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>).

Ion irradiation induced defect evolution in Ni and Ni-based FCC equiatomic binary alloys

K. Jin^{1,2}, H. Bei¹, and Y. Zhang^{1,2*}

¹ Materials Science & Technology Division, Oak Ridge National Laboratory, Oak Ridge, TN 37831, USA

² Department of Materials Science & Engineering, University of Tennessee, Knoxville, TN 37996, USA

Abstract

In order to explore the chemical effects on radiation response of alloys with multi-principal elements, defect evolution under Au ion irradiation was investigated in the elemental Ni, equiatomic NiCo and NiFe alloys. Single crystals were successfully grown in an optical floating zone furnace and their (100) surfaces were irradiated with 3 MeV Au ions at fluences ranging from 1×10^{13} to 5×10^{15} ions cm^{-2} at room temperature. The irradiation-induced defect evolution was analyzed by using ion channeling technique. Experiment shows that NiFe is more irradiation-resistant than NiCo and pure Ni at low fluences. With continuously increasing the ion fluences, damage level is eventually saturated for all materials but at different dose levels. The saturation level in pure Ni appears at relatively lower irradiation fluence than the alloys, suggesting that damage accumulation slows down in the alloys. Under high-fluence irradiations, pure Ni has wider damage ranges than the alloys, indicating that defects in pure Ni have high mobility.

Keywords: Metal and alloys; Ion irradiation; Solid solution; Defects; Energy dissipation

* To whom correspondence should be addressed: Email: zhangy1@ornl.gov

1. Introduction

Safe and reliable nuclear energy highly relies on the performance of structural materials under harsh conditions, including high-pressure, high-temperature, and continuous irradiation to high dose. As a result, seeking structural materials with superior mechanical properties, high thermal and phase stabilities, and high irradiation resistance has been a crucial but challenging task in nuclear engineering.¹⁻⁴ In order to achieve the desired properties, varieties of alloys consisting of one principal element and small amount of alloying elements have been developed and investigated. Such alloys are currently used as both in-core and out-of-core structural materials in most of the commercial reactors.¹ Despite the rapid development of the traditional alloys in the past decades, challenges from irradiation induced/accelerated degradations (embrittlement, solute segregation, corrosion, volume swelling, etc.) have not been fully conquered. These challenges become more significant in the radiation environments for most proposed types of Generation IV fission reactors and fusion reactors, in which the structural materials could be exposed to irradiation damage to over 100 displacements per atom (dpa).^{1,3,4}

Recently, a new type of concentrated solid solution alloys, including high entropy alloys, have been developed, which contain multiple principal elements with equiatomic or near-equiatomic molar ratios and exhibit supreme mechanical properties.⁵⁻¹⁰ In these alloys, traditional solute and solvent concepts do not exist, as all the elements are “majority”. In addition, the atomic radii of all the elements in these alloys are similar, e.g. 1.246, 1.252 and 1.274 Å for Ni, Co, and Fe, respectively.¹¹ These features may cause the traditional theories not suitable on explaining the alloying effects of mechanical properties or defect migration behavior (such as the dislocation migration) that have been largely applied in traditional alloys.⁶ At present, the irradiation response of such equiatomic solid solution alloys, e.g. defect production, accumulation and evolution, has not been well studied, and the underlying mechanism is unknown. In this work, equiatomic binary alloys with face-center-cubic (FCC) structure are investigated to demonstrate the impact of alloying elements on irradiation response.

Modifying the alloying elements is an important method in the design of alloys to tune their

properties and performances. Understanding whether and how they can affect the defect evolution in the alloys under irradiation is of critical importance in the “smart design” of alloys with desired irradiation response. We select elemental Ni, equiatomic NiCo and NiFe as our model systems because Ni, Co and Fe are important elements in the design of complex solid solution alloys with similar atomic masses, and single phase FCC alloys could be synthesized in Ni-Co and Ni-Fe systems with a large variation of Co or Fe concentration. In this paper, pure Ni is compared with equiatomic NiCo and NiFe alloys regarding their defect evolution process under 3 MeV Au ion irradiations. The irradiation induced damage is characterized using Rutherford backscattering spectrometry (RBS) technique, and significant chemical effects are observed.

