H⁺-induced irradiation damage resistance in Fe- and Ni-based metallic glass

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Abstract

In this study, use of 40-keV H⁺ ion for irradiating metallic glass Fe₈₀Si_{7.43}B_{12.57} and Ni₆₂Ta₃₈ as well as metallic tungsten (W) at fluences of 1×10^{18} and 3×10^{18} ions/cm², respectively, was investigated. At the fluence of 1×10^{18} ions/cm², a crystalline layer appeared in metallic glass Fe₈₀Si_{7.43}B_{12.57}, with α -Fe as the major crystalline phase, coupled with a little Fe₂B, Fe₃B, and metastable β -Mn-type phase. Fe₈₀Si_{7.43}B_{12.57} exhibited good soft magnetic properties after irradiation. At the fluence of 3×10^{18} ions/cm², Ni₆₂Ta₃₈ was found to be amorphous-based, with a little μ -NiTa and Ni₃Ta phases. No significant irradiation damage phenomenon appeared in metallic glasses Fe₈₀Si_{7.43}B_{12.57} and Ni₆₂Ta₃₈. Blistering, flaking, and other damage occurred on the surface of metallic W, and the root-mean-square (RMS) roughness increased with the increase of fluence. Metallic glass Ni₆₂Ta₃₈ exhibited better resistance to H⁺ irradiation than Fe₈₀Si_{7.43}B_{12.57}, both of which were superior to the metallic W.

Keywords: H⁺ ion beam; radiation damage; metallic glass; metallic W

1. Introduction

The devices in the irradiation environment (e.g., aerospace and nuclear fusion) are subjected to a variety of plasma irradiations, of which proton irradiation is unique; for example, 85% of cosmic irradiation components are protons, and nuclear fusion reactor produces hydrogen and hydrogen isotope ions (deuterium and tritium).¹ The devices in these unconventional areas are constantly exposed to irradiation by protons, affecting their properties and lifetimes; therefore, studies on antiproton irradiation of related materials are very important in the development of space and nuclear technologies. Tungsten is the best-known plasma-facing material. Studies have shown that

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metallic W and a variety of metal alloys produced sputtering, cracking, blistering, delamination, and other damages under ion beam irradiation, which affect the mechanical properties of the materials.^{2,3} Thus, the research interest on irradiation-resistant materials has increased recently.

Metallic glasses, also known as amorphous alloys, have a metastable structure of long-range disorder and short-range order as well as good physical, chemical, and mechanical properties (e.g., high strength and corrosion resistance).⁴⁻⁶ The inherent disordered structure of metallic glass is conducive to resist ion irradiation-induced damage⁷ and it helps to avoid damages such as cracking, blistering, and delamination in tungsten under ion beam irradiation, and hence they may be used in the irradiation environment of fusion reactors or outspace. Fe-based metallic glass exhibits high glass transition temperature, good corrosion resistance, and low activation; Ni-based metallic glass has high glass-forming ability, high thermal stability, high strength, and excellent corrosion resistance.^{8,9} The addition of corrosion-resistant element Ta further enhances the thermal stability and corrosion resistance of Ni-based metallic glass.¹⁰ Therefore, studies on the H⁺ irradiation-resistant properties of Fe- and Ni-based metallic glasses are of utmost importance to the selection of aerospace and nuclear fusion materials.

This study used H⁺ beam for irradiating Fe₈₀Si_{7.43}B_{12.57} and Ni₆₂Ta₃₈ as well as metallic W, which is currently the most promising plasma-facing material and to investigate the behavior of resistance to irradiation damage in metallic glass by comparing the change in the structure and performance of metallic glasses and W before and after irradiation.

