Article

Microstructural Evolution and the Irradiation Sensitivity of Different Phases of a High Nb-Containing TiAl Alloy under He Ions Implantation at Room- and Elevated Temperatures

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Abstract: A high Nb-containing TiAl alloy Ti-45Al-8.5Nb-(W, B, Y) with nearly lamellar microstructure has been irradiated by 200 kV He²⁺ to a fluence of 1×10^{21} ions/m² with a dose of about 1.1 dpa at 298 K and 773 K in this work. It is found that an amorphous layer formed on the surface, and no helium bubbles can be observed in the alloy after room temperature ion implantation. The surface roughness of the alloy increases significantly with the bombardment of helium ions, indicating that the ion implantation increases the surface defects. The high-temperature ion implantation leads to the phenomenon of blistering on the alloy surface, and helium bubbles are observed in both α_2 and γ phases of the alloy irradiated at 773 K. The average size of the helium bubbles in the α_2 (15–20 nm) is larger than that in the γ (3–5 nm) phase, while the helium bubble density is opposite. Moreover, the growth mechanism of helium bubble is also investigated. By means of nanoindentation, an obvious irradiation hardening phenomenon is measured after the room temperature ion implantation. In addition, the irradiation sensitivity of different phases is also discussed in this work. The results show that the γ phase has the highest irradiation sensitivity, α_2 phase second, β phase minimum. The results of this work, especially microstructure evolution and the evaluation of phase-related irradiation sensitivity during ion implantation, can be expected to provide experimental evidence for the applications of TiAl intermetallic compounds in the nuclear industries.

Keywords: high Nb-containing TiAl; radiation damage; microstructural evolution; nanohardness; irradiation sensitivity

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1. Introduction

TiAl alloys with ordered structure have received special attention in high-temperature applications due to their low density and relatively high temperature strength, creep properties and good oxidation resistance [1–4]. As advanced high-temperature structural materials, TiAl alloys were initially developed for potential applications as aerospace engines and automotive components [5–7]. To further extend the service range, alloying has been successfully used as an effective way to improve the performance of TiAl alloys. Nb has been proven to improve the oxidation resistance and increase the using temperature of TiAl alloys [8–10]. Therefore, the alloying with high Nb content is expected to be one of the promising development directions of future TiAl alloys.

With the development of nuclear energy, the advanced nuclear systems such as the very high-temperature reactor (VHTR) and the gas-cooled fast reactor (GFR) are also expected to develop a higher operating temperature and higher doses than that of light water reactors (LWR) [11,12]. International Generation IV Initiative (GIF) is committed to finding new and reliable materials for future nuclear reactors. The selection of metallic

materials is essential in nuclear industry. Meanwhile, the stability of structural materials under irradiation has received more and more attention. As the most important form of radiation damage problem for metallic materials, the creation of helium is produced by transmutation reactions. The formation and growth of helium bubbles depend on the neutron spectrum, fluence and materials [13]. Therefore, the radiation effects of neutron irradiation on materials can be simulated by helium ion radiation. A great deal of work has been done in the nuclear industry to illustrate the nucleation and growth of helium bubbles [14,15]. The relationship between microstructure and helium bubble formation of metallic materials has attracted considerable attention during past decades.

Based on the excellent radiation resistance and low neutron activation, TiAl alloys are considered as potential advanced structural materials for the Generation IV Initiative [16–18]. TiAl intermetallic compounds can be used to deal with different conditions for various extreme environments in advanced nuclear systems by microstructure regulation, such as He gas turbine, tubing in nonirradiated regions and the control rod in the irradiated core region [17,19]. Due to helium produced in a reactor, the radiation defects caused by helium have a serious influence on mechanical properties and result in hardening, swelling and blistering of structural materials [20]. Magnusson et al. studied the creep behavior of helium-irradiated TiAl alloy (ABB-2) and found the strong embrittlement phenomenon by helium [12]. Significant irradiation hardening has also been observed in TiAl alloy [20]. At present, scholars are mainly concerned with the irradiation performance of TiAl alloys after irradiation. However, the study of the TiAl alloy microstructural evolution during irradiation is still limited, especially for the evolution of helium bubbles in high Nb-containing TiAl alloys. As a β -solidified, high Nb-containing TiAl alloy, Ti-45Al-8.5Nb-(W, B, Y) can be expected to be used as structural material in the nuclear industry due to its excellent comprehensive performance. In the service environment of nuclear power, the microstructure and stability of the material are very important, and the microstructure plays a decisive role on the performance. The objective of the present work is to investigate the microstructural evolution of a β -solidified, high Nb-containing TiAl alloys during He irradiation.

