposed for the irradiated Fe-Cr system, but the singular temperature and damage dose investigated does not provide sufficient information to confirm such a claim. Given this, the precipitation of α' in all irradiated Fe-Cr-Al alloys suggests the solubility limit exists below 10 wt % Cr and ~3-5 wt % Al near 382°C.

The proposed $\alpha - \alpha'$ phase boundaries for the Fe-Cr-Al system [17, 18, 61] indicates additions of aluminum will change the anticipated volume fraction of α' at a given temperature and constant chromium content. The result is the possibility for dynamic changes in the α' volume fractions when the chromium and aluminum content are varied. Here, the presented results are similar to those of Bachhav et al. where linearity was observed in the phase stability based on chromium content in simple Fe-Cr alloys. The minimal effect observed based on varying aluminum content in the results presented in Table 3 could be the result of the small wt % variation in Al between alloys and in the fidelity of the SANS measurements, discussed previously, not fully revealing the influences of aluminum on the precipitation of α' in the Fe-Cr-Al alloys studied. Higher fidelity measurements and/or studies on different Fe-Cr-Al compositions may not adhere to the presented trends. Future work would be needed to fully map the Fe, Cr, and Al composition-temperature-irradiation space to fully determine the composition dependencies and mechanism(s) for the $\alpha - \alpha'$ phase separation in irradiated Fe-Cr-Al alloys.

The irradiated Fe-Cr-Al alloys showed a change not only in the microstructure after irradiation but also in the mechanical performance. The change in the mechanical performance can be rationalized using a structure–property relationship based on a radiation-induced hardening model. Here, the hardening model is proposed based on the interaction of mobile dislocation with a dense network of radiation-induced defects. The radiation-induced defects are those defects that were visible using the characterization techniques presented in this study and quantified in Tables 2 and 3: Cr-rich α' , $a/2 \langle 111 \rangle$ dislocation loops, $a \langle 100 \rangle$ dislocation loops, black dots, and line dislocations or tangles/networks.

The observed hardening is related to these defects using the simplified dispersed barrier hardening (DBH) model [62],

$$\Delta\sigma_y = M\alpha\mu b\sqrt{\rho d} \ , \quad (1)$$

where

$\Delta\sigma_y$ is the change in yield strength compared with the unirradiated value,

M is the Taylor factor (3.06) [63],

α is the defect barrier strength during interactions with the $a/2 \langle 111 \rangle$ dislocations,

μ is the shear modulus (82 GPa),

b is the magnitude of the Burgers vector of $a/2 \langle 111 \rangle$ dislocations interacting with the radiation-induced defects during deformation (2.49×10^{-10} m),

ρ is the defect number density, and

d is the defect diameter.

The parameter α accounts for the fact that some obstacles may be partially cut or sheared by the dislocation as it bows out during the glide process, with a resistance less than an impenetrable Orowan obstacle ($\alpha = 1$). This parameter α strongly depends on barrier size, defect geometry, and defect type as well as temperature and has a value in a range of 0 to 1. Other models can be used, such as the Friedel-Kroupa-Hirsch (FKH) model [64, 65] that is suitable for weaker obstacle barriers, but the DBH model has been more widely used in accounting for radiation-induced defects similar to those observed here.

Note that Eq. (1) is valid only for one specific defect cluster with a fixed diameter. However, as indicated by the results, the microstructure of the neutron-irradiated Fe-Cr-Al alloys involves a complex combination of various defects. To account for the obstacles of different types, sizes, and number densities within a single microstructure, a model must be developed which superimposes the combined effects. The two common cases for superposition are the linear superposition, $\Delta\sigma_y = \sum_i \Delta\sigma_{y,i}$, and root-sum-square superposition, $\Delta\sigma_y = \sqrt{\sum_i (\Delta\sigma_{y,i})^2}$. By standard convention, radiation-induced hardening is calculated using the root-sum-square superposition for obstacles with similar strengths and the linear superposition for obstacles with dissimilar strengths [66-68]. In the present work, it is unclear what superposition rule best represents the physical mechanisms for radiation-induced hardening increases; therefore, both the linear and the root-sum-square superposition rules are considered.