2. Experimental Methods

Metallic glass strips of Fe₈₀Si_{7.43}B_{12.57} and Ni₆₂Ta₃₈ with a uniform thickness of 35 μm and widths of 10 and 2 mm, respectively, are prepared by quenched melt spinning. Tungsten with a purity of 99.95% is machined into thin blocks with dimensions of 8 × 8 × 2 mm. After mechanical cleaning, these blocks are put into acetone solution for ultrasonic cleaning. Irradiation experiments were conducted on a low-energy, high-current ion implanter provided by the Institute of Heavy Ion Physics, Peking University. During irradiation, the sample holder was water-cooled. The proton beam spot size was Φ 100 mm, ion beam flux density was 5 mA, duty ratio was 1/10, energy of H⁺ ion was 40 keV, and irradiation fluencies were 1 × 10¹⁸ and 3 × 10¹⁸ ions/cm².

SRIM program was used for the simulation of ion irradiation to calculate the radiation damage parameters. X-ray diffraction (XRD) was used to measure the phase structure change of metallic

glass before and after irradiation; a transmission electron microscope (TEM) was used to measure the changes of structure, component, and the distribution of hydrogen bubble in metallic glass after irradiation; and a scanning electron microscope (SEM) and atomic force microscope (AFM) were used to determine the surface morphology and root-mean-square (RMS) roughness of metallic glass and W, respectively. The magnetic properties of $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$ were analyzed by a vibrating sample magnetometer (VSM).

3. Results and discussion

3.1 SRIM simulation results

On the basis of Monte Carlo approach, SRIM was used to simulate the interaction of ions with materials by tracking a large number of incident ions. Table 1 shows the range, target density, displacement per atom (dpa), and sputtering yields of H^+ irradiation in the aforementioned three materials. It is evident from the table that the range of H^+ in metallic W was the least. Equi-energy H^+ ions were irradiated on metallic glasses and metallic W, and faster ions remained in the metallic W after a series of cascading collisions; H^+ ions transferred all of their energy in a shorter range to the W atoms. H^+ irradiation might result in more severe irradiation damage to metallic W than to metallic glasses.

Dpa is an important measure of irradiation-induced damage on materials.¹¹ According to Table 1, the ratio $((dE/dx)_n/\rho)$ of the linear energy transfer (LET) per ion to target by nuclear processes and target density in the three materials were close, and their displacement threshold energies (E_d) were quite different, leading to a significant impact on dpa. At the fluence of 1×10^{18} ions/cm², the dpa peaks corresponding to $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$, $\text{Ni}_{62}\text{Ta}_{38}$, and W were 6.22, 4.49, and 2.80, respectively. The dpa peaks of these three materials during proton irradiation with the energy of 40 keV were higher than those with the energy of 250 keV, indicating that the H^+ irradiation with a lower energy may cause more damage to the materials. In addition, the sputtering yields of metallic glass $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$ was less than that of metallic W. It was clear that the sputtering from $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$ caused less pollution to the plasma environment.

3.2 Evolution of surface morphology with H^+ irradiation in metallic glass

Figure 1 shows the SEM images of $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$, $\text{Ni}_{62}\text{Ta}_{38}$, and W before and after H^+ irradiation at different fluences. It is evident from the figure that no obvious signs of irradiation damage were observed on the surface of $\text{Ni}_{62}\text{Ta}_{38}$ at different fluences. At the fluence of 1×10^{18}

ions/cm², a small amount of impurities appeared on the surface of Fe₈₀Si_{7.43}B_{12.57}. When the fluence reached 3×10^{18} ions/cm², white crystal particles were found to be uniformly distributed on the surface of Fe₈₀Si_{7.43}B_{12.57}, but there was no cracking, blistering, delamination, and other severe irradiation damage phenomena. At the fluence of 1×10^{18} ions/cm², the surface of metallic W became rough, with obvious grain boundaries and local blistering damage. The bubbles were regularly shaped, mostly rounded. When the fluence reached 3×10^{18} ions/cm², almost the entire treated area of metallic W suffered blistering damage, and the surface became rougher. The bubbles were irregularly shaped, with significantly increasing diameters. It is also evident from the figure that the bubble cover had mostly separated itself from W surface, and chipping and shedding occurred at the edges of the bubble cover.

