***In-situ TEM/heavy ion irradiation on ultrafine-and nanocrystalline-grained tungsten:
effect of 3 MeV Si, Cu and W ions.***

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Abstract

Plasma facing components for future fusion applications will experience helium- and neutron-induced structural damage. Direct observation of the in-situ dynamic response of such components during particle beam exposure assists in fundamental understanding of the physical phenomena that give rise to their irradiation resistance. We investigated the response of ultrafine and nanocrystalline-grained tungsten to 3 MeV heavy ion irradiations (Si^{2+} , Cu^{3+} and W^{4+}) for the simulation of neutron-induced damage through transmutation reactions via in-situ ion irradiation-transmission electron microscopy experiments. Defect densities as a function of irradiation dose (displacement per atom) and fluence were studied. Four stages of defect densities evolution were observed, as a function of irradiation dose: 1) increase in defect density at lower doses 2) higher defect production rate at the intermediate doses (before saturation), 3) reaching the maximum value, and 4) drop of the defect density in the case of W^{4+} , possibly due to defect coalescence and grain boundary absorption of small defect clusters. The effect of grain size on defect densities were investigated and found that defect densities were

independent of grain size in the ultrafine and nanocrystalline region (60- 400nm). These results were compared to other heavy ion irradiation studies of structural materials.

1. Introduction

There has been significant interest over the past several years for increasing radiation resistance of certain materials, specifically in structural materials for the use in the extreme radiation environment of nuclear reactors [1, 2]. Currently, tungsten is being considered as one of the best candidate materials for use in nuclear fusion reactors due to its mechanical strength, high melting point, good thermal conductivity, high sputter threshold, and low sputtering yield [3-5]. However, exposing tungsten materials to extreme irradiation environment, certain atomistic changes are known to develop such as the formation of defect clusters, voids, and bubbles [6, 7]. These atomistic changes will then manifest into microstructural changes [8-11], that can compromise the structural integrity of the material [12, 13]. To mitigate this severe drawback, innovative solutions are proposed [14-16], and increasing grain boundary density is conjectured to achieve higher irradiation tolerance [17]. The grain boundaries of a material are known to serve as sinks for both self-interstitial atoms as well as helium [18-21]. It was also demonstrated through molecular dynamic simulations that grain boundaries assist the process of interstitial-vacancy annihilation by re-emitting previously loaded interstitials into the bulk of the material effectively annihilating nearby vacancies and healing the damaged bulk material [22]. Several other studies demonstrated higher irradiation tolerance of nanocrystalline materials [23-25]. Accordingly, ultrafine and nanocrystalline tungsten samples should exhibit a higher radiation tolerance than commercially available

tungsten due to their increased density of grain boundaries. Thus, ultrafine and nanocrystalline grained tungsten should prove to be effective options for use as plasma facing components (PFCs). In this context, there are recent irradiation studies [26, 27] of ultrafine-grained (100 nm to 500 nm grains) and nanocrystalline-grained (less than 100 nm grains) tungsten, which is also shown to have improved mechanical properties than commercial coarse-grained counterpart [28].

In addition to helium-induced morphology change (irradiation damage) in tungsten, neutron irradiation (thermal neutrons in this case) can exacerbate the damage to the material bulk through defect generation via displacement cascades and transmutation reactions [29]. Numerous studies are available in literature regarding helium-induced [13, 30, 31] and neutron-induced [32, 33] irradiation damage on tungsten. However, fundamental understanding of atomistic processes, and the dynamic behavior of defects, that give rise to irradiation-induced damage can be further expanded through real time studies [26, 29, 34-36], where irradiation and material response are observed *in situ* (in real time).

The present study observes microstructural effects and displacement damage aspects due to room temperature, high-energy (3 MeV) heavy ion bombardment of nanocrystalline and ultrafine-grained samples for three ion species: silicon, copper, and tungsten. Heavy ion bombardment is used to provide a first-order imitation of atomic displacements and damage cascades formed due to neutron damage. Since in the case of neutron bombardment all of the neutron's energy is transferred to atoms through an absorption process, heavy ion bombardment should generate similar material damage to what neutron bombardment would generate, with the exception of transmutation

processes (transmutation reaction). The development of microstructure changes were monitored *in situ* with the incorporation of transmission electron microscopy (TEM). *In situ* TEM is one of the most promising techniques for observation defect formation and dynamics, as this allows for real-time analysis of microstructural changes due to radiation. By real-time and precise evaluation of defect development and microstructures during radiation, a better understanding of defect dynamics and microstructural evolution can be directly correlated to the fundamental driving mechanisms for these changes. Moreover, observing the response of various grain sizes under the same irradiation conditions can assist in predicting how different grained tungsten materials will ultimately behave in high-energy radiation environments.