The inclusion of prior cold work, which shows partial recovery under irradiation, complicates the overall evaluation using the DBH model. The conventions for cold work in austenitic stainless steels [69-71] are used here to account for the preexisting dislocation networks in the Fe-Cr-Al specimens. It is

assumed that the only obstacles to dislocation motion in the unirradiated case are the network dislocations due to cold working. Therefore, the strength of the cold-worked material can be represented as

$$\sigma_y^{unirr} = \sigma_y^0 + \Delta\sigma_y^{CW} , \quad (2)$$

where σ_y^0 is the intrinsic yield strength of the material before cold working and $\Delta\sigma_y^{CW}$ is the yield strength increase due to cold working, with the form

$$\Delta\sigma_y^{CW} = M\alpha\mu b\sqrt{\rho_{unirr}} , \quad (3)$$

where ρ_{unirr} is the initial dislocation density measured in the unirradiated specimens in Table 2. After neutron irradiation, the yield strength can be described as

$$\sigma_y^{irr} = \sigma_y^0 + \Delta\sigma_y^{irr} , \quad (4)$$

where $\Delta\sigma_y^{irr}$ is the contribution to the yield strength due to radiation-induced defects and is determined using either superposition rule. Combining Eqs. (2–4) and using the different superposition rules, the hardening model for linear superposition for the irradiated Fe-Cr-Al alloys can be rewritten as

$$\begin{aligned} \Delta\sigma_y^{linear} = M\mu b \bigg(& \alpha_{\alpha'}\sqrt{\rho_{\alpha'}d_{\alpha'}} + \alpha_{a/2\langle 111 \rangle}\sqrt{\rho_{a/2\langle 111 \rangle}d_{a/2\langle 111 \rangle}} + \alpha_{a\langle 100 \rangle}\sqrt{\rho_{a\langle 100 \rangle}d_{a\langle 100 \rangle}} \\ & + \alpha_{bd}\sqrt{\rho_{bd}d_{bd}} + \alpha_d(\sqrt{\rho_{irr}} - \sqrt{\rho_{unirr}}) \bigg) \end{aligned} \quad (5)$$

and the hardening model using the root-sum-square superposition can be written as

$$\begin{aligned} \Delta\sigma_y^{sum-root-square} = & \\ M\mu b \bigg(& \sqrt{\alpha_{\alpha'}^2\rho_{\alpha'}d_{\alpha'} + \alpha_{a/2\langle 111 \rangle}^2\rho_{a/2\langle 111 \rangle}d_{a/2\langle 111 \rangle} + \alpha_{a\langle 100 \rangle}^2\rho_{a\langle 100 \rangle}d_{a\langle 100 \rangle} + \alpha_{bd}^2\rho_{bd}d_{bd} + \alpha_d^2\rho_{irr} -} \\ & \alpha_d\sqrt{\rho_{unirr}} \bigg) . \end{aligned} \quad (6)$$

Among the parameters in Eqs. (5) and (6), only the strength factor for each defect type is unknown based on experimental methods. The strength factor has been shown to be strongly dependent on the defect type [72], and so a generalized value for all defects observed cannot be used within the model. Therefore, the strength factors must be determined based on empirical fitting to the experimental results in Tables 2–4.

By linearizing Eqs. (5) and (6) for each alloy investigated, it can be seen that the system is under-determined; as a result, any calculation of the strength factors produces an inconclusive result. To

overcome this issue, it is assumed that the black dots observed in the STEM-based analysis are either $a/2 \langle 111 \rangle$ or $a \langle 100 \rangle$ dislocation loops. Therefore, the population of black dots can be partitioned into the $a/2 \langle 111 \rangle$ and $a \langle 100 \rangle$ dislocation loop populations based on the dislocation loop ratios for each alloy. Partitioning of the black dot data into the dislocation loop data results in a critical case where an explicit value for each strength factor can be determined by minimizing the variance between the calculated and measured yield strength values. Further characterization of other irradiation conditions for the same alloys will result in an overdetermined system whereby assumptions of the black dot population could be neglected; that investigation falls outside the scope of work for this study.

The resulting barrier strength factors for each defect are listed in Table 5. These barrier strength factors were calculated based on finding the best fit to Eqs. (5) and (6) and taking into account the uncertainty levels provided in Tables 2–4. The α values obtained from the two different superposition rules are not significantly different. By applying a similar method to study the hardening of neutron-irradiated Fe-Cr model alloys, Bergner et al. [72] proposed that α is 0.048 ± 0.012 for Cr-rich α' precipitates and 0.40 ± 0.16 for dislocation loops (without distinguishing the difference between $a/2 \langle 111 \rangle$ and/or $a \langle 100 \rangle$ dislocation loops) using the Bacon, Kocks, and Scattergood (BKS) model [73]. Similar values were proposed using either the linear or root-sum-square superposition of the DBH model. The corresponding barrier strengths obtained from the present study are consistent with their results.