The H⁺ ions were mainly captured in grain boundaries, vacancies, and other defects inside metallic W. These ions combined with the vacancies to move and gather, and thus a large number of H⁺ ions gradually gathered into hydrogen bubbles. When the pressure inside the bubble was higher than the yield strength of metallic W, surface blistering was caused by deformation.¹² With the continued accumulation of fluence, the hydrogen bubbles inside W continued to absorb the surrounding H⁺ ions and vacancies, or were merged to grow into larger hydrogen bubbles. Therefore, the diameter and number of bubbles increased at the fluence of 3×10^{18} ions/cm². Constant bombardment by H⁺ increases the temperature and pressure inside the hydrogen bubbles, and the stress generated by internal pressure was released by breaking the bubble cover away from the surface of metallic W (Fig. 1). Because of the inherently disordered structure and the absence of grain boundaries, dislocations, and other defects, metallic glasses Fe₈₀Si_{7.43}B_{12.57} and Ni₆₂Ta₃₈ were capable of accommodating more H⁺ ions, which were not easy to gather in the metallic glasses; therefore, there was no obvious blistering damage on the surface of metallic glasses.

Figure 2 shows the AFM surface morphology images of Fe₈₀Si_{7.43}B_{12.57}, Ni₆₂Ta₃₈, and W. After H⁺ irradiation at different fluences, the surfaces of these three materials were roughened, and the RMS roughnesses increased with the increase of fluence. At the fluence of 3×10^{18} ions/cm², the change in RMS roughness of Ni₆₂Ta₃₈ was the least (2.5 nm), the RMS roughness of Fe₈₀Si_{7.43}B_{12.57} reached 14.7 nm because of crystal formation, and that of metallic W was the highest (15.8 nm) because of blistering-induced damage. The evolution of ion beam-irradiated solid surface was the result of a competition between the roughening process caused by ion bombardment and the smoothing process caused by material transport.¹³ It was clear that the roughening process resulted

from ion bombardment played a predominant role in metallic glass and W with the increase of fluence.

3.3 Effect of phase structure with H^+ irradiation in metallic glass

Figure 3 shows the GXRD patterns of metallic glasses $Fe_{80}Si_{7.43}B_{12.57}$ and $Ni_{62}Ta_{38}$. It is evident from the figure that the original $Fe_{80}Si_{7.43}B_{12.57}$ and $Ni_{62}Ta_{38}$ exhibited good amorphous structure. As expected, Figure 3a shows a crystal diffraction peak at $2\theta \approx 45^\circ$ in $Fe_{80}Si_{7.43}B_{12.57}$ at the fluence of 1×10^{18} ions/cm². When the fluence reached 3×10^{18} ions/cm², the crystal diffraction peak increased significantly, indicating the occurrence of crystallization in $Fe_{80}Si_{7.43}B_{12.57}$,¹⁴ and was identified as α -Fe phase by the PDF card. As shown in Figure 3b, $Ni_{62}Ta_{38}$ maintained good amorphous state after H^+ irradiation at different fluences. Width of the supercooled liquid region affects anti-crystallization behaviors of metallic glasses under irradiation. A wider supercooled liquid region for the metallic glass indicates higher thermal stability against crystallization.¹⁵ The metallic glass $Fe_{80}Si_{7.43}B_{12.57}$ is unstable, and hence its glass transition temperature cannot be measured definitely, whereas $Ni_{62}Ta_{38}$ is stable and has a wide supercooled liquid region (63 K¹⁶). ~~On the contrary,~~ The onset temperature of crystallization of $Fe_{80}Si_{7.43}B_{12.57}$ (779 K) is lower than that of $Ni_{62}Ta_{38}$ (1056 K¹⁶). The metallic glass $Fe_{80}Si_{7.43}B_{12.57}$ is more easily crystallized under the same experimental conditions, and the phase structure of $Ni_{62}Ta_{38}$ is more stable. It is clear that the metallic glass with high onset temperature of crystallization has better resistance to H^+ irradiation-induced damage.