2. Methods

Tungsten samples were produced by a surface plastic deformation technique called extrusion machining, which refines the microstructure of the workpiece subsurface and the resulting chips into a mixture of nanocrystalline and ultrafine grains by imposing large strains during the plastic deformation. Details about the starting material, processing conditions and final microstructures were previously reported [37].

For transmission electron microscopy sample preparation, tungsten chips having ultrafine and nanocrystalline grains were mechanically polished to 100 μm thickness. The final sample preparation step was performed via electropolishing (jet polishing) the sample with 0.5% aqueous sodium hydroxide solution at room temperature. It should be noted that there was no variation of mass thickness contrast other than what is expected from a slightly wedged sample (due to electropolishing), suggesting no preferential

etching of grains due to orientation or grain size dependence. Therefore, quantification of defects was not biased by the proximity of the surfaces especially for adjacent grains.

Irradiation experiments were performed at Sandia National Lab in the in-situ irradiation/ transmission electron microscopy facility which includes a JEOL 2100 TEM (200 KV), a 6 MV Tandem Accelerator, and a 10 kV Colutron [38]. Irradiations were performed at room temperature with 3 MeV Si, Cu, and W ions and the fluxes are presented in Table 1. During the experiment, the sample is tilted nominally 15 degrees with respect to the normal of the electron gun. The angle between the accelerator beam and the sample was around 75 degrees. Electron microscopy was performed utilizing bright field mode and select area diffraction.

Post-irradiation imaging were performed in bright field mode using FEI Titan 80/300 field emission TEM, operating at 300 kV, equipped with a Tridiem GIF. Defect density and defect area calculations on specific grains were performed using transmission electron microscopy micrographs taken under multi beam conditions as described by Jenkins [39]. In order to calculate the defect density, two different individuals independently counted the number of defects within the grain or part of it and then divided by the area, assuming that the defect density was uniform within the grain. The reported defect density is an average of two numbers with the difference between them is within the error calculated with means of Poisson statistics. The defect cluster area calculation was determined using Image J software.

3. Results and Discussion

3.1. Microstructure type

The ultrafine-and nanocrystalline-grained tungsten samples used in this study are made of elongated grains. The reported grain sizes are determined by the smallest distance between grain boundaries for a given grain (grain width). Therefore, a defect in a 100 nm grain will be within 50 nm from a grain boundary. Around 40% of the grain boundaries in the samples are high-angle type, as adjusted by the processing conditions used in the sample preparation [37]. High angle grain boundaries were previously postulated as high defect cluster sinks [36].

3.2. Interaction range and defect generation

Irradiation of a material with heavy ion irradiation causes displacement damage from which both interstitials and vacancies are created. While interstitials migrate at RT in tungsten, monovacancies and cluster vacancies need higher temperatures to migrate [40, 41]. Coalescence of interstitial defects is then possible at RT, leading to defect clusters. To study the effect of MeV metal irradiation and the subsequent defect cluster formation on ultrafine- and nanocrystalline-grained tungsten, TEM samples of the microstructure described above were irradiated in-situ at RT with three different species: silicon ions, copper ions, and tungsten ions of the same energy (3MeV) and nominal incident angle of 75 degrees. The thickness of the electron transparent region investigated did not exceed 200 nm. The average damage per incident ion per angstrom within the sample thickness (200 nm) and the penetration depth in the bulk sample for all three species were calculated using TRIM simulation software [42]. Table 1 shows the experimental parameters and the calculated data for each of the ion species. Figure 1 shows the profiles for damage production within the sample thickness of the TEM

samples based on the data simulated in TRIM [43]. The charge state of the metal ions were chosen to optimize the ion current. The ion fluxes were similar and low (Table 1); therefore, no overlapping cascades can occur within a time frame of damage production or initial relaxation mechanisms. The nominal doses, displacement per atom (dpa), for the experiments were calculated based on the measured fluences and simulated values of the average damage production using the relations published elsewhere [44]. As demonstrated in Figure 1, the calculated range for silicon and copper ions in the bulk tungsten sample is larger than the maximum sample thickness, thus both species pass through the tungsten sample, depositing just a part of their energy within the specimen. No vacancy production gradient along the sample thickness is expected in the case of copper and silicon ions (Figure 1). On the other hand, for the sample irradiated with the tungsten ions, the penetration depth is shorter than the sample thickness, and the ions deposit their entire energy inside the material. Therefore, one could expect to have the highest defect production rate and final defect density for the tungsten irradiation specimen. The tungsten ions, however, produces a vacancy concentration gradient through the sample with the maximum vacancy production occurring at 10 nm sample thickness (Figure 1).