The preliminary results based on two different superposition rules indicate that Cr-rich α' and $a/2 \langle 111 \rangle$ dislocation loops are two relatively weak barriers compared with $a \langle 100 \rangle$ dislocation loops and line dislocations. Hence the Cr-rich α' precipitates with α of 0.04 (linear) and 0.06 (sum-root-square), respectively, represent a relatively weak barrier and in small number densities do not significantly contribute to the hardening. It is of interest that $a \langle 100 \rangle$ dislocation loops on average are a stronger barrier to dislocation motion than $a/2 \langle 111 \rangle$ dislocation loops, according to both superposition rules. The difference in the barrier strengths of the two dislocation loops could be due to their sessile and glissile nature under deformation [74, 75]. The difference between the barrier strengths of the two dislocation

loop populations suggests the ratio between the two dislocation loop types can have a significant impact on the hardening of Fe-Cr-Al alloys during and after irradiation, although the correlation drawn here between their known glide behavior and calculated barrier strengths could be fortuitous to a degree.

Substituting the α values in Table 5 into Eqs. (5) and (6) leads to the prediction of the yield strength increase. The comparison of the model predictions and the experimentally measured yield strength change is shown in Figure 6, which shows a good agreement for both superposition rule. Therefore, no distinction regarding which is the ideal superposition rule to use for irradiated Fe-Cr-Al alloys can be provided. Based on the empirically determined barrier strength of each defect type using the linear superposition rule, as shown in Eq. (5), the individual contribution of each defect to the hardening in this case can be computed, as seen in Figure 7. Because of the nonlinear relationship of the contributions of each type of defect in the root-sum-square case, individual contributions are not shown here.

Several important features of the system can be derived from Figure 7. The first is that although α' precipitates are the weakest barrier in the irradiated model Fe-Cr-Al alloys, the large number density ($\geq 10^{24} \text{ m}^{-3}$) at high chromium contents ($>10 \text{ wt \% Cr}$) and moderate aluminum contents ($\sim 3\text{-}5 \text{ wt \% Al}$) leads to significant contributions to the hardening response of the alloys. Such an interplay between the hardening effect and the number density change, as a function of chromium content, drives the chromium composition hardening response seen in Figure 5 and Table 4. Based on this conclusion, it becomes clear the chromium content of irradiated Fe-Cr-Al alloys with $\sim 3\text{-}5 \text{ wt \% Al}$ can be optimized to limit radiation-induced hardening from α' precipitates. The second feature is that the dislocation loops are intermediate-strength barriers, but their influence on hardening is limited by their low density. Finally, the line dislocations are strong barriers to dislocation motion and, based on the hardening contributions for α' precipitates and dislocation loops, radiation-induced softening can be observed, as in the measurement of the Fe-10Cr-2.9Al alloy.

The resulting model and analysis assumes the average strength of each defect type based on the average size and number density of the defects observed. Recent work indicates the barrier strength can

be dependent on the defect size [76, 77]; future work will consider the calculated size distributions in the radiation-induced hardening model. The model presented here takes into account only the visible, or resolvable, defects using STEM or SANS. STEM-based techniques have a finite lower limit on resolvable defects; therefore, it can be assumed defects below this limit were not resolved in this study. These “non-resolvable” defects, such as small solute or vacancy clusters, can also contribute to the radiation-induced hardening response. The effects of these non-resolvable defects were not included in the current analysis, so some error is intrinsic in the values given for the barrier strength values. Finally, the uncertainty in the experimentally measured yield-strength changes was not accounted for, but some uncertainty is expected. Further testing would be needed to determine the uncertainty bounds in the experimentally determined changes in yield strength after irradiation. Given the errors associated with the experimental analysis and simplifications assumed in the model, the model presented here serves only as a first-order approximation of the system; some variability is expected, although the model does provide initial insight into the correlation between microstructural and mechanical changes in the irradiated model Fe-Cr-Al alloys.