2. Experimental methods

2.1. Single crystal growth and sample preparation.

Ingots of elemental Ni, equiatomic NiFe and NiCo were first mixed by arc melting the pure nickel, iron or cobalt starting materials (>99.9% pure) with the buttons flipped and remelted five times to ensure good mixing before drop casting into a copper mold. Single crystals were grown in an optical floating zone furnace from the polycrystalline drop-cast ingots.¹² During single crystal growth, the diameter of the molten zone was carefully reduced to produce a neck that prevented the slower growing grains from propagating.^{13,14} X-ray diffraction and metallographic examination were used to confirm the single crystal nature of the as-grown rods with FCC structures.¹⁰ The orientations of all crystals were measured by using backscattered Laue diffraction. Fig. 1 (a) and (b) show, as an example, a photograph of the as grown Ni single crystals with its corresponded Laue diffraction pattern. The as-grown crystals were re-oriented to $\langle 100 \rangle$ direction in a goniometer with assistance of backscattered Laue diffraction technology. Disks with thickness of ~1 mm cut normal to the $\langle 100 \rangle$ crystallographic directions using an electro-discharge machine (EDM) and followed by a careful electrochemical polishing to eliminate the damage layers of the machining. Before performing the irradiations, the orientation of the single crystalline samples were further examined by RBS channeling. Fig. 1 (c) shows the 2-D ion channeling image of the single crystalline Ni with $\langle 100 \rangle$ surface normal; the images of NiCo and NiFe are similar to pure

Ni. Wide axial $\langle 100 \rangle$ and planar channels (100) and (110) are clearly evident, indicating high-quality crystals.

2.2. Ion irradiation

The single crystals were irradiated with 3 MeV Au^{2+} ions at the Ion Beam Materials Laboratory (IBML) at the University of Tennessee in partnership with Oak Ridge National Laboratory.¹⁵ The irradiation fluences range from 1×10^{13} to 5×10^{15} ions cm^{-2} . The flux was controlled at 5.2×10^{11} ions $\text{cm}^{-2} \text{s}^{-1}$ for the fluences up to 1×10^{14} ions cm^{-2} ; for higher fluences, the flux was set at 2.8×10^{12} ions $\text{cm}^{-2} \text{s}^{-1}$. The samples were tilted $\sim 7^\circ$ off normal to minimize the channeling irradiation effects. In addition to the $\langle 100 \rangle$ axial channels, the planar channels, as shown in Fig. 1(c), were also carefully avoided to the largest extent. All the irradiations were performed at room temperature.

The displacement and implanted ion profiles of Au irradiation are simulated using The Stopping and Range of Ions in Matter (SRIM-2013)¹⁶, as shown in Fig. 2. Both “Full cascade” and “quick calculation” modes were used, and 40 eV is assumed as the threshold displacement energy.^{17,18} It is clearly shown that different simulation modes give different dpa values, as well as different shape of damage profile: their damage levels are similar at the surface region, while the peak dpa value given by the “full cascade” mode is about as twice large as that given by the “quick calculation” mode. As reported in Ref. ¹⁹, the “quick calculation” dpa agrees with the conventionally used NRT dpa, and the “full cascade” dpa is overestimated. For convenient comparison with literature results, the NRT dpa level is used to estimate the dpa level in this work. Because the atomic number of Ni, Co and Fe are similar, only the simulation results in pure Ni are shown here to represent the other cases.

