The effect of injected interstitials on void formation in self-ion irradiated nickel
containing concentrated solid solution alloys

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Abbreviations

SP-CSAs, single-phase concentrated solid solution alloys

Keywords

Injected interstitials; Irradiation effects; Ni containing concentrated solid solution alloys; Void swelling

Abstract

Pure nickel and three nickel containing single-phase concentrated solid solution alloys (SP-CSAs) have been irradiated using 3 MeV Ni^{2+} ions at 500 °C to fluences of 1.5×10^{16} and $5.0 \times 10^{16} \text{cm}^{-2}$. The radiation-induced voids were characterized using cross sectional transmission electron microscopy that distributions of voids and dislocation loops were presented as a function of depth. A various degree of void suppression was observed on the tested samples and a defect clusters migration mechanism was proposed for NiCo. In order to sufficiently understand the defect dynamics in these SP-CSAs, the injected interstitial effect has been taken into account along with the 1-dimentional (1-D) and 3-dimentional (3-D) interstitial movement mechanisms.

1. Introduction

Nickel and various single-phase nickel alloys have been used in studying the alloying effects in irradiation induced defect evolution [1-4]. Heavy-ions and neutrons are commonly used as the irradiation source to investigate the effect of alloying elements on defect cluster formation, such as swelling and radiation-induced segregation [1-4]. In some cases, minor elements in nickel alloys existed in the form of ordered solutes [1,4-7] and their distributions were not completely random. Recently, a family of single-phase concentrated solid solution alloys (SP-CSAs) with two to five equiatomic principle elements (Ni, Co, Fe, Cr, Mn) has been developed. These compositionally complex alloys were fabricated in the form of single-phase face-centered cubic (fcc) crystal structure with completely random atomic distributions. The alloys and single crystal nickel were then irradiated to various conditions to study the irradiation induced defect dynamics due to composition changes [8-12].

Previous studies have shown the changes in irradiation-induced defect behavior and mechanical properties for different SP-CSAs through experimental and simulation results. A variation in defect migration range and migration energy has been demonstrated in a series of recent studies [10,13-18]. In a study conducted at room temperature, three sets of nickel, NiCo and NiFe SP-CSAs were irradiated with 3 MeV Au^+ ions to fluences of 2×10^{13} , 1×10^{14} and $5 \times 10^{15} \text{cm}^{-2}$. The research found that as the fluence increased, the defect migration range increased dramatically; this result was supported by cross-sectional transmission electron microscopy (TEM) images and Rutherford backscattering spectrometry (RBS-C) channeling spectra [13]. The extended defect migration range, which is deeper than predicted by SRIM simulation, was explained by enhanced defect migration. The highly mobile defects such as interstitials and small vacancy clusters were found to migrate further than the predicted range, with the assistance of mechanical stress from implanted ions. They also determined that the deeper migration range found in nickel than in NiCo and NiFe for all testing conditions was due

to a higher defect migration rate in nickel; this conclusion was supported by a simulation study conducted by Béland et al. [19] using atomistic modeling.

To better study the dynamics of irradiation induced defects, high temperature irradiation was conducted at 500 °C ($\sim 0.45 T_m$, melting point of Ni). Defect mobility and migration in nickel and nickel- containing SP-CSAs (NiCo, NiFe, NiCoFe, NiCoFeCr and NiCoFeCrMn) were studied by Lu et al. [20] with 1.5 and/or 3 MeV Ni^{2+} ions irradiated to 3×10^{15} and $5 \times 10^{16} cm^{-2}$, respectively. In that study, the authors found that void formation was suppressed in multicomponent single phase alloys and categorized the materials into two groups to explain the defect migration mechanism. The criteria were based on the relative depth distribution of voids and dislocation loops along with the average void size presented in the samples. The group containing nickel and NiCo showed large voids in the irradiated region closer to the sample surface while interstitial type dislocation loops were separated from voids and existed in a deeper region. The other group containing NiFe, NiCoFe, NiCoFeCr and NiCoFeCrMn showed much smaller voids distributed beyond the end of the predicted ion range, while dislocation loops existed closer to the surface. Increasing the irradiated ion energy or fluence did not alter the observed distribution of various types of defect clusters. The two different types of defect clusters were attributed to the two different modes for the interstitial cluster motion in alloys; from a long range one-dimensional (1-D) mode in nickel to a short range three-dimensional (3-D) mode in NiFe, revealed by molecular dynamic simulations. It was also determined that the void swelling resistance found in the SP-CSAs is correlated to the segregation between interstitial and vacancy type defect clusters. The self-organization of irradiation induced defects was affected by the different atomic sizes of principal alloying elements in these SP-CSAs.