5. Conclusions

Model Fe-Cr-Al alloys were fabricated, irradiated in the HFIR, and investigated to determine the radiation tolerance of irradiated Fe-Cr-Al alloys. It was found the microstructural and mechanical responses of model Fe-Cr-Al alloys to irradiation at LWR-type conditions are similar to the responses of model Fe-Cr alloys. The visible radiation-induced microstructure consisted of Cr-rich α' precipitates, black dots, line dislocations, and $a/2 \langle 111 \rangle$ and $a \langle 100 \rangle$ dislocation loops. Limited or a complete absence of voids/bubbles was noted. The observed microstructural features impacted the mechanical properties, and radiation-induced hardening was observed and was highest in the Fe-18Cr-2.9Al alloy. It was demonstrated that a DBH model could be developed to establish a first order structure-property relationship. Based on the analysis, the following statements can be made:

- Cr-rich α' precipitates are a weak barrier to dislocation motion, but their contribution to radiation hardening can become significant with sufficiently high number densities ($\geq 10^{24} \text{ m}^{-3}$).

- Dislocation loops contribute to the hardening response, and it is possible that the sessile/glissile nature of the defect loops influences their barrier strength. It is proposed that sessile $a\langle 100 \rangle$ dislocation loops are stronger barriers to dislocation motion than glissile $a/2 \langle 111 \rangle$ dislocation loops.
- The strong dependence on chromium composition of radiation hardening is directly related to the mechanism(s) controlling the $\alpha - \alpha'$ phase separation. The increase in number density of α' precipitates is linearly related to a change in chromium composition when there is a minimum change in aluminum content; and at high chromium contents (>10 wt % Cr) and moderate aluminum contents ($\sim 3\text{-}5$ wt % Al), the α' precipitates dominate the hardening response in neutron-irradiated Fe-Cr-Al alloys.

Based on these results it can be concluded that the composition of Fe-Cr-Al alloys needs to be optimized to control Cr-rich α' precipitation under the presence of neutron irradiation at typical LWR temperatures if minimal radiation-induced hardening is requisite.

Acknowledgements

The authors would like to thank the Irradiated Materials Examination and Testing facility and Low Activation Materials Development and Analysis laboratory staff for their continuing support of this research, S. J. Zinkle for guidance on many of the concepts and techniques presented within this work, and R. Howard for the continuing design and support of the irradiation test train. K. G. Field would like to thank C. M. Parish for fruitful discussions on the imaging of dislocation loops.

Research was sponsored by DOE Office of Nuclear Energy, Advanced Fuel Campaign of the Fuel Cycle R&D program. Research on the CG-2 General Purpose SANS and neutron irradiation of Fe-Cr-Al alloys at ORNL's HFIR user facility was sponsored by the Scientific User Facilities Division, Office of Basic Energy Sciences, DOE. Transmission electron microscopy was partially sponsored by ORNL's Center for Nanophase Materials Sciences, which is sponsored by the Scientific User Facilities Division, Office of Basic Energy Sciences, DOE.

Tables

Table 1: Summary of model Fe-Cr-Al alloy compositions in weight percent (wt %)

Alloy	Composition (wt %)									
	Fe	Cr	Al	Y	C	S	O	N	P	Si
Fe-10Cr-4.8Al	85.15	10.01	4.78	0.038	0.005	0.001	0.0013	0.0003	0.006	<0.01
Fe-12Cr-4.4Al	83.56	11.96	4.42	0.027	0.005	0.0013	0.0017	0.0009	0.0	0.01
Fe-15Cr-3.9Al	80.99	15.03	3.92	0.035	0.005	0.0004	0.0025	0.0007	<0.002	0.01
Fe-18Cr-2.9Al	79.52	17.51	2.93	0.017	0.005	0.0006	0.0015	0.0011	<0.002	<0.01

Table 2: Summary of average size/number density of STEM-identified defects in Fe-Cr-Al specimens irradiated in the HFIR reactor at 382°C to a dose of 1.8 dpa