2. Experimental Procedures

The experimental β -solidified, high Nb-containing TiAl alloy with the nominal composition of Ti-45Al-8.5Nb-(W, B, Y) (the content of W, B, Y elements are 0.2, 0.2, 0.02 respectively) was prepared by plasma arc melting (PAM) twice to make the composition homogeneous. The specimens used in postimplantation microstructure observation experiments were mechanically polished to a thickness of 2 mm with a rectangular shape. The sample size for implantation at 298 K was 10 mm $\times$ 10 mm $\times$ 2 mm and 20 mm $\times$ 20 mm $\times$ 2 mm for implantation at 773 K. The prepared transmission electron microscope (TEM) samples, as circular discs with a diameter of 3 mm, were subjected to corresponding ion implantation experiments and then observed by transmission electron microscope.

Helium ion implantations were carried out at a MEVVA Source Ion Implanter (China Nuclear Tongchuang (Chengdu) Technology Co., Ltd., Chengdu, China) using 200 kV He²⁺ for homogeneously implanting the samples at 298 K and 773 K. Helium ion implantations were performed directly on specimens for scanning electron microscope (SEM) and TEM observations. The total fluence was 1×10^{21} ions/m². The damage dose and He ions concentration with implantation depth were calculated by the Stopping and Range of Ions in Matter (SRIM) [21] in Figure 1a. Based on the component of the alloy and the total fluence, the maximum damage dose is estimated to be 1.1 dpa, and the peak value of damage occurs at about 765 nm from the sample surface. The maximum implantation depth by He is 1.02 μ m. The He concentration distribution displays a relatively sharp peak at a depth of 810 nm. The formation of helium bubbles is closely related to the concentration distribution of helium ion. The theoretical He ions concentration and defects profiles calculated by SRIM [14] are shown in Figure 1b. It can be observed that the defects peak around 780 nm and the irradiated region can be divided into three regions, as indicated in Figure 1b. Region I corresponds to the area where the ion slows down due to the electron energy loss

process [22]. Peak region II corresponds to the area where the He ions interact with the alloy atoms by the nuclear collision and finally stop in the crystal lattice. Region III is the nonimplanted area.

Figure 1. Depth profiles of the displacement damage (a), helium concentration distribution and vacancies (b) in irradiated samples with helium ions to a fluence of 1×10^{21} ions/m² (I, II and III represent three regions).

Surface roughness was conducted by Atomic Force Microscope (AFM) (Bruker Dimension Icon, Karlsruhe, Germany), which can provide three-dimensional surface morphology features in nano scale. Microstructural observations were carried out by SEM (VEGA-3 TESCAN, Tescan, Brno, Czech Republic), TEM (FEI Tecnai G2 F30, FEI, Hillsborough, America) and high-resolution transmission electron microscopy (HRTEM) analyses. The hardness measurements were performed using a nanoindenter (U9820A Keysight Nano Indenter G200, Agilent Corp, Palo Alto, CA, USA) for irradiated and unirradiated specimens. The continuous stiffness measurement mode was adopted, and the maximum indentation depth was 1100 nm. At least six tests were performed under the same testing condition to determine the hardness and relative standard deviations.

3. Results and Discussion

3.1. Microstructure Evolution

Figure 2 presents the SEM and the corresponding AFM 3D images before and after ion implantation. The as-received alloy exhibits typical fine-grained lamellar microstructure with γ and β phase along the colony boundaries, as shown in Figure 2a. Compared with the original and room temperature injection microstructures, as shown in Figure 2a,c, respectively, no significant microstructure difference can be observed. According to the existing research results [12,20,23], it is believed that the effect of ion implantation on microstructure occurs at the nanoscale. Further increasing the ion implantation temperature to 773 K, the microstructure is shown in Figure 2e. The phenomenon of blistering is observed in the microstructure, which is different from the results in previous studies [24–26]. As for the foaming position, it can be seen that the phenomenon appears at the interior of the lamellar and the lamellar interface. It is indicated that the damages of TiAl under ion implantation are observed both in nanoscale and micronscale.