$Fe_{80}Si_{7.43}B_{12.57}$ at the fluence of 1×10^{18} ions/cm² and $Ni_{62}Ta_{38}$ at the fluence of 3×10^{18} ions/cm² were chosen for cross-sectional TEM analysis. Figure 4 shows the results of TEM analysis of $Fe_{80}Si_{7.43}B_{12.57}$. In Figure 4a, region A is located in the crystalline layer and region B is near the ion range. The cross-sectional TEM morphology of $Fe_{80}Si_{7.43}B_{12.57}$ in Figure 4a shows a clear crystalline layer within the range of approximately 90 nm from its surface. A large number of hydrogen bubbles were distributed in the crystalline layer, whose shapes were mostly circular, but rarely irregular. The average size range of the hydrogen bubbles was approximately 4–7 nm. The hydrogen bubbles with smaller sizes (~1 nm) were distributed at the end of the crystalline layer. No large hydrogen bubble was formed in the near surface layer of $Fe_{80}Si_{7.43}B_{12.57}$, and hence no blistering, cracking, or other damage phenomena occurred in the SEM morphology. It could be seen from the EDS line scan graph of $Fe_{80}Si_{7.43}B_{12.57}$ in Figure 4b that element Si was linearly distributed along the direction of depth, but element Fe was slightly decreased within the range of 90 nm from the surface. Because the evaluation of element B by TEM-EDS is significantly difficult, analysis results of only Fe and Si are presented. The through-focus imaging shown in Figure 4c and d

revealed Fresnel contrast at the round defects. In an under-focused condition (Fig. 4c), the bubbles exhibited bright Fresnel fringes, whereas in an over-focused condition (Fig. 4d), they exhibited dark Fresnel fringes. This phase contrast was associated with a change in the electronic potential, which was manifested as bubbles or voids.¹⁷ Figure 4e shows a high-resolution image of the hydrogen bubbles at region A in Figure 4a, where the interconnected growth process of the two hydrogen bubbles with larger diameters (~6 to 7 nm) could be clearly seen. In our previous studies, the interconnected growth phenomenon of helium bubbles was also observed after irradiation of 500-keV He²⁺ irradiation on Zr-based metallic glass.¹⁸

Figure 5 shows the high-resolution images and the corresponding Fourier spectra for different positions (regions A and B in Figure 4a) from the surface of Fe₈₀Si_{7.43}B_{12.57}. It is evident from the figure that the crystalline phases of Fe₈₀Si_{7.43}B_{12.57} after irradiation, including Fe₂B phase with tetragonal structure (Fig. 5a), Fe₃B phase with orthogonal structure (Fig. 5c), and metastable β -Mn-type phase (Fig. 5b), were observed in addition to α -Fe. Figure 5d shows that the arrangement of atoms near the ion range in Fe₈₀Si_{7.43}B_{12.57} was disordered, and the Fourier transform spectrum was a diffuse halo, which indicates that there was also amorphous state near the range.

Metallic glasses are thermodynamically metastable structures; therefore, when activation energy exceeds the glass transition temperature, metallic glasses might undergo a transition to steady-state crystalline phase.¹⁹ The surface of Fe₈₀Si_{7.43}B_{12.57} was crystallized by H⁺ irradiation, mainly because the temperature reached its glass transition temperature. In the experiment, the temperature of the sample holder did not exceed 100°C, but the local or transient temperature of Fe₈₀Si_{7.43}B_{12.57} might be higher than its glass transition temperature.²⁰ The α -Fe in metallic glass Fe₈₀Si_{7.43}B_{12.57}, as a main crystalline phase, was precipitated directly from the amorphous matrix, which often appeared in other Fe-Si-B metallic glasses.^{21,22} Metastable β -Mn-type phase was a complex phase that appeared in the process of crystallization. The β -Mn-type phase was composed of Frank–Kasper-type polyhedron, and its element concentration was similar to the one in amorphous matrix. The Frank–Kasper-type polyhedron, a common type in alloys, appeared as a local structure in alloys. The precipitation of β -Mn-type phase from Fe₈₀Si_{7.43}B_{12.57} only needed a short-range atomic diffusion process.²³ Ion implantation was a nonequilibrium process and a metastable crystalline phase was easily formed in Fe₈₀Si_{7.43}B_{12.57}.