Property	Alloy			
	Fe-10Cr-4.8Al	Fe-12Cr-4.4Al	Fe-15Cr-3.9Al	Fe-18Cr-2.9Al
Cr content (wt %)	10.01	11.96	15.03	17.51
ρ_{unirr} (m^{-2})	$6.3 \pm 1.0 \times 10^{13}$	$1.5 \pm 0.7 \times 10^{14}$	$1.5 \pm 0.6 \times 10^{14}$	$1.0 \pm 0.5 \times 10^{14}$
ρ_{irr} (m^{-2})	$2.7 \pm 0.9 \times 10^{13}$	$4.7 \pm 1.3 \times 10^{13}$	$2.9 \pm 1.2 \times 10^{13}$	$6.0 \pm 0.9 \times 10^{13}$
$d_{a<100>}$ (nm)	52.5 ± 23.8	29.7 ± 13.7	41.1 ± 24.8	32.0 ± 15.6
$\rho_{a<100>}$ (m^{-3})	$3.4 \pm 0.5 \times 10^{19}$	$2.3 \pm 0.2 \times 10^{20}$	$1.0 \pm 0.6 \times 10^{20}$	$2.0 \pm 0.3 \times 10^{20}$
$d_{a/2<111>}$ (nm)	31.9 ± 18.7	19.8 ± 11.7	30.4 ± 16.8	25.5 ± 15.4
$\rho_{a/2<111>}$ (m^{-3})	$2.6 \pm 0.6 \times 10^{20}$	$1.5 \pm 0.3 \times 10^{21}$	$7.9 \pm 1.2 \times 10^{20}$	$7.2 \pm 1.0 \times 10^{20}$
d_{bd} (nm)	9.1 ± 3.2	8.2 ± 4.3	9.8 ± 3.8	8.4 ± 3.4
ρ_{bd} (m^{-3})	$1.0 \pm 0.1 \times 10^{20}$	$1.6 \pm 0.1 \times 10^{21}$	$3.2 \pm 1.2 \times 10^{20}$	$8.8 \pm 1.1 \times 10^{20}$

ρ , dislocation density; d , effective diameter

Subscripts: unirr, dislocation line density in unirradiated condition; irr, dislocation line density in irradiated condition; $a<100>$, $a\langle 100 \rangle$ dislocation loops; $a/2<111>$, $a/2\langle 111 \rangle$ dislocation loops; bd, black dots

Table 3: Summary of Cr-rich α' size and number density determined using SANS in Fe-Cr-Al specimens irradiated in the HFIR reactor at 382°C to a dose of 1.8 dpa

Property	Alloy			
	Fe-10Cr-4.8Al	Fe-12Cr-4.4Al	Fe-15Cr-3.9Al	Fe-18Cr-2.9Al
Cr content (wt %)	10.01	11.96	15.03	17.51
$\square_{\alpha'}$	0.009	0.051	0.119	0.180
$\rho_{\alpha'}$ (m^{-3})	2.0×10^{23}	2.0×10^{24}	4.8×10^{24}	6.8×10^{24}
$d_{\alpha'}$ (nm)	4.4	3.7	3.6	3.7

$\square$, volume fraction; ρ , dislocation density; d , effective diameter

Table 4: Summary of tensile properties of Fe-Cr-Al specimens irradiated at HFIR at 382°C to a dose of 1.8 dpa

Property	Alloy			
	Fe-10Cr-4.8Al	Fe-12Cr-4.4Al	Fe-15Cr-3.9Al	Fe-18Cr-2.9Al
Cr content (wt. %)	10.01	11.96	15.03	17.51
YS^{unirr} (MPa)	540	576	566	519
YS^{irr} (MPa)	465	667	766	801
UTS^{unirr} (MPa)	570	602	612	539
UTS^{irr} (MPa)	568	719	793	819
TE^{unirr} (%)	12.6	12.0	10.5	11.3
TE^{irr} (%)	17.1	10.3	13.5	9.2
UE^{unirr} (%)	2.2	0.6	0.6	0.8
UE^{irr} (%)	7.0	2.8	3.5	1.5

YS, yield strength; UTS, ultimate tensile strength; TE, total elongation; UE, uniform elongation

Superscripts: unirr, unirradiated condition; irr, irradiated condition

Table 5: Barrier strength factors for various defects computed by using the linear and sum-root-square superposition rules based on input from Tables 2–4

Superposition rule	Barrier strength of visible defects (α)			
	Cr-rich α'	$\alpha\langle 100 \rangle$	$\alpha/2\langle 111 \rangle$	line dislocations
Linear	0.04	0.31	0.05	0.68
Sum-root-square	0.06	0.33	0.17	0.56

Figures

Figure 1: SS-J2 sub-size tensile specimen geometry for HFIR irradiation.

Figure 2: Dislocation structure in unirradiated (a) Fe-10Cr-4.8Al, (b) Fe-12Cr-4.4Al, (c) Fe-15Cr-3.9Al, and (d) Fe-18Cr-2.9Al. STEM-ABF tilted to the $[100]$ zone axis used to image dislocation structures. Red arrows indicate the $[002]$ directions and blue arrows indicate the $[011]$ directions. Measured foil thickness (t) using CBED analysis reported for reference. (For interpretation of the references to color, the reader is referred to the Web version of this article.)