The study of mechanical properties for nickel containing SP-CSAs was performed by Jin et al. [21]. The materials used in their study included single crystal nickel, NiCo, NiFe, NiCoFe, NiCoCr, and NiCoFeCr, as well as polycrystalline NiCoFeCrMn with millimeter grain size. The irradiation at 500 °C was conducted using 3 MeV Ni^{2+} ions to a peak damage dose about 53 dpa. The swelling induced by irradiation was measured by a step-height method using optical profilometer. The results showed that nickel had the highest swelling, followed by NiCo, NiCoCr, NiFe and NiCoFeCr then NiCoFe and NiCoFeCrMn. The correlation between irradiation induced hardening and step-height based swelling was previously studied [21], atomic level insights on dislocation loop movements are highly desirable to reveal the correlation between the two measurements.

Even though the defect cluster movement and the swelling behavior at elevated temperature in Ni-containing SP-CSAs have been reported, the depth dependence of swelling has not been studied in detail. For example, in the study performed by Jin et al. [21] only the step-height based overall swelling is calculated, which did not reveal the commonly considered depth-dependent or peak swelling information. In ion irradiated materials, the damage and injected ion concentration vary with depth, and their interaction can result in complicated swelling behavior.

Depth dependent information can be achieved through preparing TEM samples in a cross-sectional method. A full understanding of the swelling behavior in the Ni-containing SP-CSAs requires a detailed analysis of voids as a function of depth over the entire ion range and even beyond. The objective of this study is to explore the depth dependent swelling behavior of the irradiated material and to further explore its relationship to the 1-D and 3-D defect migration mechanism based on quantitative microstructure characterization.

2. Experiments

2.1. Ion Irradiation experiments

Single crystal nickel and equiatomic nickel-containing SP-CSAs NiCo, NiFe and NiCoFeCr were investigated in this study. The single crystal materials were prepared using pure elemental metal Ni, Co, Fe and Cr (> 99.9% pure) by arc melting. The single crystals were then grown from the polycrystalline ingots using a high temperature optical floating-zone directional solidification method. The actual composition of Ni (50.5 at%) and Fe (49.5 at%) determined by atom probe tomography (APT) were close to the nominal composition, and no impurities other than trace O, N, and Ga, which were induced from APT sample preparation, were detected in all the samples. The nominal composition is, therefore, used for all the materials unless otherwise stated. Details of materials preparation and single crystal growth can be referred to reference [22]. Two sets of materials were irradiated with 3 MeV Ni^{2+} ions at 500 °C to fluences of 1.5×10^{16} and $5.0 \times 10^{16} \text{ cm}^{-2}$, respectively. The ion flux was controlled at $2.8 \times 10^{12} \text{ cm}^2/\text{s}^{-1}$. The irradiation used a rastered beam with scanning frequencies of 517 and 64 Hz for the horizontal and vertical direction, respectively [23]. The irradiation induced damage profile and injected ion concentration were calculated using SRIM-2013 in Quick Kinchin Pease Mode with a displacement threshold energy of 40 eV. Since atoms were displaced in a completely random form in nickel-containing SP-CSAs, nickel was chosen to be the target material in SRIM simulation. The simulation was also performed for the rest of tested materials, but no significant difference was observed. The calculated peak damage dose is about 17 and 53 dpa for the two ion fluences, respectively. The results of the calculations are plotted in Figure 1 as a function of depth.

