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Motion of Defect Clusters and Dislocations at a Crack Tip of Irradiated Material

Won-Jin Moon,a Sang Chul Kwon,b Whung Whoe Kim,b
a Gwang-ju Center, Korea Basic Science Institute, Yongbong-dong, Buk-gu, Gwangju
b NMTD, Korea Atomic Energy Research Institute, Duckjin-dong 150, Yusong-gu, Taejeon

1. Introduction

Effects of defect clusters on mechanical properties of irradiated materials have not been clarified until now. Two radiation hardening models have been proposed. One is a dispersed barrier hardening mechanism [1] based on the Orowan hardening model. This explains defect clusters as barriers to a dislocation motion. Generally the dislocation would rather shear or remove the defect clusters than make so-called Orowan loops. And the other is a cascade induced source hardening mechanism [2], which explains defect clusters as a Cottrell atmosphere for dislocation motions. However, the above mechanisms can not explain the microstructure of deformed material after irradiation and the phenomenon of yield softening. These mechanisms are based on an immobility of clusters. But we observed defect clusters could move into a specific crystallographic direction easily. Through 3 times of High Voltage Electron Microscope analysis, defect clusters have been observed to make one dimensional motion without applied external stress. If very small defect clusters could move under a stress gradient due to interactions between clusters, we can suggest that the clusters will move more actively when a stress gradient is applied externally. In-situ tensile test at TEM, we confirmed that kind of motion. We suggest defect clusters can move into crack tip, a stress-concentrated area due to tensile stress gradient and dislocations move out from the area by shear stress. Therefore radiation hardening can be explained agglomeration of defect clusters at stress concentrated area prohibits a generation of dislocation and make an increase of yield point.

2. Methods and Results

2.1 Experimental

The starting foil for this study was pure Fe (99.994%) with a thickness of 0.5 mm obtained from Johnson-Matthey. To relieve machining effect, the foil was annealed at 850 ℃ for 3 hrs in a vacuum condition. The After TEM disks were ground into 40 μ m in thickness, they were electro-polished into 20 μ m in thickness. In situ TEM straining specimens of dimensions 11.6 mm × 2.5 mm were punched with using a specially designed punch. The specimens were thinned to perforation at room temperature using a solution of 5% perchloric acid and 95% acetic acid, cooled to -40 ℃ with an applied potential of 16 V. Straining was performed with a straining stage after irradiation with 1.25 MeV electrons, at a rate of 4.7 x 10^{23} /m^2s for 5 minutes. The motion of the clusters was recorded through a CCD camera with a time interval of 1/30 s.

2.2 Defect Cluster Glide into Crack Tip

Before stress was applied, defect clusters were not formed at the crack tip as shown in Fig. 1. Because of the image force, the formation of clusters at the crack tip is difficult. When tensile stress was applied and increased slowly, the clusters came to the crack tip and agglomerated as shown in Fig. 2.

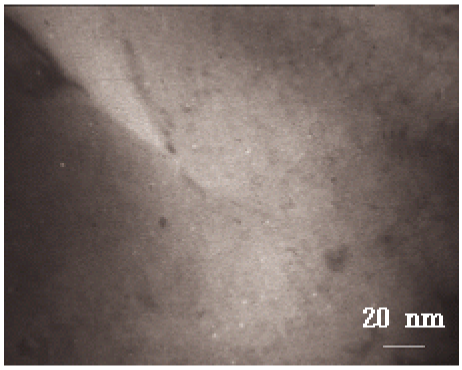

Fig. 1. Defect clusters near crack tip before application of stress.

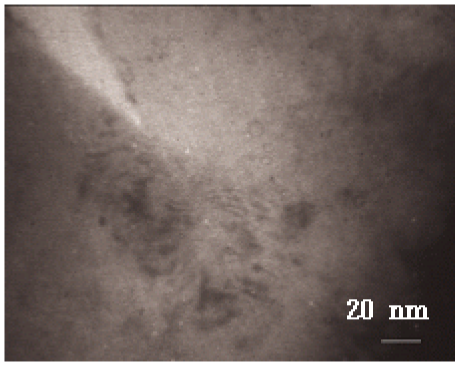

Fig. 2. Agglomeration of defect clusters after application of stress.