2.3. RBS/channeling

The depth resolved ion irradiation induced damage was characterized with RBS/channeling technique using 3.5 MeV He^+ ions. The scattering angle was set at 155° . In the RBS measurements, the backscattered He ions from the collision with target atoms are detected by an energy-resolved silicon detector. The energy of detected He ions along with the known stopping power provides the depth where the backscattering occurs. If the target is a

damage-free single crystal and the incident ions are well aligned along the low-index crystalline axes or planes, e.g. <100> in this study, the backscattering yield will be significantly reduced from that of an amorphous target or when the incident beam direction is “randomly” chosen. This influence of crystal lattice on the trajectory of energetic ions, i.e. the significant reduction of backscattering yield, is called channeling effect.²⁰

The imperfections (defects, distortion, etc.) can lead to the increase of backscattering yield, attributed to two mechanisms. One is direct scattering, by which the channeled ions are directly backscattered by an atom not sitting at the pristine lattice points. The other one is termed as dechanneling, by which the channeled ions is deflected by the distortion of lattice, become a “randomly incident” ion. Different from the dechanneling mechanism, one of the features of direct backscattering is the appearance of a damage “peak” in channeling spectra, because the backscattering yield comes from the location where the defects exist, and thus it drops with decrease of the defect density.²⁰ It has been generally described that point defects or amorphous regions have strong effects on both mechanisms, while the extended defects such as dislocations have negligible effects on direct backscattering comparing with the effects on dechanneling.²⁰ For the contribution from dechanneling, the defect depth distribution can be derived through the formula²⁰⁻²²:

$$n_D(z) = \frac{1}{\sigma_D} \frac{d}{dz} \left\{ \text{Ln} \left[\frac{1 - \chi_v(z)}{1 - \chi_D(z)} \right] \right\}, \quad (1)$$

Where z is the depth from sample surface, n_D is the defect concentration, χ_v is the channeling yield of the undamaged (virgin) sample and χ_D is the normalized channeling yield of the sample containing defect. σ_D stands for a constant termed as the dechanneling cross section or dechanneling factor, which is related to the defect configuration and its relative orientation to the incident beam. The product of $n_D(z)\sigma_D$ is termed as (differential) dechanneling parameter, indicating the overall lattice distortion or disorder intensity.

In the case of metals subjected to the ion irradiation at room temperature or higher, the channeling spectra of both previous²³⁻²⁶ and present (Sec. 3) studies exhibit no significant damage “peak”, suggesting that the direct backscattering mechanism is not significant. In

other words, dispersed point defects and local amorphization are not observed to contribute much to the increase of the backscattering yield. This agrees with the TEM observation, which has been extensively applied to study the defect structure of FCC metals or alloys (dilute or concentrated), including the Ni-based FCC systems.²⁷⁻³⁰ In the low-fluence regime, black-dot cluster defects with a few nm-sizes are usually observed in TEM. And they were identified either dislocation loops (both interstitial- or vacancy-types) or stacking fault tetrahedra (SFT). In the high-fluence regime, extended dislocation loops or networks are formed. As a result, Eq. (1) should be applied to the data analysis in this study, as has been previously performed.^{21,23}

However, the accurate quantification of σ_D and then the absolute defect density is extremely difficult. For the simplest case, pure dislocation line, σ_D can be described by:

$$\sigma_d = K \left(\frac{abdE}{2Z_1Z_2e^2} \right)^{1/2} \quad (2)$$

where a is the Thomas-Fermi screening distance, b is the Burgers vector, d is the atomic spacing, e is the charge of an electron, E is the He beam energy, Z_1 and Z_2 are atomic numbers of the analyzing ion and target atom, and K is a constant related to the dislocation type and the angle between the channel and the dislocation. And it obviously relies on the relative orientation between the dislocation and the incident beam, which is hardly a single value. The analysis will be even more complicated if SFT is included into consideration. As a result, most attempts of quantifying the dechanneling factor rely on crude assumptions, and the conclusion may not be exclusive. More seriously, they can hardly describe the complicated cases when different defect structures (e.g. dislocation loops and SFT) are mixed. Fortunately, the dechanneling parameter calculated from Eq. (1) could represent the overall lattice distortion, and as a result, the depth profile under similar experimental conditions provides a trend of damage accumulation and evolution. Thus, in this work, we present the results of dechanneling parameter, rather than giving the absolute defect density.