Table 1. Experimental parameters and TRIM simulated data for in-situ irradiation and TEM imaging.

Ion species	Si ²⁺	Cu ³⁺	W ⁴⁺
Ion flux, ions/cm ² sec	3.4x10 ¹¹	1.4x10 ¹¹	1.8x10 ¹¹

Final fluence, ions/cm ²	6.2×10^{14}	5.8×10^{14}	3.3×10^{14}
Final dose, dpa	1.03	3.79	5.72
Ion range in bulk, nm	330	301	86
Average damage production within the TEM sample thickness (200nm), vacancies/ion A	1.05	4.15	11

Figure 1. Calculated vacancy production per ion within the sample thickness for three irradiating species.

3.3. Defect density calculations

The defect cluster density was calculated from the TEM micrographs taken in-situ from the same grain during the entire irradiation period for each specimen. Figures 2, 3, and 4 show TEM micrographs for the three tungsten samples before irradiation and at three different doses. The quantification of the defects for the three irradiation cases (Si^{2+} , Cu^{3+} and W^{4+}) was performed in similar types of elongated nanocrystalline and ultrafine grains (marked in the figures) .

Figure 2. TEM micrographs of similar adjacent regions in the sample irradiated with Si^{2+} at: a) zero dose; b) 0.13 dpa; c) 0.67 dpa; d) 1.03 dpa

Figure 3. TEM micrographs of similar adjacent regions in the sample irradiated with 3 MeV Cu³⁺ at: a) zero dose; b) 0.38 dpa; c) 2.14 dpa; d) 3.79 dpa

Figure 4. TEM micrographs of the same region in a sample irradiated with 3 MeV W^{4+} at: a) zero dose; b) 0.72 dpa; c) 2.9 dpa; d) 5.72 dpa

Generally, the TEM image characteristics highly depend on the imaging conditions chosen. In a given TEM micrograph, portion of the defect clusters are visible. Some will show weak or no contrast (g.b = 0 condition, where g is the diffraction vector and b is the Burgers vector). To minimize this quantification error in defect cluster density, defect-imaging conditions were performed using multi beam electron imaging and therefore, high proportions of resolvable loops can be obtained. Other factors decreasing defect density such as image forces [45], where glissile prismatic loops glide to the surface (observed in heavy ion irradiation of bcc materials) [46], can still exist. The

above factors are well summarized by Jenkins [39]. Foil thickness can also affect the defect density, and was studied by Li et al. [47]. The factors listed above will affect the absolute value of the calculated defect density in different tungsten samples. Therefore, in the following discussion, we will compare trends and rates at which the defect density changes in each specimen, but not the calculated values of the defect density.

3.4. Defect density vs. dpa

Figure 5 shows how the defect density in the tungsten sample changes with the irradiation fluence and doses for silicon, copper, and tungsten ions.

Figure 5. Defect density vs. fluence (left) and dose (right) for ultrafine-grained tungsten: Irradiated with Si^{2+} at flux 3.4×10^{11} (ions/cm²s), with Cu^{3+} at flux 1.4×10^{11} (ions/cm²s); and with W^{4+} at flux 1.8×10^{11} (ions/cm²s). The lines are a guide for the eye only. Grains used for defect density calculations are marked in figures 2, 3 and 4

Despite the higher damage production rate (Table 1) of Cu^{3+} , the higher defect density in the Si^{2+} irradiated sample compared to the Cu^{3+} at low doses may be related to the effect of the grain size and/or crystallographic orientation of the grain, as well as to the characteristics of the TEM images discussed above. However, the defect multiplication (increase) rate in the case of Si^{2+} irradiated sample is slower comparing to the two other samples: from 8×10^{13} ions/cm² (0.134 dpa) to 6×10^{14} ions/cm² (1.03 dpa),