Figure 2. SEM and AFM images of the irradiated samples with helium ions to a fluence of (a,b) 0; (c,d) 1×10^{21} ions/m² at 298 K; (e,f) 1×10^{21} ions/m² at 773 K.

Quantitative roughness analyzed by atomic force microscopy (AFM) is proposed to denote the degree of irradiation damage, with arithmetic average roughness (Ra) as parameters, and the three coordinates represent the length, width and height, respectively. Figure 2b shows the original AFM 3D images of the experimental alloy that the microstructure of the alloy is smooth and its Ra is 0.69 nm. After the room temperature ion implantation, the AFM 3D image of the alloy is shown in Figure 2d. The remarkable fluctuation and roughness decrease can be observed in the microstructure, and the Ra is 1.89 nm. Compared with the original one, its roughness increases by 174%. It indicates that the ion implantation has a very significant effect on the microstructure of the alloy, which greatly improves the surface roughness of the alloy. AFM 3D of the alloy-irradiated image is shown in Figure 2f, and the Ra is 12.9 nm. Compared with room temperature injection, the roughness of the alloy increases exponentially, which is mainly due to the blistering phenomenon on the surface of the alloy. The roughness increases more than ten

times, indicating that ion implantation is sensitive to temperature, and the effect is more pronounced at high temperature.

3.2. Formation and Growth of Helium Bubbles

TEM images of the specimens before and after helium ion irradiation at 298 K are displayed in Figure 3. Figure 3a shows the untreated sample with a fine lamellar structure arrangement of α_2 and γ phase. Comparing the original microstructure in Figure 3a and microstructure after room temperature irradiation in Figure 3b, no helium bubble is found in the microstructure. However, a transparent film is formed on the surface of the alloy. As a result of direct ion implantation, it is believed that the formation of this film is caused by ion implantation. Furthermore, the diffuse scattering at “1” in Figure 3c indicates the presence of amorphous state, which has not been observed in previous studies [20,27]. This means that the experimental material cannot recover/recrystallize rapidly due to the high activation energy during helium ion implantation [28]. To further observe whether there are helium bubbles in the microstructure, Figure 3d shows the internal microstructure and reduces the influence of contrast. The HRTEM image for position 2 of Figure 3d is shown in Figure 3e with the corresponding Fast Fourier Transform (FFT) image in Figure 3f. It can be seen from the Figure 3d that no helium bubble is found in the lamellar, and the HRTEM image of Figure 3e shows that there is no significant difference in the lattice morphology for the α_2 phase. The diffraction spots in Figure 3f can also explain that there is a single phase in the figure and no precipitated phase. Therefore, it can be considered that helium bubbles are not formed in the alloy after helium ion implantation at room temperature.

Figure 3. TEM images of the irradiated samples with helium ions at 298 K to a fluence of (a) 0, (b,d) 1×10^{21} ions/m² with the diffraction pattern in (c) for position 1 in (b), the HRTEM images for position 2 shown in (e) and the corresponding FFT image in (f).

After He ion irradiation at 773 K, helium bubbles were observed in both the α_2 and γ phases, as shown in Figure 4a. It indicates that the formation of helium bubbles strongly depends on the temperature. It also can be seen that there are significant differences in the morphology of helium bubbles in the two phases. Figure 4b shows the HRTEM images

and the corresponding FFT image of the arrow position in Figure 4a. According to the diffraction spots in Figure 4b, it can explain that there is a single α_2 phase. The contrast between helium bubble area and matrix is obviously different. The average size of helium bubbles in the α_2 is 15~20 nm. After zooming in, it can be seen a large density granular helium bubbles in the γ phase. According to the HRTEM images and the corresponding FFT image of the γ phase, the average size of helium bubbles is 3~5 nm. It indicates that the helium bubble density in the γ phase is much larger than that of α_2 .

Figure 4. TEM images of the irradiated samples at 773 K with helium ions to a fluence of $(a) 1 \times 10^{21}$ ions/m² with the HRTEM images and the corresponding FFT image of the arrow position shown in (b), and the γ phase shown in (c) with the HRTEM images and the corresponding FFT image shown in (d).