2.2. Microstructure analysis

Cross-sectional TEM samples were prepared by FIB lift-out technique on an FEI Helios Nanolab workstation at the Michigan Center for Material Characterization at the University of Michigan (MC²). In order to remove defects induced by FIB preparation and focus on studying defects from the primary ion irradiation, flash polishing was performed [13]. The void distribution was characterized before flash polishing using JEM 3011TEM with under-focused condition in bright field. TEM samples were tilted away from the zone axis to enhance the contrast for voids. The dislocation loops were imaged near $g=[200]$ two beam condition after

flash polishing with a small degree of tilting and defocus to enhance void contrast. Sample thickness was measured using the electron energy loss spectroscopy (EELS) method using a JEOL 2100F aberration-corrected TEM.

The calculation method for void swelling is described as follows. ImageJ was used as the image processing software for counting voids. The voids were profiled in 100 nm increments in depth, starting from the surface, except in the case of pure nickel irradiated to $5.0 \times 10^{16} \text{cm}^{-2}$. Since the average void diameter exceeded 100 nm, a bin size of 200 nm was applied in this case. A total depth of 2000 nm from the surface was counted for each sample, but a depth of 1800 nm was used in the total swelling calculation for consistency with the previously published data [20]. The percentage of swelling in each bin was calculated using the equation listed at below.

$$S(\%) = \frac{\frac{\pi}{6} \sum_{i=1}^N d_i^3}{A \times \delta - \frac{\pi}{6} \sum_{i=1}^N d_i^3} \times 100 \quad [16]$$

The parameter A represents the imaged area; δ represents the sample thickness; d_i represents the void diameter and N represents the number of voids counted in each size group. The overall swelling was calculated using the same equation as above but has the bin size set as 1800 nm.

In the case of Ni irradiated to 1.5×10^{16} and $5.0 \times 10^{16} \text{cm}^{-2}$, the void swelling profiles were recalculated as a function of mass depth to include the density correction due to the large amount of swelling induced by ion irradiation. Mass depth is calculated by multiplying the effective density ($\rho_{\text{effective}}$) by the thickness of each depth region, defined as:

$$\rho_{\text{effective}} = (1 - \frac{\Delta V}{V}) \times \rho \quad [24]$$

where $\Delta V/V$ is the swelling and ρ is the density of the material. With the large amount of swelling, the distribution of injected ion concentration will inevitably be affected as well. Hence, after voids were profiled in depth, the resulting swelling profile was used to recalculate the injected ion concentration. By adopting the swelling profile, the SRIM calculation with modified material densities as a result of void swelling was calculated, assuming that the effect of voids is homogenous throughout each layer. The updated densities were input into SRIM as discrete layers of 100 nm and 200 nm each for the case of nickel irradiated to 1.5×10^{16} and $5.0 \times 10^{16} \text{cm}^{-2}$, respectively. The resulting nickel ion distributions will then represent the case with modified material densities. By adjusting the bulk atomic density in each layer, the nickel ion concentration is calculated. The distribution depth is converted to mass depth by multiplying with the original nickel density.

3. Results

The images of cross-sectional TEM samples are presented in Figure 2 and 3 for ion fluences of 1.5×10^{16} and $5.0 \times 10^{16} \text{cm}^{-2}$, respectively. In general, the distribution of voids is inhomogeneous and depth dependent. The average diameter decreases as the number of equiatomic principal elements increases. Among the characterized materials, the swelling behavior falls into two groups: one group has larger voids and a relatively even distribution, i.e. nickel and NiCo; the other group has much smaller voids and these voids tend to be found in a deeper region from the surface or may have few scattered voids close to the surface, i.e. NiFe and NiCoFeCr.

The number of voids counted and the area counted is provided in Table 2. The depth dependent swelling behavior for the two ion fluences of 1.5×10^{16} and $5.0 \times 10^{16} \text{cm}^{-2}$ are shown in Figure 4-6. Figure 4 shows the void number density distribution as a function of depth in pure nickel, NiCo, NiFe and NiCoFeCr. The profile of irradiation induced damage and implanted nickel ion concentration was overlaid on the plot to emphasize the effects on the void distribution. An obvious separation of two bands of voids was observed in NiCo and NiCoFeCr for both ion fluences, and swelling is suppressed by the injected interstitials. The calculated damage peak coincides with the valley of the void number density for both fluences, while the peak for injected ion concentration shifted deeper from the valley.