the defect density has increased just about twice. Defect density for Cu^{3+} irradiated samples increased 10 times for the range from 6×10^{13} ions/cm² (0.380 dpa) to 6×10^{14} ions/cm² (3.79 dpa). At the same time, the defect density of W^{4+} irradiated samples reached the maximum at the fluence around 1.7×10^{14} ions/cm² (2.9 dpa) and dropped 1.6 times when reached the final fluence of 3×10^{14} ions/cm² (5.72 dpa). Since the defect production rate (Table 1) is different among the three irradiation cases, it is expected to observe a similar trend in the defect multiplication (increase) rate (Figure 5) and the defect production rate ($\text{W}^{4+} > \text{Cu}^{3+} > \text{Si}^{2+}$) for the same irradiation time (fluence), in agreement with the results. It should be noted, however, that a similar trend was also observed for the same irradiation dose (dpa). Since for a same dpa value, a higher defect production rate requires shorter time, this observation is attributed to effect of time in the response of point defects (decreases with defect production rate) and their diffusion to nearby grain boundaries. For example, for the same dpa, a defect generated in the W^{4+} case has 11 times less response time than in the Si^{2+} case.

The change of the defect density with irradiation doses in the tungsten samples for all three species follows the same trend: 1) slowly increasing (more evident in the case of Si and Cu irradiations) at the beginning because of relatively high density of the defect's sinks (grain boundaries), 2) higher defect production rate at the intermediate doses when sinks begin to saturate, 3) reaching the maximum value, and 4) drop of the defect density in the case of tungsten, possibly due to synergetic effect of the point defects coalescence. Migration of defects clusters to the grain boundaries and to the surface may also contribute to the decrease of the defect density at the high doses.

3.5. Comparison with previous studies

Some of the recent research shows similar results for other materials irradiated with heavy ions. Sun et al. [36] have published results of in-situ TEM study Kr^+ irradiated nano-crystalline and coarse-crystalline nickel, where they observed an increase of the dislocation loop density as the dose increased, followed by saturation of the defect density at high doses due to absorption of the defects by the sinks. Brimbal et al. [29] have reported the defect density evolution in iron specimen under 1 MeV Fe^+ irradiation. Defect density evolution, in this case, followed the same stages as discussed here. The drop in dislocation loop density was attributed partially to loop coalescence and primarily to the motion of the loops to the surface and their subsequent elimination. Both mechanisms are likely to play a role in the case of W^{4+} irradiation of the tungsten sample. A similar trend was also reported by He et al. [48] for UO_2 single crystal sample irradiated with 150 keV and 1 MeV Kr^+ . The final drop in the dislocation loop density, in this case, was associated with the growth of the dislocation loops, their coalescence and the formation of large clusters.

Because silicon ions have the largest penetration depth and the lowest damage production per ion within the sample thickness, the fluences we observed in the experiment for the sample irradiated with Si^{2+} correspond to lower doses (dpa) compared to the other specimens and demonstrate the early stage of defects formation in the tungsten sample under irradiation. In the case for the copper ions, the penetration depth is shorter, more interactions happen within the sample; albeit energy is partially deposited within the specimen. Thus, we have observed early and intermediate stages before the beginning of the defect density saturation. In contrast, the tungsten ions bombarding

tungsten sample have shorter range than the sample thickness; ions deposit their entire energy within the sample, producing large number of defects even at the low fluences. All these permitted four stages of defect density evolution as detailed below.

3.6. Defect density decrease at high doses (W^{4+} irradiation case)

There are two possible explanations for the decrease of the defect density - the fourth stage - in the sample irradiated with W^{4+} at the final doses, namely, the point defects coalescence and/or migration of defects clusters to grain boundaries and sample surface. Examples of point defect coalescence and defect migration to grain boundaries are shown in Figures 6 and 7 (video is available online in supplementary material). It should be noted that defects cluster migration between two pinning defects was observed at even less than 300 keV electron beam during post irradiation TEM imaging. This suggests a high defect population, which would increase defects cluster mobility. Figure 8 shows TEM micrograph of the grain and its boundary, taken after irradiation (5.79 dpa), where it is clearly seen that the defect concentration near the boundary is higher than the rest of the grain, suggesting that there is preferential diffusion of the defects to the grain boundaries (defects sinks). Similar results were observed in-situ by Sun et al. [36], stating grain boundaries play the role of main defect sinks in the fine-grained materials under ion irradiation at room temperature.

Figure 6. TEM micrographs demonstrating (a)-(b): absorption of a defect cluster to a grain boundary for the 3 MeV W^{4+} irradiation case at around $2 \times 10^{14} \text{ cm}^{-2}$. Red circles guide the eyes to areas where these phenomena occur.