3. Results

3.1. RBS channeling spectra

The crystalline quality of the pre-irradiated samples is measured and compared using RBS

channeling technique. Fig. 3 (a) shows the channeling spectra of the pristine samples and Fig. 3 (b) shows the angular scan of the $\langle 100 \rangle$ channel. In Fig. 3 (a), surface peaks of both Ni and Fe can be clearly identified in the NiFe spectrum, while the peak of Co is close to Ni. The overlap of three channeling spectra in Fig. 3 (a) and the three angular scan curves in Fig. 3 (b) demonstrate that the crystals used in this study have comparable initial crystalline quality. Meanwhile, the ratio of channeling to amorphous level, $\chi_{\min}$, after the surface peak, see Fig. 3 (a), is below 4%, and this is also revealed by the large dip depth shown in Fig. 3 (b). Such low $\chi_{\min}$ indicates high crystalline quality of all the samples. As the damage-free levels are similar among these materials, only the damage free spectrum of pure Ni will be shown in the following figures in this paper for clarity.

Typical RBS channeling spectra of 3 MeV Au^{2+} ion irradiated Ni, NiCo and NiFe are shown in Fig. 4. Five spectra are shown in each figure: besides the “fully amorphous level” at top and the “damage-free level” at bottom, the three other spectra correspond to Ni, NiCo and NiFe under the Au ion irradiation to each fluence.

Comparing the damage levels between the three materials directly from these channeling spectra, three correlations are observed at different irradiation fluence stages. First, at low fluences (i.e. up to 2×10^{13} ions cm^{-2} as shown in Fig. 4 (a)), the damage level in both alloys are lower than pure Ni. Among the two alloys, the damage level in NiFe is lower than in NiCo. This result shows that NiFe has higher irradiation resistance at low fluences while pure Ni is the most radiation-sensitive among these three materials. With further irradiation, shown in Figs. 4 (b) and (c), the damage levels become comparable. For example, at the fluence of 5×10^{13} ions cm^{-2} , the damage level of Ni and NiCo appears similar, while the damage level of NiFe is still lower. At the fluence of 1×10^{14} ions cm^{-2} , the damage levels of all these three materials become similar. Under the irradiation to higher fluences, i.e. up to 5×10^{15} ions cm^{-2} as shown in Fig. 4 (d), two phenomenon are observed. On one hand, the damage level in the alloys is higher than pure Ni before 600 nm, which is opposite to the cases at low fluences. Meanwhile, the damage range (identified here by the “knee”²⁴⁻²⁶ in the channeling spectra) of pure Ni is deeper than that for NiCo and NiFe.

3.2. Dechanneling parameters

In order to intuitively demonstrate the damage evolution process, the calculated dechanneling parameter, which is more closely related to the defect density, is shown in Fig. 5 (a)-(c) at different Au ion fluences of pure Ni, NiCo and NiFe, respectively. The calculation is based on Eq. (1), combined with the best Spline fitting of the channeling spectra. Fig. 5 (d) shows the comparison between the three materials regarding their dechanneling parameters at the displacement peak position (between ~150 to 200 nm) as a function of irradiation dose. As stated in Sec. 2.3, the dechanneling profile can qualitatively stand for the profile of lattice distortion or the defects concentration. Thus, the results in Fig. 5 provide better comparisons of overall damage buildup in Ni and the two binaries.

Three stages in damage evolution can be clearly identified from Fig. 5. In the low-fluence irradiation, damage accumulation is observed in all the three materials. In this stage, the dechanneling parameter profiles (considered as the damage profile) in all these three materials agree well with the SRIM prediction of the displacement events profile shown in Fig. 2. This result indicates that no significant defect propagation towards either surface or deeper region occurs in this irradiation fluence range.