Figure 3: Dislocation structure in irradiated (a) Fe-10Cr-4.8Al, (b) Fe-12Cr-4.4Al, (c) Fe-15Cr-3.9Al, and (d) Fe-18Cr-2.9Al. STEM-ABF tilted to the $[100]$ zone axis used to image dislocation structures. Red arrows indicate the $[002]$ directions and blue arrows indicate the $[011]$ directions; $a/2 \langle 111 \rangle$ dislocation loops appear as ellipses parallel to the $[011]$, and $a\langle 100 \rangle$ dislocation loops appear edge-on and parallel to the $[002]$ direction. Measured foil thickness (t) using CBED analysis reported for reference. (For interpretation of the references to color, the reader is referred to the Web version of this article.)

Figure 4: Scattered intensities and best-fit model of Fe-Cr-Al specimens obtained from SANS after irradiation in the HFIR at 382°C to a dose of 1.8 dpa.

Figure 5: Engineering stress-strain curves for Fe-Cr-Al specimens before and after irradiation in the HFIR at 382°C to a dose of 1.8 dpa.

Figure 6: Comparison of the yield strength increase obtained from modeling and experiment for the irradiated specimens.

Figure 7: Radiation-induced hardening contributions due to different measured defects based on the linear superposition of the DBH model.

References:

- [1] B.A. Pint, K.A. Terrani, M.P. Brady, T. Cheng, J.R. Keiser, *Journal of Nuclear Materials*, 440 (2013) 420–427.
- [2] K.A. Terrani, S.J. Zinkle, L.L. Snead, *Journal of Nuclear Materials*, 448 (2013) 420–435.
- [3] S.J. Zinkle, K.A. Terrani, J.C. Gehin, L.J. Ott, L.L. Snead, *Journal of Nuclear Materials*, 448 (2014) 374–379.
- [4] M.T. Farmer, L. Leibowitz, K.A. Terrani, K.R. Robb, *Journal of Nuclear Materials*, 448 (2014) 534–540.
- [5] Y. Yamamoto, B.A. Pint, K.A. Terrani, K.G. Field, L.L. Snead, Development and property evaluation of nuclear grade wrought FeCrAl fuel cladding for light water reactors, In-preparation for journal submission, (2015).
- [6] B.A. Pint, S. Dryepontdt, K.A. Unocic, D.T. Hoelzer, *JOM*, 66 (2014) 2458–2466.
- [7] B.A. Pint, K.A. Terrani, Y. Yamamoto, L.L. Snead, Accepted in - *Metallurgical and Materials Transactions E*, (2015).
- [8] B.A. Pint, K.A. Unocic, K.A. Terrani, *Materials at High Temperature*, 32 (2015) 28–35.
- [9] J. Ejenstam, M. Thuvander, P. Olsson, F. Rave, P. Szakalos, *Journal of Nuclear Materials*, (2014).
- [10] J.R. Regina, J.N. Dupont, A.R. Marder, *Welding Journal*, 86 (2007) 170–178.
- [11] J. Tusek, A. Kosmac, U. Nartnik, A. Dejan, *Z. Metallkd.*, 93 (2002) 310–314.
- [12] K.G. Field, M.N. Gussev, Y. Yamamoto, L.L. Snead, *Journal of Nuclear Materials*, 454 (2014) 352–358.
- [13] N.M. George, K. Terrani, J. Powers, A. Worrall, I. Maldonado, *Annals of Nuclear Energy*, 75 (2015) 703–712.
- [14] J. Chen, P. Jung, W. Hoffelner, H. Ullmaier, *Acta Materialia*, 56 (2008) 250–258.
- [15] E.A. Little, D.A. Stow, *Journal of Nuclear Materials*, 87 (1979) 25–39.
- [16] M. Kangilaski, The effects of neutron radiation on structural materials, in, *National Aeronautics and Space Administration*, Columbus, Ohio, 1967, pp. 94–103.
- [17] S. Kobayashi, T. Takasugi, *Scripta Materialia*, 63 (2010) 1104–1107.
- [18] W. Li, S. Lu, Q.-M. Hu, H. Mao, B. Johansson, L. Vitos, *Computational Materials Science*, 74 (2013) 101–106.
- [19] H.G. Read, H. Murakami, *Applied Surface Science*, 94 (1996) 334–342.
- [20] H.G. Read, H. Murakami, K. Hono, *Scripta Materialia*, 36 (1997) 355–361.