Crystal structure is one of the important factors that can influence the helium bubbles formation and growth [29]. Zinkle et al. illustrated that the defect formation in face centered cubic (fcc) metals is easier than that in body centered cubic (bcc) metals, and close-packed hexagonal (hcp) metals show variable behavior [30]. The research shows that in the Ti–Al system, the formation of Ti vacancies is easier than Al vacancies [31]. On another hand, the activation energy of Ti vacancy for diffusion in the α_2 phase is lower than that in γ phase [31], which provides the easiest diffusion path for Ti in α_2 phase. From this point of view, the higher vacancy mobility in the α_2 phase leads to the defect clusters' rapid growth and coalescence, resulting in the larger He bubbles formation in the α_2 phase than that of γ phase. Furthermore, Zhu et al. have indicated that crystal structure plays a dominant role in determining the irradiation response, as formation and growth of helium bubbles, between α_2 and γ phases [29]. As other interstitial elements, helium atoms can also reside in interstitial or substitutional sites of the γ -TiAl [30]. It has been reported that interstitial atoms prefer to occupy octahedral sites “Ti6-type”, which exist in the Do₁₉

structure of α_2 phase [32]. Therefore, helium atoms have higher solubility in the α_2 phase than that of γ phase in the investigated high Nb-containing TiAl alloys [20]. It also indicates that the black contrast of helium bubbles in α_2 phase may be related to the displacement damage [23]. Meanwhile, interstitial elements prefer to migrate in the α_2 phase rather than the γ phase due to different ordered crystal structures [33]. Accordingly, with the large enough pressures to spontaneously swallow the periphery cluster [34], the helium bubbles grow more rapidly and form larger bubbles in the α_2 compared with those in the γ phase.

The process of ion implantation at high temperature results in defects such as helium bubbles in the microstructure and other microstructural changes, as shown in Figure 5. After He ion implantation at 773 K, high-density irradiation-induced dislocation pile-up can be observed in the α_2 and γ phases, as shown in Figure 5a,b. It indicates that the ion implantation increases the defect density in the alloy. Based on the previous experimental results, it can be concluded that the surface defects of the alloy increase with ion implantation, which leads to the increase of alloy surface roughness. For the nucleation positions of helium bubbles, it has been shown that helium bubbles prefer to nucleate at interstitial [29] and defect sites [35]. Figure 5c shows that helium bubbles nucleate on the irradiation-induced dislocations. In addition, helium atoms can also nucleate at lamellar boundaries, as shown in Figure 5d, considering the fact that boundaries always act as sinks to absorb defect clusters in the alloy at high temperatures [27]. For the growth mechanism of helium bubbles, it is found that the helium bubbles can grow by migration and coalescence of bubbles, as illustrated in Figure 5e. The morphology of the helium bubble is approximately circular in general, but a rod-like shape of helium bubble was observed, as shown in Figure 5f. It can be found that there are two growing helium bubbles that form a squeeze on it. This indicates that the morphology of helium bubble is closely related to its location.

Figure 5. TEM images of irradiation-induced dislocations (a,b), helium bubbles nucleate on dislocations (c) and boundaries (d), and growth (e,f) after He ion implantation at 773 K.

Figure 6 illustrates in situ HRTEM images of a single He bubble in the α_2 phase with per-ten-second capture speed after He ion implantation at 773 K. The moiré fringes in the Figure 6 are caused by the helium bubble region. After He ion irradiation, the boundaries between the helium bubbles and the α_2 phase are clearly observed, as indicated in Figure 6a. However, the defects caused by the helium bubble are gradually reduced from ~20 nm to 4 nm, shown in Figure 6b–e, and eventually disappear completely in Figure 5f with the increase of irradiation time. It has been reported that temperature is one of the main factors affecting the size of helium bubbles [20]. The experimental results show that the average size of helium bubbles can be affected by electron irradiation. It indicates that the atoms in the helium bubbles defect can rearrange to reach a stable state by the electron energy.

Figure 6. In situ HRTEM image from (a–f) of a single He bubble during TEM observations in high vacuum with per-ten-second capture speed after He ion implantation at 773 K.

3.3. Irradiation Sensitivity of Different Phases

In addition to microstructure evolution, ion implantation inevitably preserves the impact on mechanical performance of the experimental TiAl alloy. In order to understand the sensitivity of each phase after ion implantation, nanoindentation experiments were carried out at different locations of the experimental high Nb-containing TiAl alloy. In the nanoindentation experiment, more than six positions were randomly selected for different phases in the range of all samples. The experimental results are shown in Table 1.