Figure 5 shows the depth distribution of the average diameter of voids. For nickel, the void size gradually increased with depth and peaked at around 800 nm from the surface for both fluences before gradually decreasing. The size distribution for NiCo at lower fluence showed two peaks; one peak is closer to the surface at around 300 nm, while the second peak is much deeper, appearing at around 1000 nm. For NiCo at higher fluence, there is only one peak at around 600 nm with a relatively larger void size around 100 nm. Additionally, for NiFe at higher fluence, the void size increased with depth and showed a peak at around 1800 nm before dropping abruptly to zero.

Figure 6 presents the depth distribution of swelling. Data are presented in two groups due to a scale difference of two orders. For pure nickel, the correction method can yield different statistical results [24] since swelling higher than 10% was observed. The swelling vs. depth and nickel ion concentration is replotted in Figure 7 using the effective thickness method as described in Section 2.2. The swelling distribution of Ni in Figure 6 shows the original depth vs. swelling without density correction for better comparison with data that has swelling values much lower than 10%. NiCo at lower fluence barely shows any degree of swelling at the same depth of the swelling peak for pure nickel. In NiCoFeCr, there is no swelling in the depth range between 600-1000 nm for both fluences. At lower fluence, the two swelling peaks are separated just as observed in NiCo, but only wider. The swelling for NiFe is entirely located at the deeper region compared with the other samples.

The overall swelling data (calculated using the method described in Section 2.2) is provided in Figure 8 to show the relative swelling resistance in the materials. Among all the testing samples, the swelling resistivity is highest for NiCoFeCr follow by NiFe, NiCo and then nickel. It concludes that NiFe has similar amount of swelling as NiCoFeCr at lower fluence, but NiCoFeCr clearly showed better swelling resistance at higher fluence. The swelling resistance for testing SP-CSAs is consistent for overall and depth dependent swelling.

4. Discussion

Table 1 lists the overall swelling calculated from this study using cross-sectional TEM images and the swelling measured by Jin et al.[21] using optical profilometer. Under the same irradiation condition, the overall swelling measured by different methods shows a similar trend between Ni containing SP-CSAs. A variety of measurement errors such as TEM resolution, EELS thickness measurement and void diameter measurement can contribute to the uncertainty. More importantly, due to the different defect cluster motion behavior in the different materials, the swelling may appear at different depths in different SP-CSAs, thus making the selection on the overall material volume counted for swelling challenging. The overall swelling quoted in Table 1 is a comparison among different alloys and serves as a reference for detailed investigation of depth dependent swelling. Also note that the swelling calculated for nickel in this study has the highest amount of documented swelling for heavy ion irradiation using a rastered beam, whereas the studies conducted previously for pure nickel mainly used a defocused beam [25-26]. It should be pointed out that the swelling in nickel has the tendency to saturate with increasing dose, as has been observed in several neutron irradiation studies [1, 27-29]. However, the results of nickel in this study are limited to two fluences and therefore it remains uncertain as to when the saturation may occur. An accurate swelling rate also could not be determined due to the limited fluences studied.

The results reported by Lu et al.[20] is further confirmed in our study by presenting the irradiation induced voids as a function of depth. According to Lu et al., there are several factors that can impact the defect distribution response, such as the type of defect clusters and the different defect clusters migration energies in the SP-CSAs. It is stated that the heterogeneous segregation of dislocation loops and voids results from different migration energies of interstitial clusters for various materials. This leads to 1-D defect movement dominant in nickel and NiCo, and 3-D defect movement dominant in NiFe and other SP-CSAs with higher complexity. Table 3 shows the defect migration energy barriers [20] for direct comparison.

In this study, we would like to note that the influence of injected interstitials should also be taken into account when studying the void distributions in heavy ion irradiated materials. It has long been recognized that injected interstitials can suppress the swelling by recombining with vacancies. The source of extra interstitials varies but is mainly due to the injected ions

during heavy ion irradiation [30-34]. The effect of injected interstitials can be seen clearly in the depth distribution of void number density and swelling, in Figure 4 and Figure 6, respectively. The overall influence of the injected interstitial on the void distributions will be discussed for different SP-CSAs.

