EVOLUTION OF CLEARED CHANNELS IN NEUTRON-IRRADIATED PURE COPPER AS A FUNCTION OF TENSILE STRAIN—D. J. Edwards (Pacific Northwest National Laboratory)* and B. N. Singh (Risø National Laboratory, Denmark)

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EXTENDED ABSTRACT

The occurrence of localized deformation via dislocation channeling in irradiated materials continues to present an interesting challenge to the fusion materials community. While this phenomenon has been known for over 40 years [1-4], until recently little progress has been made in understanding the origin of the channels and the variables that influence their formation and propagation. More recent work [5-10] has led to a re-evaluation of how channels are formed and propagate in irradiated materials, as well as in unirradiated materials. Some of the outstanding issues that remain unanswered include how to precisely relate the occurrence of the yield point to the microstructure formed during irradiation, how channels are cleared of defects and leave no dislocation debris behind, and how to account for the increase in stress when no conventional work hardening processes are evident.

As part of a broader set of experiments aimed at studying localized deformation in pure metals, a simple experiment was conducted wherein tensile tests of neutron-irradiated copper were stopped at intermediate strains. The material used in the present investigation was thin (0.3 mm) sheet of oxygen-free high conductivity (OFHC) copper containing 10, 3, < 1 and < 1 ppm, respectively, of Ag, Si, Fe and Mg, respectively. The oxygen content of this copper was found to be 34 appm. Five tensile samples of OFHC copper were irradiated in the DR-3 reactor at Risø National Laboratory. Prior to irradiation, the OFHC copper samples were given a solution annealing treatment of 823K for 2 hours in a vacuum of 10^{-6} torr. The resulting grain size and dislocation density were about 30 μm and $\sim 10^{12}$ m^{-2}, respectively. The tensile specimens were irradiated at 323K to a dose level of 0.3 dpa (NRT). All specimens were irradiated at a displacement damage rate of $\sim 5 \times 10^{-8}$ dpa (NRT)/s. The specimens were tensile tested at room temperature in an Instron machine at a strain rate of 1.2×10^{-3} s^{-1}. One specimen was tested to failure, one specimen loaded to ~90% of the upper yield stress, unloaded and then removed from the test fixture, and the remaining 3 specimens strained plastically to 1.5, 5 or 14.5% elongation, unloaded and then removed from the tensile machine. The full tensile curve and the associated stress-strain curves for the other four samples are shown in Figure 1.

The sample stopped at ~90% of the yield stress exhibited a few narrow channels scattered at random in isolated grains. Defect free zones (DFZ) around grain boundaries were common, and were definitively not diffusion based because the width of the zones varied considerably along the length of an individual grain boundary and sometimes disappeared altogether. A check of the untested sample revealed no such features, indicating the defect zones were a consequence of grain boundary migration or sliding under the high stress reached before yielding. The main mode of plastic strain accumulation was via by dislocation channels initiated at interfacial stress concentrations that then propagated quickly through the grain. Examination of the samples strained to successively higher plastic strains confirmed that new channels continued to form even in the necked region of the failed tensile sample, illustrating that the dense population of defects effectively inhibits the global movement of dislocations to very high plastic strains and stresses. Examples of the progressive formation of slip steps on the surface of strained specimens are shown in Figure 2. These slip steps are thought to correspond directly to individual dislocation channels. The yield point is thought to be related to the initial generation of isolated channels from stress concentrators (not necessarily unpinning of Frank-Read sources), which is then followed by the continual generation of new channels from interfacial stress concentrators as the plastic strain increases to the point of necking and failure. Dislocations are not produced uniformly inside the grains by

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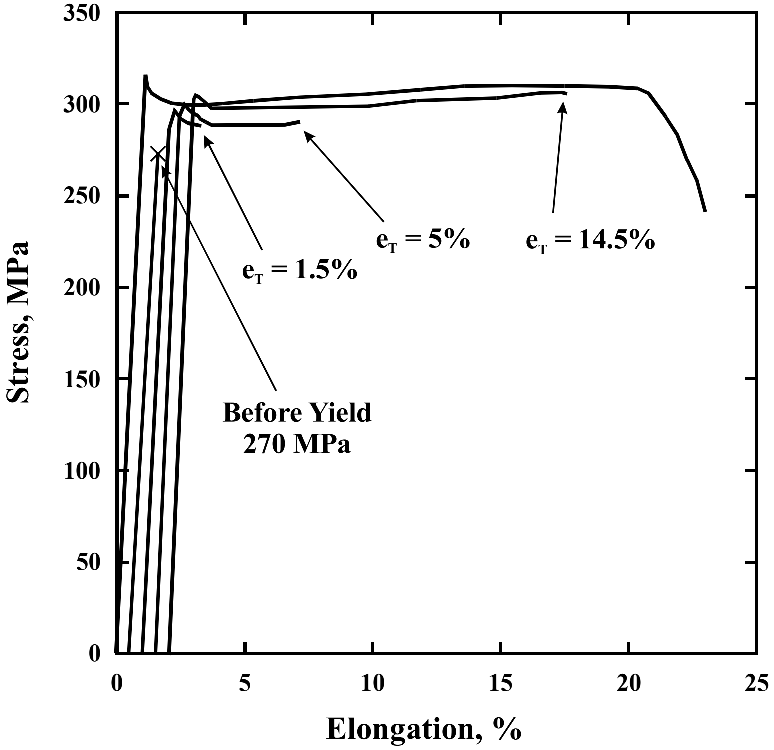