Once nucleation occurred on the surface of metallic glass Fe₈₀Si_{7.43}B_{12.57}, the energy of more

incident H^+ ions would be absorbed in the process of crystal growth. The defects generated from crystal formation, such as grain boundaries, would capture the later implanted H^+ ions, resulting in H^+ retention in the crystalline layer; the H^+ ions were combined with the vacancies to move and gather into hydrogen bubbles. The holes formed by hydrogen bubbles slightly reduced the proportion of Fe within the crystallization range of $Fe_{80}Si_{7.43}B_{12.57}$. Beyond the crystalline layer range, $Fe_{80}Si_{7.43}B_{12.57}$ remained in a good amorphous state, and no H^+ ions gather in large numbers to form hydrogen bubbles, possibly due to the disordered structure of metallic glass and the high solubility and diffusivity of H^+ in the solid.¹² In ref. [24], under the 250-keV H^+ irradiation of $Fe_{80}Si_{7.43}B_{12.57}$ at the same fluence, different depths from the surface were kept in an amorphous state. The beam current with the energy of 40 keV (~ 5 mA) was found to be hundreds of times higher than the one with 250 keV (~ 10 μ A), and hence the damage cascade was more serious when the energy of the beam current was 40 keV. The damage cascade induced by ion irradiation increased the atomic mobility within the irradiated region, which ~~in turn~~ lead to crystallization.²⁵ Moreover, the thermal effect generated from the energy deposition of a large amount of H^+ ions in the surface layer may lead to the increase of local or transient temperature of the metallic glasses higher than their glass transition temperature, thereby facilitating crystallization. The synergistic effect of thermal annealing and a thermal effect from irradiation result in crystallization of $Fe_{80}Si_{7.43}B_{12.57}$.

Figure 6 shows the TEM results of $Ni_{62}Ta_{38}$ at the fluence of 3×10^{18} ions/cm². It can be observed from the cross-sectional morphology in Figure 6a that a phenomenon of crystallization appears within the range of approximately 85 nm from the surface. From the SAD image of different regions in Figure 6a, it can be observed that many crystalline phases were deposited on the amorphous matrix in the crystalline layer, and the diffraction pattern near the range was a diffuse amorphous halo. Figure 6b shows the EDS line scan graph for the composition of $Ni_{62}Ta_{38}$ along the direction of depth, from which it can be seen that element Ni in the near-surface region was undulating, indicating that the local element distribution of $Ni_{62}Ta_{38}$ was uneven.

Figure 7 shows the high-resolution image and Fourier transform spectrum of the crystalline layer of $Ni_{62}Ta_{38}$ (region A in Figure 6a). It is evident from the figure that the observed crystalline phases of $Ni_{62}Ta_{38}$ were μ -NiTa and Ni_3Ta . According to the Ni-Ta phase diagram, the most common crystalline phases of $Ni_{62}Ta_{38}$ were μ -NiTa and Ni_2Ta . Because the experiment used micro-TEM observation, Ni_2Ta phase could not be observed. Ion implantation was an unbalanced