- [21] C. Capdevila, J. Chao, J.A. Jimenez, M.K. Miller, *Materials Science and Technology*, 29 (2013) 1179–1184.
- [22] C. Capdevila, M.K. Miller, J. Chao, *Acta Materialia*, 60 (2012) 4673–4684.
- [23] C. Capdevila, M.K. Miller, F.A. López, G. Pimentel, J. Chao, *Philosophical Magazine*, 93 (2013) 1640–1651.
- [24] C. Capdevila, M.K. Miller, G. Pimentel, J. Chao, *Scripta Materialia*, 66 (2012) 254–257.
- [25] C. Capdevila, M.K. Miller, K.F. Russell, J. Chao, J.L. González-Carrasco, *Materials Science and Engineering: A*, 490 (2008) 277–288.
- [26] G. Bonny, D. Terentyev, L. Malerba, *Scripta Materialia*, 59 (2008) 1193–1196.
- [27] P.J. Grobner, *Metallurgical Transactions*, 4 (1973) 251–260.
- [28] R.O. Williams, *Trans. Met. Soc. AIME*, 212 (1958) 497–502.
- [29] J.J. Heger, *Metal Progress*, (1951) 55–61.
- [30] J. Chao, C. Capdevila, M. Serrano, A. Garcia-Junceda, J.A. Jimenez, M.K. Miller, *Materials and Design*, 53 (2014) 1037–1046.
- [31] M.H. Mathon, Y. de Carlan, G. Geoffroy, X. Averty, A. Alamo, C.H. de Novion, *Journal of Nuclear Materials*, 312 (2003) 236–248.
- [32] M. Bachhav, G. Robert Odette, E.A. Marquis, *Scripta Materialia*, 74 (2014) 48–51.
- [33] L. Malerba, A. Caro, J. Wallenius, *Journal of Nuclear Materials*, 382 (2008) 112–125.
- [34] M. Matijasevic, A. Almazouzi, *Journal of Nuclear Materials*, 377 (2008) 147–154.
- [35] S.I. Porollo, A.M. Dvoriashin, A.N. Vorobyev, Y.V. Konobeev, *Journal of Nuclear Materials*, 256 (1998) 247–253.
- [36] D. Terentyev, G. Bonny, C. Domain, G. Monnet, L. Malerba, *Journal of Nuclear Materials*, 442 (2013) 470–485.
- [37] S. Zinkle, J. Busby, *Materials Today*, 12 (2009) 12–19.
- [38] R.L. Klueh, D.R. Harries, *High-chromium ferritic and martensitic steels for nuclear applications*, ASTM, 2001.
- [39] N. Baluc, R. Schäublin, P. Spätig, M. Victoria, *Hardening mechanisms in ferritic/martensitic steels*, in: M. Grossbeck, T. Allen, R.G. Lott, A.S. Kumar (Eds.) *Effects of Radiation on Materials: 21st International Symposium*, ASTM International, West Conshohocken, PA, 2004.
- [40] B.A. Pint, *Oxidation of Metals*, 45 (1995) 1–37.
- [41] B.A. Pint, *Journal of American Ceramic Society*, 86 (2003) 686–695.