Effects of injected interstitials in different SP-CSAs

As can be seen in the cross-sectional TEM images (Figure 2 and 3) along with the depth dependent distribution of void number density and swelling, the void separation is the most apparent in NiCo and NiCoFeCr, but the dominating mechanism seems to differ in these two cases. The separation band where no voids are present is narrower in NiCo but much wider in NiCoFeCr. The defect migration energy E_m is considered in an attempt to explain the observed difference. In Table 3, the defect migration energy ($E_m^{i,v}$, i represents interstitials while v represents vacancies) calculated by MD simulations [13] shows that nickel has much lower E_m^i for interstitial clusters than NiFe (0.002 eV versus 0.974 eV for 4 interstitial clusters); while NiFe has lower E_m^v for vacancies and their clusters than nickel (0.515 eV versus 0.9 eV for 2 vacancy cluster). Mansur and Yoo [34] have suggested that the influence of injected interstitials toward void growth rate is more pronounced in systems with high vacancy migration energy. The calculated E_m^v for a single vacancy is 1.32 eV and 0.756 eV for pure nickel and NiFe [20], respectively. According to Mansur and Yoo, this E_m^v is relatively low and a significant injected interstitial effect is not expected. However, Plumton and Wolfer [35] have investigated the injected interstitial effect on void nucleation and have concluded its effect is most pronounced where defect recombination dominates. Garner [33] also has proposed that the injected interstitial effect would be enhanced as long as the chance of an extra interstitial finding a void prior to finding other sinks was increased.

Ni

In the case of Ni, it has the lowest E_m^i due to the least amount of lattice distortion and highest swelling rate in all tested material. At first sight, the suppression effect on void formation and growth from injected interstitials may not seem significant from the direct observation of TEM images. However, the swelling and modified Ni ion concentration versus mass depth in Figure 7 shows an obvious suppression on void swelling in the region of injected interstitials. The same trend was observed in both high and low doses for Ni. The influence of sputtering yield from 3 MeV Ni^{2+} ions is less than 10 nm for the tested fluences and is therefore negligible in this case. The irradiation response in Ni and NiCo will be compared later.

NiCo

In the previous result published by Lu [20], NiCo was categorized with the same group of defect migration mechanism as nickel for their similar void and dislocation loop distributions.

However, as the peak dose increased from 4 dpa in their case [20] to 17 and 60 dpa in this study, it was found that in NiCo, there are significant amount of dislocation loops closer to the surface and existing in the same region as voids. This observation contradicted the previous result of NiCo irradiated to 4 dpa from Lu [20], that only voids appeared closer to the surface. Based on the defect distribution from the TEM images, it is proposed that the defect migration mechanism for NiCo actually acted as a transition in between the 1-D and 3-D motion. This can be supported by observing the region closer to the surface, with the amount of dislocation loops increasing as the peak dose increased from 17 to 60 dpa. Furthermore, the observation of relatively ordered dislocation loops in the region of void denuded zone at lower dose provides interesting insight into how the injected interstitials can suppress voids formation. However, with the current data, it is hard to distinguish if the ordered dislocation loops were formed solely due to injected interstitials. The voids appearing after the denuded zone might be due to the fact that it takes longer for the injected interstitials to reach a concentration that is sufficient to have an effect on nucleation; while the highly mobile interstitials induced by irradiation already form a vacancy supersaturated zone in the cascade collision region.

Another interesting result is that the difference between the injected interstitial effect on Ni and NiCo. In Ni, the suppression on void swelling was dominant at the peak of Ni ion concentration and no void denuded zone was observed; while in NiCo, the injected interstitial resulted in a void denuded zone between two regions with a large amount of swelling. This is likely because of the difference in E_m^i , which is lower for Ni but may be relatively higher in NiCo due to the increased chemical complexity. In Ni, the injected interstitials have the same mobility as the interstitials generated from irradiation displacement and would quickly migrate away. They would either escape toward the surface or toward a much deeper region through 1-D motion. Furthermore, the concentration of injected interstitials accumulated throughout the irradiation would have a minimal effect toward swelling suppressing if a material has a higher void nucleation rate. Hence, the recombination effect would not be significant enough to form a void denuded zone. In NiCo, it is suspected that with increasing chemical complexity, there will be a transition in the migration mechanism of defect clusters. In Figure 9 and 10, it is observed that the defect distribution still possesses the features of 1-D motion, with the majority of voids closer to the surface and dislocation loops in the much deeper region. However, there is also an increased amount of dislocation loops distributed in the region closer to the surface, which is the main feature for 3-D motion. The trend is enhanced with increased dose in NiCo and contradicted with the model proposed by Lu [20] where they suggested Ni and NiCo have the same defect migration mechanism. With the new data from this study, it is proposed that NiCo has the defect clusters migration mechanism in a transition between 1-D and 3-D motion resulting from increasing locally irradiation-induced disorder.