process, the composition distribution in metallic glass after irradiation was uneven, and the ratio of Ni to Ta possibly changed, and thus Ni_3Ta phase, which did not exist in the $\text{Ni}_{62}\text{Ta}_{38}$ phase diagram, could be experimentally observed. Crystalline phases $\mu\text{-NiTa}$ and Ni_3Ta formed in $\text{Ni}_{62}\text{Ta}_{38}$ were less in content and exist only within the range of 85 nm from the surface, and hence its GXRD spectrum after irradiation was an amorphous package, without the formation of crystalline phases. A combination of thermal annealing and cascade effect from irradiation led to the crystallization of $\text{Ni}_{62}\text{Ta}_{38}$. In addition, the uneven distribution of the Ni and Ta atoms contributed to the crystallization.

3.4 Effect of H^+ irradiation on the magnetic properties of $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$

Figure 8 shows the hysteresis loop of $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$ before and after H^+ irradiation at the fluence of 3×10^{18} ions/cm². It can be seen that the original metallic glass $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$ had low coercivity, high permeability, high saturation magnetization, and other soft magnetic characteristics. At the fluence of 3×10^{18} ions/cm², with a slight increase of coercivity, but no significant change in saturation magnetization, $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$ still exhibited appreciable soft magnetic properties. The coercivity of a soft magnetic material was structure-sensitive. The analysis of phase structure using GXRD shows that there was a clear crystallization phenomenon in $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$ at the fluence of 3×10^{18} ions/cm². The occurrence of crystalline phase played a pinning role in the movement of the magnetic domains, which increased the coercivity of $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$.¹³ In addition, stable crystalline phase Fe_2B exhibited high magnetic anisotropy, which also lead to the increase of coercivity^[26].

After H^+ irradiation, crystallization occurred in both metallic glasses $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$ and $\text{Ni}_{62}\text{Ta}_{38}$. The difference was that the amorphous matrix of $\text{Ni}_{62}\text{Ta}_{38}$ contained a small amount of crystal particles deposited on it, which were still amorphous-based with a good amorphous state at the fluence of 3×10^{18} ions/cm² according to the GXRD patterns. A crystalline phase clearly appeared in $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$ at the fluence of 3×10^{18} ions/cm². After H^+ irradiation at different fluences, there was no blistering, flaking, delamination, or other irradiation-induced damage on the surface of the two metallic glasses, but crystal-precipitated white spots appeared in $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$ at the fluence of 3×10^{18} ions/cm². Blistering-induced damage occurred in metallic W at the fluence of 1×10^{18} ions/cm², which was particularly more severe at the fluence of 3×10^{18} ions/cm². Although higher dpa values and crystallization appeared in metallic glasses $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$ and $\text{Ni}_{62}\text{Ta}_{38}$, no serious irradiation-induced damage occurred. Thus, the H^+ irradiation-resistant

properties of the two metallic glasses were better than those of metallic W.

4. Conclusion

After low-energy and strong-beam proton irradiation in metallic glasses, different degrees of crystallization occurred in the surface layers of Fe- and Ni-based metallic glasses because of the increase of atomic mobility induced by damage cascade, thermal effect generated from energy deposition, and the local uneven distribution of composition. The main crystalline phase in $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$ at the fluence of 1×10^{18} ions/cm² was α -Fe, coupled with a little Fe_2B , Fe_3B , and β -Mn-type phase; the hydrogen bubbles and the interconnected growth phenomenon of hydrogen bubbles were observed in the crystalline layer. After irradiation at the fluence of 3×10^{18} ions/cm², uniformly distributed crystalline spots appeared on the surface of $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$, still exhibiting appreciable soft magnetic properties. $\text{Ni}_{62}\text{Ta}_{38}$ was still amorphous-based at the fluence of up to 3×10^{18} ions/cm², and a small amount of μ -NiTa and Ni_3Ta phases appeared, but the surface morphology did not change significantly. Blistering-induced damage appeared on the surface of metallic W, and the phenomenon that the bubble cover was separated from the surface was observed. The RMS roughnesses of the three materials increased with the increase of fluences, wherein the change in that of $\text{Ni}_{62}\text{Ta}_{38}$ was the least. The dpa peaks of metallic glasses $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$ and $\text{Ni}_{62}\text{Ta}_{38}$ were stronger than that of metallic W. After irradiation, different degrees of crystallization occurred, but no blistering, cracking, or other significant irradiation-induced damage appeared. The H^+ irradiation-resistant properties of $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$ and $\text{Ni}_{62}\text{Ta}_{38}$ were superior to those of metallic W; furthermore, $\text{Ni}_{62}\text{Ta}_{38}$ exhibited better H^+ irradiation-resistant properties than $\text{Fe}_{80}\text{Si}_{7.43}\text{B}_{12.57}$. Fe- and Ni-based metallic glasses may have good prospects in irradiation environments.