Figure 7. TEM micrographs demonstrating (a)-(b): the coalescence of two defect clusters for the W^{4+} irradiation case at around $2 \times 10^{14} \text{ cm}^{-2}$. Red circles guide the eyes to areas where these phenomena occur.

Figure 8. TEM image of the grain boundary decorated with the defects in the sample irradiated with W^{4+}

3.7. Cluster planer area

The evolution of defect cluster planer area with dose for all three species is shown in Figure 9 a. Large size (up to 200 nm^2) defect clusters were observed in the sample irradiated with Cu^{3+} possibly due to the larger grain size (250 nm), compared to the grain sizes in the samples irradiated with Si^{2+} and W^{4+} (170 nm). Single point defects and small defect clusters in that case, were more likely to coalesce than to migrate and be absorbed by the grain boundaries. Damage production rate is smallest for the case of Si^{2+} irradiation. The radiation dose is low even at the highest fluence (Figure 9 b). This allowed enough time for the growth of defect clusters due to migration and coalescence of single defects and small clusters. In general, the dose dependent evolution of the defect cluster area was following the trend published elsewhere for other materials irradiated with heavy ions [29,36,48]

Figure 9. Average size of the defect clusters for three irradiated samples a) vs. dose b) vs. fluence. The lines are a guide for the eye only

3.8. Effect of grain size

In order to study the effect of the grain size on the defect cluster density, we have investigated ex-situ TEM micrographs of different regions of the three samples irradiated to the final doses. Grain size change did not change with irradiation dose, which is expected due to the low temperature (RT) used in this study. Figure 10 shows the defect density with grain size for the sample irradiated with Cu²⁺ to the final dose of 3.79 dpa.

Figure 10. The defect density dependence on the grain size for the sample irradiated with Cu^{3+} at 3.79 dpa

This particular sample also contained fine grains (larger than 500 nm) which enabled the defect density comparison with the ultrafine and nanocrystalline grains. This sample at the final dose corresponds to the intermediate stage of the tungsten irradiation case, before defect density saturation. We observed an almost exponential decay of the defect density from 8×10^{-3} to 5.5×10^{-5} defects/ nm^2 as the grain size increased from 200 nm (ultrafine grains) to 2768 nm (fine grains). However, if we focus on the ultrafine region of less than 400 nm grain width, the defect density varies between 4×10^{-3} to 8×10^{-3} defects/ nm^2 with no particular trend in the data distribution. Similar results were observed in the case of Si^{2+} and W^{4+} irradiation within the same grain width regime.

In order to eliminate the possible effect of the sample thickness' fluctuation as well as the effect of the resolution limit at the different magnifications for the grains of the different sizes, we compared only those grains from the same part of the sample at the final dose. Figure 11 shows the TEM images of the sample region with the grains of the different size for the specimens.

Figure 11. a) Sample irradiated with Si^{2+} : 1) 140 nm – 7.2×10^{-3} defects/ nm^2 , 2) 122 nm – 6.9×10^{-3} defects/ nm^2 , 3) 63 nm – 4.7×10^{-3} defects/ nm^2 , 4) 367 nm – 6.4×10^{-3} defects/ nm^2 ; b) Sample irradiated with Cu^{3+} : 1) 228 nm – 4.3×10^{-3} defects/ nm^2 ; 2) 202 nm – 5.9×10^{-3} defects/ nm^2 ; c) Sample irradiated with Cu^{3+} : 1) 228 nm – 4.3×10^{-3} defects/ nm^2 ; 2) 202 nm – 5.9×10^{-3} defects/ nm^2 .

defects/nm²; 3) 137 nm – 6.1×10^{-3} defects/nm²; c) Sample irradiated with W⁴⁺: 1) 372 nm – 2.3×10^{-3} defects/nm², 2) 128 nm – 4.5×10^{-3} defects/nm². The TEM micrographs were taken at the final dose.

The defect density calculated for adjacent grains for each sample seems to follow the same trend as the defect density of the grains of different grain widths over the entire specimen: random distribution for the sample irradiated with Si²⁺, decreasing defect density with increasing grain width for the sample irradiated with Cu³⁺, lower defect density for the large grain and higher defect density for the smaller grain in the sample irradiated with W⁴⁺.