NiFe and NiCoFeCr

NiFe and NiCoFeCr have relatively low interstitial migration mobility compared to nickel. Based on the 3-D motion [20], these sessile interstitials tend to form dislocation networks

in the cascade region. The amount of injected interstitials is relatively low compared with the amount of interstitials created from irradiation displacement, hence, the injected interstitial effect should be relatively minor compared with the case of nickel and NiCo. It is proposed in this study that the defect migration mechanism in NiFe and NiCoFeCr is dominated by the 3-D motion with a minor effect from injected interstitials. The isolated voids found in NiCoFeCr but not NiFe can be explained from the different degree of lattice distortion. As the lattice distortion is relatively higher in NiCoFeCr than NiFe, this makes it harder for the vacancies in NiCoFeCr to escape from the surface and they tend to form isolated voids in the region close to the irradiated surface instead.

While the injected interstitial effect behaves differently based on material composition in testing SP-CSAs, the influence of other irradiation parameters should not be ignored. This includes temperature, displacement rates, ion energies and raster frequencies, etc. The void separation band was not observed in another set of SP-CSAs irradiated to 4×10^{14} and $3 \times 10^{15} \text{ cm}^{-2}$ (with peak damage equal to about 0.53 and 4 dpa, respectively) using 1.5 MeV Ni ions at 500 °C. In this case, the suppression on void swelling was not observed in the region where injected interstitials come to rest. A potential explanation would be that the low concentration of injected interstitials was not enough to affect the voids nucleation process. Further study would be required to determine the effects of different irradiation parameters on injected interstitials.

5. Summary

The depth dependent void distributions in self ion irradiated nickel and three nickel containing SP-CSAs has been presented in this study. It has been found that the influence of injected interstitials should be considered along with the so called 1-D or 3-D interstitial movement models to interpret the defect migrations in different SP-CSAs. NiCo exhibits a defect migration behavior between 1-D and 3-D motion due to increased local strain or disorder. The influence of injected interstitials appears differently for each testing material. The suppression effect from injected interstitials on void swelling is most pronounced in Ni and NiCo, but the resulting defect distribution is different due to the difference on interstitial migration energy and chemical complexity of materials. A void denuded zone overlaid with the injected ion concentration and a band of relatively ordered dislocation loops induced by injected interstitials are observed in NiCo, but not in Ni. A broader void denuded zone found in NiCoFeCr is believed to be a combination of 3-D or sluggish interstitial movement and minor injected interstitials effect. The voids in NiFe are found only in the deeper region, but not near the surface. The reason that isolated voids at the surface are only found in NiCoFeCr but not in NiFe is likely due to a surface effect and less lattice distortion in NiFe compared with NiCoFeCr. The surface effect and less lattice distortion allow vacancies to escape easier to the surface. It has been demonstrated in this study that when discussing defect dynamics induced by ion irradiation, the injected interstitial effect should be taken into account, and the effect can be altered by

displacement rate, ion energy, vacancy migration barrier, interstitial mobility that lead to various swelling behaviors.

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Figure 1. SRIM prediction profile of depth distribution of displacement damage and implanted ions for 3.0 MeV Ni²⁺ ions irradiated to (a) 1.5×10^{16} ions/cm² and (b) 5.0×10^{16} ions/cm².

Figure 2. Under-focus cross-sectional BF TEM images of Ni and Ni-alloys irradiated by 3 MeV Ni²⁺ at 500 °C to 1.5×10^{16} ions/cm².

Figure 3. Under-focus cross-sectional BF TEM images of Ni and Ni-alloys irradiated by 3 MeV Ni^{2+} at 500 °C to $5 \times 10^{16} \text{ ions/cm}^2$.