Table 1. The nanohardness of as-cast alloy at different positions before and after ion implantation (GPa).

Position	Before Implantation	Room Temperature Implantation	Elevated Temperature Implantation
Lamellar Colony	5.79 ± 0.10	7.16 ± 0.08	5.35 ± 0.55
γ phase	5.34 ± 0.15	7.05 ± 0.01	5.31 ± 0.35
β phase	6.63 ± 0.49	7.38 ± 0.04	5.72 ± 0.01

Figure 7 shows the nanoindentation hardness of the experimental high Nb-containing TiAl alloy. The indicator A shown in Figure 7a corresponds to lamellar structure. The phase-related nanoindentation hardness curves of the experimental alloy before and after helium ion implantation are compared in Figure 7b. The nanohardness of the lamellar structure before injection first increases, then decreases, and finally tends to be stable. It has a peak within the depth of 100 nm, then gradually decreases and finally tends to be stable. The average value of lamellar structure nanohardness is 5.79 ± 0.10 GPa. After the room temperature ion implantation, the overall trend of the lamellar structure is consistent with but higher than that before injection, and the average value of lamellar structure nanohardness is 7.16 ± 0.08 GPa. The nanohardness increases by 23.7% compared with that of the nonimplanted one. The results indicate that the lamellar structure of the alloy has obvious irradiation hardening phenomenon after room temperature ion implantation. The emergence of irradiation hardening phenomenon has two main causes. Firstly, prevalent defects, such as helium bubble clusters and dislocations, are produced in the alloy after ion implantation [17]. At room temperature, helium clusters and dislocations are relatively stable. Secondly, the irradiation hardening phenomenon occurs due to the accumulation of helium ions in the alloy [22].

Figure 7. Nanoindentation hardness of the experimental high Nb-containing TiAl alloy. (a) The position of nanoindentation tests in the alloy: A lamellar colony, (b) The curve of nanohardness with depth in lamellar colony before and after ion implantation of as-cast alloy.

The nanohardness curve of the alloy after ion implantation at high temperature shows different characteristics from that of room temperature implantation. Its nanohardness shows a gradual upward trend within 300 nm, but the overall level is lower than that of room temperature implantation and nonimplantation samples. After 300 nm, its trend tends to be consistent with and slightly higher than that of nonimplantation samples. After ion implantation at 773 K, the nanohardness of the lamellar colony is calculated to be 5.35 ± 0.55 GPa. The decrease rate is 7.6% compared with the nonimplanted specimen. It indicates that the phenomenon of irradiation hardening is not observed in the lamellar structure of the alloy at high temperature. This means that the irradiation hardening of the alloy is closely related to the temperature. At 773 K, the defects in the alloy are not stable and absorbed by the dislocations [17]. The defect density in the alloy decreases remarkably, which leads to the disappearance of the irradiation hardening. In the present work, the reduction or disappearance ultimate temperature of the TiAl alloy radiation hardening is different from the results in [17,20]. This is mainly due to the change of phase constitutes, which is primarily determined by the alloying composition.

The indicator B shown in Figure 8a corresponds to lamellar structure, γ phase. The phase-related nanoindentation hardness curves of the experimental alloy before and after helium ion implantation are compared in Figure 8b. The average value of γ phase nanohardness is 5.34 ± 0.15 GPa. Compared with the lamellar structure, the nanohardness of the γ phase is slightly lower, which mainly due to the lower phase intensity of the γ phase

than the α_2 phase [36]. After the room temperature ion implantation, the average value of the γ phase nanohardness is 7.05 ± 0.01 GPa. The nanohardness increases by 32.02% compared with that of the nonimplanted γ phase. It indicates that the γ phase has obvious irradiation hardening phenomenon after room temperature ion implantation. Comparing with that of the lamellar structure, the irradiation hardening rate of the γ phase is higher, indicating that the γ phase is more sensitive to the ion implantation than that of α_2 phase and it is more likely to induce radiation hardening. According to the above experimental results, the helium bubble density in the γ phase is much larger than that of α_2 . It means that there is a correlation between the sensitivity of radiation damage and the density of defects in different phases. Because the solid solubility of the helium ion in the α_2 phase is higher, it also indicates that the effect of helium ion concentration on irradiation hardening is the secondary reason, and the defect density is the dominant reason. The results show that high radiation tolerance of α_2 is determined by its ordered crystal structure D0₁₉ [37]. In other words, the radiation resistance of different phases is determined by their crystal structure. After ion implantation at 773 K, the nanohardness of the γ phase was calculated to be 5.31 ± 0.35 GPa. The decrease rate was 0.5% compared with that of the nonimplanted one. It indicates that the phenomenon of irradiation hardening is not observed in the γ phase at high temperature.