Overall, no clear trend is observed for the grain width as a function of defect density for the ultrafine and nanocrystalline region (60 to 500 nm), while a decreasing trend is observed going from the ultrafine region (200 nm) to the fine region (few micrometers grain width). The defect density for each particular grain may also depend on its crystallographic orientation. Ion channeling [49], the phenomenon in which ions incident on or near a low-index zone axis penetrate anomalously deeply into a crystal by traveling down the channels between atomic columns, plays significant role in the case of ion bombardment of crystalline samples such as tungsten [50]. As an ion channels through the specimen, it produces much fewer defects than the non-channeling ion. The fraction of non-channeled ions depends on crystallographic orientation of the specimen. The relations between crystallographic orientation of the grains and the defect density for the given samples will be investigated in a future experiment. Moreover, the efficiency of grain boundary in absorbing defects depends on its type (small angle vs large angle) [36]. It should be noted that all experiments were performed at RT (possible increase in temperature can happen due to ion bombardment), and therefore vacancy migration did

not occur [41, 51, 52]. However, interstitial migration has a very low activation energy (0.054 eV) in tungsten and would occur even at room temperature [40]. Then, the interstitial-vacancy annihilation is slower and will be enhanced at high irradiation doses. In addition, loop migration and annihilation to the grain boundaries can occur through loop gliding or climb. For Frank type loops, glide is forbidden and the migration can occur through diffusion controlled process (climb) [53] where vacancy absorption and re-emission are important [36]. Such mechanism will be very slow at RT in tungsten due to vacancies being immobile. However, it was reported that the temperature in the divertor region in the International Tokamak Experimental Reactor (ITER) will be low (less than 200 °C) except for the strike point of the outer divertor [5]. Therefore, our experiments and results simulate neutron damage on the divertor region of ITER.

Formation of voids and/or any subsequent microstructural evolution on the sample surfaces were also investigated via out-of-focus Fresnel contrast imaging of the TEM samples and scanning electron microscopy (SEM) imaging of the samples' surfaces. No voids were observed, which is expected at the low temperature used in this study (RT) due to impossible vacancy coalescence. SEM of the sample surfaces revealed no change in the surface microstructure after irradiation.

Future work will focus on high temperature W irradiation and dual beam irradiations with low energy helium ions. The effect of grain boundaries can be better studied by high temperature experiments to facilitate loop glide and diffusion controlled climb to grain boundaries. Concurrent He implantation and W irradiation will elucidate any synergistic effects, possible grain growth, void formation and surface microstructural changes.

4. Conclusion

In conclusion, displacement damage experiments on ultrafine- and nanocrystalline-grained tungsten were performed via heavy ion irradiation with an in-situ TEM imaging at ambient temperature. The response of the samples can be summarized as follows:

- The dose dependent evolution of the defect density involves four stages: 1) the early stage corresponding to the low defect production rate, 2) intermediate stage with the higher defect production rate, 3) reaching the maximum value of the defect density and 4) for W, a decrease due to synergetic effect of the defect coalescence and defect absorption by the grain boundaries.
- The dose dependent evolution of the area of the defect clusters was following the trend reported in literature of heavy ion irradiation on nuclear materials with almost linear increase at lower doses and leveling at higher doses.
- We have observed no correlation between the grain width and the defect density for the ultrafine and nanocrystalline region. The defect density, however, demonstrated a decreasing trend (exponential decay) in the range of 200-2768 nm. No void formation, grain growth and surface microstructural changes were observed.

Shortly, observation of the dynamic response of ultrafine- and nanocrystalline-grained tungsten during heavy metal irradiation, summarized in this work, gives rise to fundamental understanding of the atomistic processes (defect nucleation, growth,

coalescence, etc.) that can take place in these materials when exposed to neutron irradiation damage in future fusion reactors.

Acknowledgements

O. El-Atwani would like to thank Prof. Jean Paul Allain for his help in funding him while performing part of the work. K. Hattar acknowledges the Division of Materials Science and Engineering, Office of Basic Energy Sciences, U.S. Department of Energy. Sandia National Laboratories is a multi-program laboratory managed and operated by Sandia Corporation, a wholly owned subsidiary of Lockheed Martin Corporation, for the U.S. Department of Energy's National Nuclear Security Administration under contract DE-AC04-94AL85000

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Figure 1

Figure 2

Figure 3

Figure 4

Figure 5

Figure 6

$t = 0$ seconds

$t = 0.5$ seconds

Figure 7

$t = 0$ seconds

$t = 0.5$ seconds

Figure 8

Figure 9

Figure 10

Figure 11