Figure 4. Depth distribution for void number density is overlaid on the damage dose (orange dotted line) and implanted ion concentration (black smooth line) for 3 MeV Ni^{2+} irradiation in Ni and Ni-containing equiatomic alloys to (a) 1.5×10^{16} and (b) $5 \times 10^{16} \text{ ions/cm}^2$.

Figure 5. Depth distribution for average void diameter using 3 MeV Ni^{2+} irradiation in Ni and Ni-containing equiatomic alloys to (a) 1.5×10^{16} and (b) 5×10^{16} ions/cm^2 .

Figure 6. Depth distribution for swelling on $1.5 \times 10^{16} \text{ ions/cm}^2$ (a) Ni and NiCo (b) NiFe and NiFeCoCr; and $5 \times 10^{16} \text{ ions/cm}^2$ (c) Ni and NiCo (d) NiFe and NiFeCoCr.

Figure 7. Swelling profile and modified Ni ion concentration for Ni as a function of mass depth irradiated to (a) $1.5 \times 10^{16} \text{ ions/cm}^2$ and (b) $5.0 \times 10^{16} \text{ ions/cm}^2$.

Figure 8. Fluence dependence of overall swelling in Ni, NiCo, NiFe and NiCoFeCr irradiated using 3.0 MeV Ni²⁺ at 500°C.

Table 1. The comparison of SP-CSAs swelling measured by cross-sectional TEM images and optical profilometer. All materials were irradiated to $5.0 \times 10^{16} cm^{-2}$ using 3MeV Ni²⁺ ions at 500 °C.

Material	Swelling (%)*	Swelling (%)**
Ni	12.59±2.26	6.7
NiCo	3.60±0.67	3.5
NiFe	0.38±0.07	0.33
NiCoFeCr	0.016±0.005	<0.2

*The swelling calculated in this study using cross-sectional TEM images.

**The swelling calculated by Jin [21] using optical profilometer.

Figure 9. Distribution of voids and dislocation loops at under-focus cross-sectional BF TEM images of Ni and Ni-alloys irradiated by 3 MeV Ni^{2+} at 500 °C to $1.5 \times 10^{16} \text{ ions/cm}^2$.

Figure 10. Distribution of voids and dislocation loops at under-focus cross-sectional BF TEM images of NiCo and NiCoFeCr irradiated by 3 MeV Ni^{2+} at 500 °C to $5.0 \times 10^{16} \text{ ions/cm}^2$.

Table 2. Summary of voids in 3 MeV Ni^{2+} irradiated Ni containing SP-CSAs at 500 °C in the region of 0-1800 nm.

Material	Peak dose (dpa)	Number of voids counted	Area examined (μm^2)	Average void diameter (nm)	Void density (10^{20}m^{-3})	Swelling (%)
Ni	17	42	3.28	76.1 ± 1.0	1.28 ± 0.13	4.64 ± 0.84
NiCo	17	240	3.33	21.9 ± 1.0	7.21 ± 0.72	0.55 ± 0.11
NiFe	17	144	3.57	6.7 ± 1.0	4.03 ± 0.40	0.00842 ± 0.0046
NiCoFeCr	17	69	2.05	8.2 ± 1.0	3.37 ± 0.34	0.01383 ± 0.0047
Ni	53	61	6.82	131.6 ± 1.0	0.596 ± 0.06	12.59 ± 2.26
NiCo	53	355	3.5	32.7 ± 1.0	11.93 ± 1.2	3.60 ± 0.67
NiFe	53	64	3.58	44.9 ± 1.0	0.54 ± 0.05	0.38 ± 0.069
NiCoFeCr	53	97	2.88	9.0 ± 1.0	3.29 ± 0.3	0.0165 ± 0.0049

Table 3. Migration energy barriers, E_m of point defects (vacancy and interstitial) and their clusters in NiFe as compared with that in pure nickel [20].

Cluster	$E_m(\text{eV})$	
	NiFe	Nickel
I_1	0.516	0.16,0.1
I_2	0.617	0.12
I_3	0.728	0.04
I_4	0.974	0.002
V1	0.756	1.32
V2	0.515	0.9
V3	0.624	-
V4	0.747	-

Note: I represents interstitial and V represents vacancy.