When the ion fluence reaches the range of 5×10^{13} to 1×10^{14} ions cm^{-2} , saturation in the irradiation-induced damage level is observed in all these three materials. It can also be identified that pure Ni reaches a relative low saturation level at a relatively lower fluence (corresponding to the peak dose of ~0.3 NRT dpa estimated from the quick SRIM calculation as suggested by Ref.¹⁹). The saturation of defect cluster density in pure FCC metals has been observed under neutron irradiation and reviewed in Ref.²⁹. For example, the saturation dose for Cu is observed as ~0.1 NRT dpa, and the saturation dose for Ni is suggested larger than 0.1 NRT dpa but has not been captured.²⁹ Our results qualitatively agree to the previous results, regardless the different damage mechanisms between neutron irradiation and heavy ion irradiation. Under high-fluence irradiations, besides the increased damage range, the damage level at the depth of initial damage peak at ~200 nm (corresponding to the displacement peak) is decreased in all these three materials. The trend of the damage evolution of the three materials appears similar, but it is clearly delayed in the alloys, more significant in NiFe than in NiCo.

4. Discussion

The results presented in this study demonstrate that the alloying elements significantly affect the irradiation response. Comparing with NiCo, NiFe more effectively delays the irradiation induced damage accumulation and evolution. The reduction of defect yield and the delay in damage evolution comparing with pure metals have been observed in literatures for steels and some other conventional alloys.^{27,28,31} An explanation for steels is that the alloying elements/impurities reduce the probability or the length of the focused replacement collision sequences.³¹ As a result, the vacancy-interstitial separation is reduced in the alloys, and thus the defect recombination is enhanced. However, such explanations for traditional alloys does not apply for the alloys used in this study, because the key character in this mechanism is the interstitial solute atoms (e.g. carbon), which does not exist in our alloys that contains multi-principal elements. As a result, there must be different mechanisms underlying the observed alloying effects.

Robinson et al²⁸ has observed that the Ni-Cr binary alloys (8%-17%Cr) has lower defect yield under 80 keV Ni ion irradiation comparing with pure Ni, although no significant difference in defect sizes was observed. The authors speculate that the reduction in defect yield is due to the low probability of loop nucleation in the cascade core, but the underlying physics was not understood. One of the assumptions made in that paper is that the average number and the spatial distribution of Frenkel pairs created in each cascade are almost the same under the same irradiation condition due to the similar atomic masses for the studied materials, but this assumption might be questionable.

It is important to emphasize that, the observed defect yield is an integrated result of the defect production from the initiation of primary knock-on atom (PKA), thermal and athermal recombination, and the migration of the survival defects. Thus, the physics behind the observed damage reduction could be related to multiple stages including defect formation, accumulation and migration. Generally speaking, at least two mechanisms may be responsible to the observed alloying effects. First, with higher chemical disorder, the electron mean free path and electron thermal conductivity of the alloys are decreased. For example, the residual electronic resistance of NiFe is significantly higher than pure Ni,³² while NiCo

sits in between these two materials but closer to pure Ni.³³ Such difference in the transport properties may intrinsically modify the energy dissipation process through both electron and phonon subsystems and directly affect the defect evolution process. Second, different from pure metals, in the alloys with multi-principal elements, different atoms have different local atomic environment and thus subject to different force field. This may lead to the change of defect behavior such as the formation energy and the migration energy from that in pure metals.

