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# Helium irradiation behavior of tungsten-niobium alloys under different ion energies

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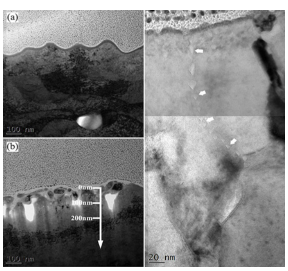
GRAPHICAL ABSTRACT



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ABSTRACT

Tungsten is a primary material in plasma-facing components (PFCs), but several disadvantages limit its application in fusion reactors. This work examined tungsten–niobium (W–Nb) alloys as potential PFC materials and investigated their response to low-energy helium ion irradiation. W–Nb powders were milled for 25 and 36 h, and the W–Nb samples were exposed to 50 and 80 eV helium ion irradiation with a flux of $1.5 \times 10^{22}$ ions/m$^2$ s at 1230 °C. The sample with a dense structure and obtained after a 25-h milling exhibited enhanced anti-irradiation performance. Sufficient energy can be provided to promote helium ion migration in the matrix, resulting in the formation of nanotendrils. Furthermore, in W–Nb alloys, the Nb-rich regions are more attractive to helium ions and thus provide protection to the tungsten matrix.

## 1. Introduction

With the progress of research on Tokmak and the development of divertors, the use of tungsten has become a research hotspot. Tungsten possesses high melting point, low hydrogen isotope retention, and high resistance to physical sputtering [1–3]. Tungsten-based plasma-facing components (PFMs) are not only affected by heat load but also by a large number of particles. Tungsten undergoes severe surface morphologies when subjected to low-energy helium and deuterium irradiation. At present, numerous studies on bubbles, blisters, and nanometric filamentary structures (fuzzes) are being conducted [4–8]. Upon exposure of tungsten to helium, some helium atoms become trapped in vacancy impurities, while other helium atoms cause surface modifications, forming helium bubbles, fuzzes, and even holes. The present study shows that nanofibers form on tungsten surfaces that are exposed to high fluxes of low-energy (20–60 eV) helium ions at

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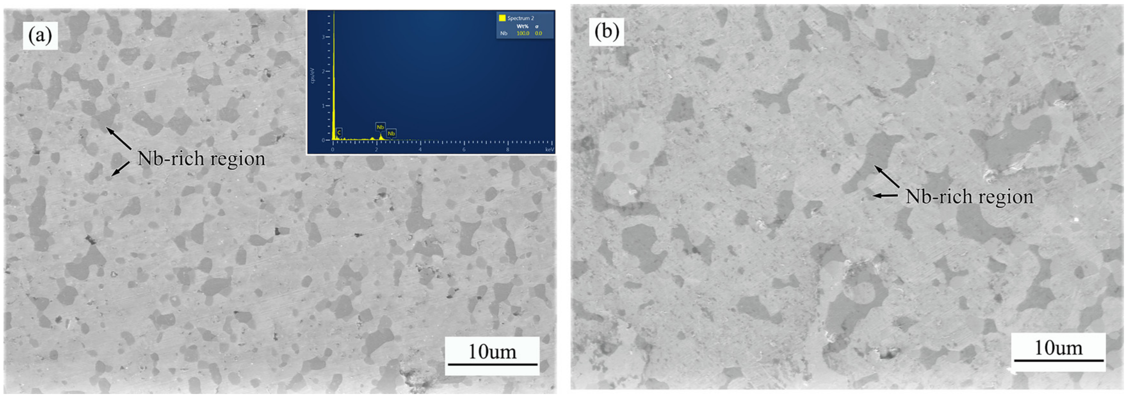


**Fig. 1.** SEM images of non-irradiated W-Nb alloys with different milling time: (a) 25 h and (b) 36 h.

temperature exceeding 900 K [9,10]. These investigations on the fundamental mechanisms and the physics behind this extreme morphological evolution are vital in the development of PFCs that are suitable for harsh fusion environments.

High ductile-to-brittle transition temperature, low recrystallization temperature, irradiation hardening, and brittleness limit the application of pure tungsten in fusion devices. As a result, tungsten alloys have been investigated as potential PFC materials. Tungsten alloys containing materials, such as rhenium and tantalum, demonstrate high ductility [11,12]. Other alloys, such as oxide dispersion-strengthened heavy alloys, demonstrate enhanced mechanical properties, decreased hardening, and embrittlement when exposed to high-energy neutrons [13–15]. Among the alloying elements, niobium presents with high melting point, corrosion resistance, and low thermal neutron absorption cross-section. In addition, niobium has high affinity for impurities (such as O, C, and N) and can bind with these impurity elements to form compounds with high melting temperatures, which can reduce total retention $D_2$ and clean the grain boundary of W alloys. A small amount of niobium can improve the ductility of tungsten-based materials and alter their behaviors under irradiation [16].