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Figures captions:

Figure 1. SEM images of pristine and 40-keV H^+ ion-irradiated $Fe_{80}Si_{7.43}B_{12.57}$, $Ni_{62}Ta_{38}$, and metallic W at different fluences.

Figure 2. AFM images of pristine and 40-keV H^+ ion-irradiated $Fe_{80}Si_{7.43}B_{12.57}$, $Ni_{62}Ta_{38}$, and metallic W at different fluences (x scale 2 μm /division, z scale 44 nm/division).

Figure 3. GXRD pattern of pristine and 40-keV H^+ ion-irradiated $Fe_{80}Si_{7.43}B_{12.57}$ and $Ni_{62}Ta_{38}$ at different fluencies.

Figure 4. (a) Cross-sectional TEM morphology and (b) EDS analysis image, (c) under-focused and (d) over-focused morphology of hydrogen bubbles, and (e) interconnected growth process of the two hydrogen bubbles of $Fe_{80}Si_{7.43}B_{12.57}$ at the fluence of 1×10^{18} ions/cm².

Figure 5. HRTEM image and the corresponding diffraction images (SAD) of (a), (b), (c) crystalline layer region, (d) the area near the ion range in $Fe_{80}Si_{7.43}B_{12.57}$ at the fluence of 1×10^{18} ions/cm².

Figure 6. (a) Cross-sectional TEM morphology, (b) EDS analysis image of $Ni_{62}Ta_{38}$ at the fluence of 3×10^{18} ions/cm².

Figure 7. HRTEM image and the corresponding diffraction images (SAD) of crystalline layer region (Zone A Figure 6) in $Ni_{62}Ta_{38}$ at the fluence of 3×10^{18} ions/cm².

Figure 8. Hysteresis loops of $Fe_{80}Si_{7.43}B_{12.57}$ before and after irradiation at the fluence of 3×10^{18} ions/cm².

Table title:

Table 1. Ranges, sputtering yields, displacement energy, target density, and dpa induced by different fluences of H^+ irradiation using SRIM program simulation in $Fe_{80}Si_{7.43}B_{12.57}$, $Ni_{62}Ta_{38}$, and metallic W.

Figure 1.

Figure 2.

Figure 3.

Figure 4.

Figure 5.

Figure 6.

Figure 7.

Figure 8.

Table 1.

Energy	Sample	Fe ₈₀ Si _{7.43} B _{12.57}	Ni ₆₂ Ta ₃₈	W
40keV	Projected range (μm)	0.23	0.21	0.15
	Target Density (atoms/cm ³)	8.46×10 ²²	6.77×10 ²²	6.34×10 ²²
	Sputtering Yield (Atoms/Ion)	0.000833	0.000934	0.000967
	(dE/dx) _n (eV/cm)	4.60×10 ⁷	3.91×10 ⁷	4.00×10 ⁷
	Displacement energy (eV)	35	55	90
	dpa (1×10 ¹⁸ ions/cm ²)	6.22	4.49	2.80
250keV	dpa (1×10 ¹⁸ ions/cm ²)	4.87	2.97	1.74