Under the high-fluence irradiations, the material response exhibits two major differences from the low-fluence irradiations. First, all three materials show damages deeper than predicted ranges. This effect is called long-range effect, which has drawn intensive interest for decades in both TEM and channeling studies.³⁴⁻³⁷ Friedland et al. have shown damage ranges over 4 times deeper than expected in pure FCC metals (e.g. Cu) from their channeling studies,³⁴ and Sharkeev et al., in their TEM studies on varieties of metals and alloys, have observed that the peak dislocation density could appear at the depth several times deeper than the ion implantation region^{36,37}. Although the mechanism of the long-range effect has not been fully understood, it has been generally attributed to the high mechanical stress (static and dynamic) generated due to accumulation of irradiation damage, which could exceed the yield strength of materials.³⁷ Such stress could be high enough to initiate dislocation formation in the region deeper than the ion range. Besides, the high stress in the implanted region could also lead to the propagation of the dislocations generated in the ion implanted region.³⁴ This mechanism has been supported by the structural dependence: the long-range effect has been observed more significant in FCC metals than BCC and HCP metals, in agreement with the fact that the Peierls force is relatively smaller in FCC structures.³⁴ In addition, these explanations are further supported by our new observations of deeper damage range in Ni than that in NiCo and NiFe. The difference in the damage ranges with Ni being the widest and NiFe the shallowest is consistent with the recent findings that Ni is strongly strengthened by alloying with Fe, while the strengthening effect of NiCo is weak.⁶

Another effect for the high-fluence irradiation is that the backscattering yield or the dechanneling parameters of all the three materials are decreased at the SRIM-predicted

displacement peak, ~200 nm. Such observations by ion channeling technique have been frequently reported in different irradiation tolerant materials, including ceramics such as ZrO_2 ,³⁸ SrTiO_3 ³⁹ and MgO ⁴⁰. The interpretation of this phenomenon could be associated with several contributions. With increasing fluences, the defect structure will change: e.g. the population of small dislocation loops may decrease while long dislocation lines or networks are formed. Such more organized defects could induce less dechanneling, causing the decrease in backscattering yield. Moreover, as discussed above, the high-fluence irradiation induced stress can cause the moving/propagation of dislocations, which makes the highest damage level, determined using the RBS channeling technique, appear at a depth away from the most intense displacement cascade region (~200 nm). Considering the complexity of the defect structure in the high-fluence irradiations, and the difference in defect migration between these three materials, the quantification and parallel comparison of dechanneling parameter should be made mainly for the low fluence cases where damage accumulation is the main contributing factor and long-range defect migration is insignificant. For the high-fluence irradiations, the disorder level of Ni at the displacement peak, determined by RBS, is lower than the alloys, especially NiFe. It is worth noting that ion channeling technique provides depth resolved quantification of the lattice distortion, but it does not directly illustrate the nature of defects when complex defects are mixed together. As a result, further investigations, including detailed TEM analysis on defect structures, are desired to understand microstructure evolution.

5. Conclusions and Summary

Defect evolution in Ni, NiCo and NiFe is studied with Au ion irradiation over a wide range of irradiation fluence at room temperature using ion channeling technique. Significant delay in damage accumulation and evolution has been observed in the alloys especially in NiFe. In the low-fluence region, the irradiation resistance of equiatomic NiFe alloy is higher than NiCo and pure Ni. With further irradiation, “saturation” in damage level is observed in all these materials and the saturation in pure Ni is observed at lower irradiation fluence comparing with the alloys. Under high-fluence irradiations, the damage in all these materials extends to deeper depth and the damage range in pure Ni is larger comparing with the alloys. This result

indicates that the defects in pure Ni have higher mobility than in the alloys. The underlying mechanisms are expected to be different from that for traditional alloys with dilute solutions. The chemical disorder induced modification in the electron and phonon transport properties may intrinsically change the energy deposition through both electron and phonon subsystems. The unique local atomic environments could affect the defect behavior. Both issues may be responsible to the observed chemical effects. The results in this study suggest that the defect evolution under heavy ion irradiation could be tuned at different fluence stages with modifying the alloying elements, and it may provide guidance on the smart design of alloys with desired irradiation response.

Acknowledgements

This work was supported as part of the Energy Dissipation to Defect Evolution (EDDE), an Energy Frontier Research Center funded by the U.S. Department of Energy, Office of Science, Basic Energy Sciences. Ion beam work was performed at the University of Tennessee–Oak Ridge National Laboratory Ion Beam Materials Laboratory (IBML) located at the campus of the University of Tennessee, Knoxville.