Figure 8. Nanoindentation hardness of the experimental high Nb-containing TiAl alloy. (a) The position of nanoindentation tests in the alloy: B γ phase. (b) The curve of nanohardness with depth in the γ phase before and after ion implantation of as-cast alloy.

The indicator C shown in Figure 9a corresponds to the β phase. The phase-related nanoindentation hardness curves of the experimental alloy before and after helium ion implantation are compared in Figure 9b. The average value of β phase nanohardness in as-received alloy is 6.63 ± 0.49 GPa, as illustrated in Figure 9b. Compared with those of the lamellar structure and γ phase, it is found that the β phase has the highest strength. The average value of nanohardness of the β phase in the alloy after room temperature ion implantation is 7.38 ± 0.04 GPa. The nanohardness of β phase in the alloy ion implanted at room temperature increases by 11.3% compared with that of the one without implantation. It indicates that the irradiation hardening phenomenon is also observed in the β phase after room temperature ion implantation. The nanohardness of β phase in the alloy after ion implantation at 773 K is calculated to be 5.72 ± 0.01 GPa. The decrease rate of nanohardness is 13.7% compared with that of the nonimplanted β phase. It indicates that the phenomenon of irradiation hardening is not observed in the β phase of the alloy ion implanted at 773 K. At room temperature, the irradiation hardening phenomenon in the β phase is caused by the existence of the stable helium clusters and dislocations defect [17]. At high temperature, the defects in the β phase are believed to become unstable and are absorbed by dislocations, resulting in a decrease in the hardness of the β phase.

Figure 9. Nanoindentation hardness of the experimental high Nb-containing TiAl alloy. (a) The position of nanoindentation tests in the alloy: C β phase. (b) The curve of nanohardness with depth in β phase before and after ion implantation of as-cast alloy.

The nanohardness of the lamellar structure and γ and β phases increases at different degrees after ion implantation at room temperature, which indicates that the alloy has obvious irradiation hardening during helium ion implantation at room temperature. However, the β phase is less sensitive compared with those of the lamellar structure and γ phase. That is to say, the resistance of the β phase to ion implantation at room temperature is the strongest among the crystalline phases in the experimental high Nb-containing TiAl alloy. The γ phase is the most sensitive to ion implantation, and the sensitivity of the α_2 phase lies between the γ and β phases. The comparison of nanohardness in different phases gives evidence that the sensitivity of the irradiation hardening is related to the crystal structure. The smallest increase of nanohardness indicates the strongest resistance to radiation damage of the β phase during ion implantation at room temperature. The crystal structure difference is believed to determine the radiation damage performance of γ , α_2 and β constituents in the experimental high Nb-containing alloy.

4. Conclusions

Ti-45Al-8.5Nb-(W, B, Y) alloy was irradiated by 200 kV He^{2+} to a fluence of 1×10^{21} ions/ m^2 with a dose of about 1.1 dpa at room and elevated temperature in this work. After room temperature ion implantation, an amorphous layer formed on the surface, and the surface roughness of the experimental high Nb-containing TiAl alloy increased significantly. Blistering phenomenon was observed in the alloy after ion implantation at 773 K. The average size of helium bubbles grew more rapidly and formed larger bubbles in the α_2 phase compared with those in the γ phase due to the different interstitial vacancies. It was found that helium bubbles prefer to nucleate at defects including lattice vacancy, dislocations and lamellar boundaries. Electron energy made the helium bubble defects atoms rearrange in the alloy. Nanoindentation results show that the alloy has an obvious radiation hardening phenomenon after room temperature ion implantation. However, the radiation hardening phenomenon disappeared after high temperature ion implantation at 773 K. In addition, the irradiation sensitivity of different phases was also discussed in this work. The results show that the γ phase is the most sensitive to ion implantation, while the sensitivity of the β phase is the lowest, and the sensitivity of the α_2 phase lies between the γ and β phases. The irradiation sensitivity of different phases is considered to be related to their crystal structures.

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