- [42] A.A. Campbell, W.D. Porter, Y. Katoh, L.L. Snead, In-preparation for journal submission, (2015).
- [43] K.G. Field, L.M. Barnard, C.M. Parish, J.T. Busby, D. Morgan, T.R. Allen, *Journal of Nuclear Materials*, 435 (2013) 172–180.
- [44] C.M. Parish, K.G. Field, A.G. Certain, J.P. Wharry, *Journal of Materials Research*, 30 (2015) 1246–1274.
- [45] D.B. Williams, C.B. Carter, *Transmission Electron Microscopy: A Textbook for Materials Science*, 2 ed., Springer, New York, 2009.
- [46] S.J. Zinkle, Y. Matsukawa, *Journal of Nuclear Materials*, 329–333 (2004) 88–96.
- [47] N. Yoshida, M. Kiritani, F.E. Fujita, *Journal of the Physical Society of Japan*, 39 (1975) 170–179.
- [48] S. Kojima, S. Zinkle, *Journal of Nuclear Materials*, 179–181 (1991) 982–985.
- [49] G.D. Wignall, K.C. Littrell, W.T. Heller, Y.B. Melnichenko, K.M. Bailey, G.W. Lynn, D.A. Myles, V.S. Urban, M.V. Buchanan, D.L. Selby, P.D. Butler, *Journal of Applied Crystallography*, 45 (2012) 990–998.
- [50] B. Yao, D.J. Edwards, R.J. Kurtz, *Journal of Nuclear Materials*, 434 (2013) 402–410.
- [51] H.-H. Jin, *Journal of Electron Microscopy*, 59 (2010) 463–468.
- [52] C. Heintze, F. Bergner, A. Ulbricht, H. Eckerlebe, *Journal of Nuclear Materials*, 409 (2011) 106–111.
- [53] A. Ulbricht, C. Heintze, F. Bergner, H. Eckerlebe, *Journal of Nuclear Materials*, 407 (2010) 29–33.
- [54] S. Messoloras, B.C. Pike, R.J. Stewart, C.G. Windsor, *Metal Science*, 18 (1984) 311–321.
- [55] Z. Száraz, G. Török, V. Kršjak, P. Hähner, *Journal of Nuclear Materials*, 435 (2013) 56–62.
- [56] Z. Yao, M. Hernández-Mayoral, M.L. Jenkins, M.A. Kirk, *Philosophical Magazine*, 88 (2008) 2851–2880.
- [57] D. Gelles, *Journal of Nuclear Materials*, 108 & 109 (1982) 515–526.
- [58] Y. Katoh, A. Kohyama, D. Gelles, *Journal of Nuclear Materials*, 225 (1995) 154–162.
- [59] M. Bachhav, G. Robert Odette, E.A. Marquis, *Journal of Nuclear Materials*, 454 (2014) 381–386.
- [60] M. Bachhav, L. Yao, G. Robert Odette, E.A. Marquis, *Journal of Nuclear Materials*, 453 (2014) 334–339.
- [61] J. Ejenstam, M. Halvarsson, J. Weidow, B. Jönsson, P. Szakalos, *Journal of Nuclear Materials*, 443 (2013) 161–170.
- [62] A. Seeger, J. Diehl, S. Mader, H. Rebstock, *Philosophical Magazine*, 2 (1957) 323–350.
- [63] R. Stoller, S. Zinkle, *Journal of Nuclear Materials*, 283–287 (2000) 349–352.

- [64] J. Friedel, *The London, Edinburgh, and Dublin Philosophical Magazine and Journal of Science*, 46 (2009) 1169–1186.
- [65] F. Kroupa, P.B. Hirsch, *Discussions of the Faraday Society*, 38 (1964) 49.
- [66] G.S. Was, *Fundamentals of Radiation Materials Science*, Springer Berlin Heidelberg, New York, NY, 2007.
- [67] M. Hiratani, V.V. Bulatov, *Philosophical Magazine Letters*, 84 (2004) 461–470.
- [68] G.R. Odette, G.E. Lucas, G. Tedeski, B.D. Wirth, Development of superposition rules for hardening in alloys containing multiple defect populations, in: F.W. Wiffen (Ed.) *Fusion Materials*, U.S. DOE Office of Fusion Energy Sciences, Oak Ridge National Laboratory, 1998, pp. 221–226.
- [69] G.D. Johnson, F. Garner, H.R. Brager, R.L. Fish, A microstructural interpretation of the fluence and temperature dependence of the mechanical properties of irradiated AISI 316, in: D. Kramer, H.R. Brager, J.S. Perrin (Eds.) *Proceeding of the tenth international symposium: Effects of Radiation on Materials*, ASTM, Savannah, Ga.
- [70] C. Pokor, Y. Brechet, P. Dubuisson, J.P. Massoud, X. Averty, *Journal of Nuclear Materials*, 326 (2004) 30–37.
- [71] G.E. Lucas, *Journal of Nuclear Materials*, 206 (1993) 287–305.
- [72] F. Bergner, C. Pareige, M. Hernández-Mayoral, L. Malerba, C. Heintze, *Journal of Nuclear Materials*, 448 (2014) 96–102.
- [73] D.J. Bacon, U.F. Kocks, R.O. Scattergood, *Philosophical Magazine*, 28 (1973) 1241–1263.
- [74] S. Dudarev, R. Bullough, P. Derlet, *Physical Review Letters*, 100 (2008).
- [75] S.P. Fitzgerald, Z. Yao, *Philosophical Magazine Letters*, 89 (2009) 581–588.
- [76] X. Hu, D. Xu, T.S. Byun, B.D. Wirth, *Modelling and Simulation in Materials Science and Engineering*, 22 (2014) 065002.
- [77] D. Terentyev, Y.N. Osetsky, D.J. Bacon, *Scripta Materialia*, 62 (2010) 697–700.