References

- 1 S. J. Zinkle and G. S. Was, *Acta Mater.* **61** (2013) 735.
- 2 T. Allen, J. Busby, M. Meyer, and D. Petti, *Mater. Today* **13** (2010) 14.
- 3 S. J. Zinkle and J. T. Busby, *Mater. Today* **12** (2009) 12.
- 4 P. Yvon and F. Carré, *J. Nucl. Mater.* **385** (2009) 217.
- 5 B. Gludovatz, A. Hohenwarter, D. Catoor, E. H. Chang, E. P. George, and R. O. Ritchie, *Science* **345** (2014) 1153.
- 6 Z. Wu, H. Bei, G. M. Pharr, and E. P. George, *Acta Mater.* **81** (2014) 428.
- 7 Y. Zhang, T. Zuo, Z. Tang, M. C. Gao, K. A. Dahmen, P. K. Liaw, and Z. Lu, *Prog. Mater. Sci.* **61** (2014) 1.
- 8 X. Ye, M. Ma, W. Liu, L. Li, M. Zhong, Y. Liu, and Q. Wu, *Adv. Mater. Sci. Eng.* **2011** (2011) 1.
- 9 O. N. Senkov, G. B. Wilks, D. B. Miracle, C. P. Chuang, and P. K. Liaw, *Intermetallics* **18** (2010) 1758.
- 10 Z. Wu, H. Bei, F. Otto, G. M. Pharr, and E. P. George, *Intermetallics* **46** (2014) 131.
- 11 R. J. D. Tilley, *Understanding solids: the science of materials*. (John Wiley & Sons, England, 2004).
- 12 H. Bei and E. P. George, *Acta Mater.* **53** (2005) 69.
- 13 H. Bei, E. P. George, and G. M. Pharr, *Scripta Mater.* **51** (2004) 875.
- 14 Y. Z. Xia, H. Bei, Y. F. Gao, D. Catoor, and E. P. George, *Mater. Sci. Eng. A* **611** (2014) 177.
- 15 Y. Zhang, M. L. Crespillo, H. Xue, K. Jin, C. H. Chen, C. L. Fontana, J. T. Graham, and W. J. Weber, *Nucl. Instrum. Methods Phys. Res. B* **338** (2014) 19.
- 16 J. F. Ziegler, (www.srim.org).
- 17 A. Müller, V. Naundorf, and M. P. Macht, *J. Appl. Phys.* **64** (1988) 3445.
- 18 H. Wiedersich, *J. Nucl. Mater.* **179-181** (1991) 70.
- 19 R. E. Stoller, M. B. Toloczko, G. S. Was, A. G. Certain, S. Dwaraknath, and F. A. Garner, *Nucl. Instrum. Methods Phys. Res. B* **310** (2013) 75.
- 20 L. C. Feldman, J. W. Mayer, and S. T. Picraux, *Materials Analysis by Ion Channeling*. (Academic Press, 1982).
- 21 S. B. Dabagov, G. D. Tolstolutskaia, I. E. Kopanetz, and I. M. Neklyudov, *SPIE Proc.* **6634** (2007) 66340W.
- 22 G. Götz and B. Gruska, *Nucl. Instrum. Meth. Phys. Res.* **194** (1982) 199.
- 23 M. Vos and D. O. Boerma, *Nucl. Instrum. Meth. B* **15** (1986) 337.
- 24 M. K. Agrawal and D. K. Sood, *Nucl. Instrum. Method.* **149** (1978) 425.
- 25 J. A. Borders and J. M. Poate, *Phys. Rev. B* **13** (1976) 969.
- 26 Y. Fujino, Y. Igarashi, and S. Nagata, *Phys. Rev. B* **63** (2001).
- 27 S. X. Jin, L. P. Guo, Z. Yang, D. J. Fu, C. S. Liu, W. Xiao, R. Tang, F. H. Liu, and Y. X. Qiao, *Nucl. Instrum. Meth. B* **269** (2011) 209.
- 28 T. M. Robinson and M. L. Jenkins, *Philos. Mag. A* **43** (1981) 999.
- 29 B. N. Singh and S. J. Zinkle, *J. Nucl. Mater.* **206** (1993) 212.
- 30 D. S. Aidhy, C. Lu, K. Jin, H. Bei, Y. Zhang, L. Wang, and W. J. Weber, *Acta Mater.* **99** (2015) 69.