These phenomena show an evidence of drift motion due to tensile stress gradient. At the crack tip, we can assume the crack tip condition as Mode I crack, even though Mode I , II and III are combined practically. As shown in Fig. 3, a tensile stress gradient can be formed at the crack tip. When the applied stress is very low, the plastic zone size (r_p) will be very small, the effect of plastic zone can be disregarded.

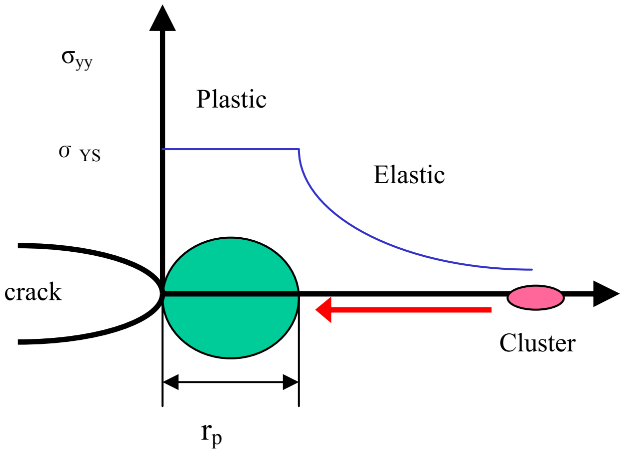

Fig. 3. Tensile stress distribution at stress intensified region of crack tip.

Before the plastic zone of the crack tip, the maximum tensile stress is yield stress. Therefore, defect clusters can move into the tip when applied stress is lower than the yield stress. Generally interstitial clusters are formed in Fe at room temperature. To decrease the compressive stress field near the clusters, the clusters will move into tensile field.

2.3 Motion of Dislocations at Crack Tip

When tensile stress is applied to specimens at mode I crack tip, dislocations will generate at crack tip. At unirradiated materials, the propagation direction of dislocation will be set at an angle where the Schmid factor is the largest as shown in Fig. 4.

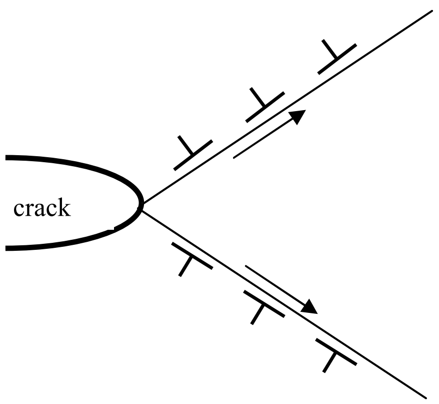

Fig. 4. Dislocations are emitted on the inclined sip planes under the tensile loading condition.

When the resolved shear stress at the slip plane reaches a critical value, the dislocations propagate from crack tip into the inner side of matrix. Ohr[3]observed that dislocation-free-zone was formed between the crack tip and the plastic zone. Therefore, just before the crack tip, the generation and propagation of dislocation cannot be observed.

2.4 Radiation Hardening

At a stress concentrated area of irradiated iron, interstitial defect clusters will agglomerate if tensile stress is applied. Due to the tensile loading, a shear stress will be formed at the crack tip. When the shear stress reaches a critical value with which the dislocation overcomes the barrier of clusters, the dislocations will be generated at the crack tip. So, the yield strength will increase. When the dislocations break away the barrier, they can move easily along the clear channel which has been formed already. Therefore, the yield softening can be explained. When the quantity of radiation damage is small, the phenomenon of a simple hardening can be explained with the combination of normal deformation and the radiation hardening.

3. Conclusion

The motion of defect clusters was observed when in-situ TEM straining was applied to an iron sample irradiated with electrons. The clusters could move without any interaction with dislocations. The direction of motion was into the crack tip, which was different from the direction of dislocation. It is considered the reason of the difference exists at stress type which induces the motion. Defect clusters move along the tensile stress gradient. On the other hand, the dislocations at the crack tip can move on the slip plane by the shear stress.

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