Figure 1. The tensile curves for each of the 5 tested conditions are shown. The stress or strain at which the tests were stopped for each specimen are indicated by the arrows. Note that some variability exists from specimen to specimen, so the yield point varies slightly. Curves are offset for clarity.

deformation, only in dislocation channels. Pre-existing loops and dislocation segments formed during irradiation are able to move and interact over very short distances, however, these dislocations do not appear to play any role in channel formation. The lack of any cell wall formation or general expansion of the DFZs around grain boundaries further testifies that dislocations produced outside of channels cannot move large distances in this irradiation condition. The continued generation of new channels past even the onset of necking subdivides the grains into smaller volumes that rotate and slide with respect to each other, much like subgrains in unirradiated and deformed copper.

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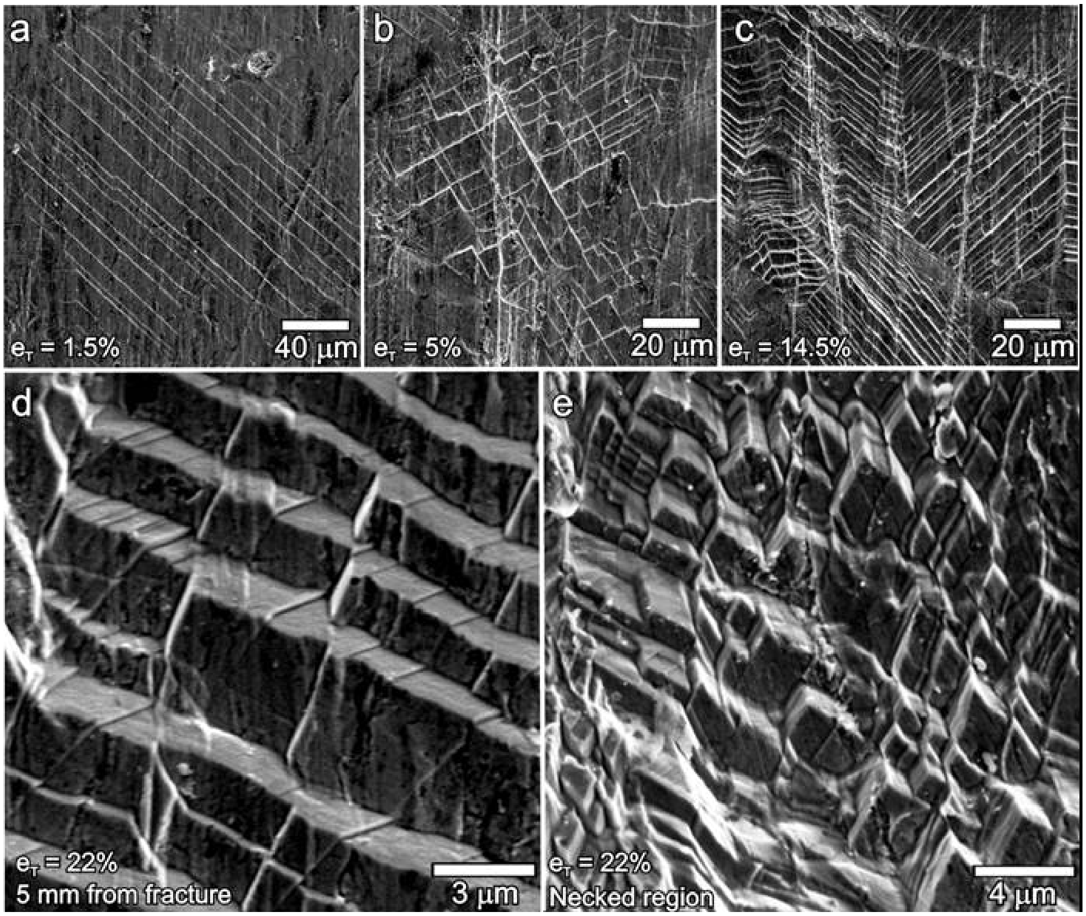

Figure 2. Examples of the slip step evolution as a function of strain are shown. As the plastic strain increases, the density of slip steps in a given grain increases and secondary slip steps begin to arise. In the last two figures (d and e), the crystal has been divided into discrete rectangular volumes that appear to slide with respect to each other. The image in (d) was taken 5 mm from the final fracture, and the image in (e) was taken in the necked region less than 1 mm from the final fracture location.

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