In a previous experiment [17], the effects of different milling times on the microstructure and radiation resistance of tungsten-niobium (W–Nb) alloys were analyzed based on surface morphology. In the present study, the two samples that showed the best and worst performances in the previous experiment were selected for further investigation of their damage behavior after irradiation.

## 2. Experiment

Powders containing 75 wt.% pure W (purity: > 99.9%, 1.0–1.3 μm) and 15 wt.% pure Nb (purity: > 99.9%, 40–50 μm) were manufactured through ball milling. Mechanical milling was performed at a rotation speed of 400 r/min and ball-to-powder ratio of 20:1 at room temperature. The jar and ball were made from tungsten carbide (WC). The milling medium used was absolute ethyl alcohol, and the milling times were 25 and 36 h. The prepared powders were consolidated through spark plasma sintering at 1700 °C for 5 min. The powder was loaded in a graphite die with a diameter of 20 mm. A DC current was then applied to the pulse, and a high uniaxial mechanical pressure was achieved. The entire sintering process was performed in a vacuum. The W–Nb specimens were cut into smaller pieces (10 mm × 10 mm × 1 mm) and mechanically polished using a diamond polishing paste. The two samples with the best and worst performance at different ball milling times in a previous work [17] were selected as objects of the present research. Samples obtained after 25 and 36 h milling were denoted as W25 and W36, respectively.

Helium exposure was performed on a large-power material irradiation experiment system, which uses a deceleration module to produce high-flux helium ion beams with 50 and 80 eV. The peak flux was $1.5 \times 10^{22}$ ions/m$^2$s, which lasted for 11 min. Exposures were conducted at 1230 °C to 1280 °C to a fluence of $9.90 \times 10^{24}$ ions/m$^2$, yielding an overall matrix under three conditions: W25 at 50 eV, W36 at 50 eV, and W25 at 80 eV. The base pressure during the irradiation was $1.0 \times 10^{-4}$ Pa. The directly irradiated region was a circle with 7 mm diameter. Scanning electron microscopy (SEM), focused ion beam (FIB), and transmission electron microscopy (TEM) were all conducted near the geometric center of the beam spot.

Changes in surface microstructure were observed using an SU8020 field-emission scanning electron microscope. FIB was used to prepare the samples for TEM. The FIB used was an FEI Helios NanoLab 650 DualBeam™ FIB-SEM. TEM was performed using a JEM-2100F transmission electron microscope.

## 3. Results

### 3.1. Effect of microstructure

Fig. 1 shows the morphology of the W-Nb samples prepared through (a) 25- and (b) 36-h ball milling. The inset (Fig. 1(a)) is the EDS pattern, which shows that the particles dispersed at the tungsten grain boundaries and within the grains (marked by an arrow in Fig. 1(a)) contain a Nb-rich phase. As shown in Fig. 1, the Nb-rich phase particles were more homogenously distributed in the tungsten grains and boundaries in W25 than in W36. Fig. 2 shows the XRD pattern of the non-irradiated W-15%Nb alloys at different milling times. $Nb_4C_3$ was evident in the XRD pattern, and it was formed due to the transformation of WC phase, which was introduced during ball milling and by carbon diffusion during SPS. Compared with the diffraction peak of standard tungsten,

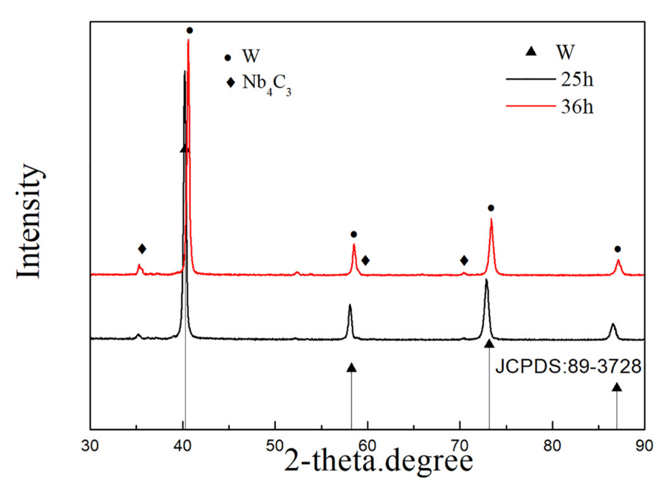


**Fig. 2.** The XRD pattern of non-irradiation W-15%Nb alloys at different milling times: (a) 25 h and (b) 36 h.