- 31 C. A. English, B. L. Eyre, K. Shoaib, and T. M. Williams, *J. Nucl. Mater.* **58** (1975) 220.
- 32 C. Y. Ho, M. W. Ackerman, K. Y. Wu, T. N. Havill, R. H. Bogaard, R. A. Matula, S. G. Oh, and H. M. James, *J. Phys. Chem. Ref. Data* **12** (1983) 183.
- 33 B. G. Tóth, L. Péter, Á Révész, J. Pádár, and I. Bakonyi, *Eur. Phys. J. B* **75** (2010) 167.
- 34 E. Friedland and M. Fletcher, *Nucl. Instrum. Meth. B* **64** (1992) 242.
- 35 E. Friedland and H. W. Alberts, *Nucl. Instrum. Meth. B* **35** (1988) 244.
- 36 Yu P. Sharkeev, A. N. Didenko, and E. V. Kozlov, *Surf. Coat. Technol.* **65** (1994) 112.
- 37 Yu P. Sharkeev and E. V. Kozlov, *Surf. Coat. Technol.* **158-159** (2002) 219.
- 38 S. Moll, L. Thome, G. Sattonnay, A. Debelle, F. Garrido, L. Vincent, and J. Jagielski, *J. Appl. Phys.* **106** (2009) 073509.
- 39 Y. Zhang, C. M. Wang, M. H. Engelhard, and W. J. Weber, *J. Appl. Phys.* **100** (2006) 113533.
- 40 S. Moll, Y. Zhang, A. Debelle, L. Thomé, J. P. Crocombette, Z. Zihua, J. Jagielski, and W. J. Weber, *Acta Mater.* **88** (2015) 314.

Figure Captions

Fig. 1. (a) Photograph, (b) X-ray Laue diffraction pattern, and (c) 2-D ion channeling scan image of single crystalline Ni. In (c), the angle intervals are 0.072° and 0.144° for the two orthogonal rotating axes; 2 MeV He ion was used in the scan for good angular resolution; different colors represent different backscattering yields.

Fig. 2. SRIM simulations of pure Ni irradiated with 3 MeV Au ions at the fluence of 1×10^{14} ions cm^{-2} . Blue solid curve is the “full cascade mode” damage distribution, green dash-dot curve is the “quick calculation mode” damage distribution, and the red dash curve is the ion distribution. Note that the two modes give the same ion distribution profile.

Fig. 3. (a) RBS channeling spectra of the pristine Ni, NiCo and NiFe, along with the normalized amorphous level. (b) The angular scans of the $\langle 100 \rangle$ channel of these materials extracted from their 2-D scans (as shown in Fig. 1(c) for the case of pure Ni).

Fig. 4. RBS channeling spectra of Ni, NiCo and NiFe, irradiated with 3 MeV Au ions to the fluences from 2×10^{13} to 5×10^{15} ions cm^{-2} .

Fig. 5. Dechanneling parameter profiles as a function of depth at different fluences, from 1×10^{13} to 5×10^{15} ions cm^{-2} , of (a) pure Ni, (b) NiCo and (c) NiFe. The dechanneling parameters as a function of irradiation dose are shown in (d).

Figures

Fig. 1.

Fig. 2.

Fig. 3.

Fig. 4.

Fig. 5.