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~~APPENDIX IV~~

THE EFFECT OF NEUTRON DAMAGE ON THE DISLOCATION DYNAMICS
OF B.C.C. METALS AND SOLID SOLUTIONS AT LOW TEMPERATURES

R. J. Arsenault and E. Pink
Engineering Materials Group, and Dept.
of Chemical Engineering
University of Maryland
College Park, Maryland

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THE EFFECT OF NEUTRON DAMAGE ON THE DISLOCATION DYNAMICS
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R. J. Arsenault and E. Pink
Engineering Materials Group
University of Maryland, College Park, Md. 20740

Introduction

It has demonstrated in the case of f.c.c. metals that neutron damage has a large effect on the dislocation dynamics, that is plastic flow. The neutron damage results in an increase in the yield stress but more significantly it results in a large increase in the temperature dependency of the yield stress. (1) However, in the case of b.c.c. metals the evidence is not as convincing and predictions based on results obtained from neutron irradiated f.c.c. metals and computer calculations indicate that the change should be small. (2)

In considering the effect of neutron damage it is necessary to take into account the fact that the yield and flow stress have to be divided into two parts: the thermal component defined as the effective stress τ^* , and an athermal component τ_μ . The damage produced by neutron irradiation always causes an increase of τ_μ , and this increase is proportional to the interstitial impurity concentration and neutron dose. The effect of neutron damage on τ^* is not monotonic in all cases.

An investigation was undertaken to determine whether a change in dislocation dynamics due to neutron irradiation could be observed in b.c.c. metals. Single crystals and polycrystalline samples of V, V-Ti, and Mo with various interstitial impurity levels were irradiated to doses ranging from 10^{19} to 10^{20} n/cm² ($E > 1$ MeV) at 60-120°C. Most of the samples were tested in compression. The yield and flow stress were investigated

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at various temperatures and strain rates, and the activation parameters of slip were also established as a function of stress and strain.

Experimental Results and Discussions

Vanadium

In a discussion of the effect of neutron damage on the dislocation motion in V the results should be considered in terms of the amount of interstitial impurities in the samples. The presence of H_2 will be considered as a separate case. It has been demonstrated that for O_2 concentrations > 100 wt-ppm the O_2 atoms are the rate controlling barriers to dislocation motion. (3) For V containing a high level of O_2 (1800 wt-ppm) the neutron damage results in only an increase in τ_μ and no change in τ^* . At the higher levels of O_2 concentration the material has such a large O_2 concentration, and even though the irradiation damage has been completely saturated with oxygen, the random oxygen concentration does not decrease appreciably due to neutron irradiation. The rate controlling mechanism is still due to the O_2 atoms, and τ^* remains approximately the same. Fig. 1 is plot of the difference in τ^* between irradiated and annealed V for various O_2 concentrations. The irradiation produced defects which have been saturated with O_2 , result in long range barriers to dislocation motion and therefore the irradiation only causes an increase in the athermal component of the flow stress.

The O_2 interstitials are still the rate controlling obstacle

to dislocation motion in the range of 100 to 300 wt. ppm. An internal friction study did show that neutron irradiation does result in a reduction of the random O_2 concentration. (4)

Therefore, it is possible that neutron irradiation can cause a decrease in τ^* , as shown in Fig. 1. However, there is not a complete removal of the random oxygen atoms as internal friction results indicate, and is also suggested by the fact that H_O is still about 1 eV (Fig. 2). Therefore, they are still the short range barriers to dislocation motion in the irradiated intermediate O_2 samples.

In the case of the annealed low O_2 concentration material the rate controlling mechanism of slip is the double kink mechanism (3) and neutron irradiation results in an increase of τ^* as shown in Fig. 1 along with an increase in τ_μ . The neutron irradiation produced defects in this material are not saturated with O_2 and as a result the neutron produced defects act as short range barriers to dislocation motion. From a consideration of the activation parameters it is possible to determine that the spacing between the defects is small $\approx 10b$, the size of the defect is also small 1 to $2b$ in diameter, and the strain has a maximum value of 0.48. These dimensions suggest that the neutron produced defect is a di-vacancy.

The presence of a small amount of H_2 (11 wt-ppm) results in a discontinuity "a bump" in the yield stress vs. temperature curve at $193^\circ K$, as shown in Fig. 3. This increase in yield stress at $193^\circ K$ is due to stress induced ordering of the H_2 atoms by the strain fields of the moving dislocations. (5) Neutron irradiation of the H - V results in a decrease of the yield stress at $193^\circ K$ as shown in Fig. 3. It has been determined that this decrease is not connected with a decrease in τ_μ as the

result of deformation at 193°K. It appears that the decrease in the yield stress is due to a decrease in τ^* .

The neutron damage also influences the deformation characteristics in other ways. In b.c.c. metals it has been established that the deformation mechanism is the same for yielding and macro flow. (6) The activation volume (v) is therefore independent of strain. In neutron irradiated vanadium v is strain dependent, as is shown in Fig. 4. There is a slight increase of v with strain to ~ 12% strain, then there is an abrupt increase in v to values obtained from annealed vanadium. This affect is temperature independent and the amount of strain required to "remove" the characteristics of the irradiation is constant for all test temperatures. The stress-strain curve exhibits a higher yield point in the irradiated condition, but the level of stress returns to that of the annealed condition at the same strain, i.e. 12% as shown in Fig. 4.

Molybdenum

Due to the very low solubility of interstitial atoms in Mo, it may be considered as being free of interstitially dissolved impurity atoms. Therefore, impurity interstitials can not be the rate controlling barrier to dislocation motion. This also means that there can not be a reduction of τ^* as a result of neutron irradiation due to a reduction of the random interstitial concentration. However, neutron irradiation of Mo samples containing 10 wt ppm O_2 , 32 wt ppm C and 5 wt ppm N_2 resulted in a slight decrease of τ^* as shown in Fig. 5. It is difficult to arrive at an explanation for this decrease in τ^* . A possible explanation is a mechanism proposed by Weertman

(7), whereby the neutron produced defects can aid in the formation of a double kink. However, the decrease in τ^* due to neutron irradiation is not observed in Mo samples containing a smaller interstitial concentration as shown in Fig. 5.

Vanadium Titanium Alloys

From the investigation of unalloyed V it was shown that the presence of impurity interstitials resulted in a reduction or no change in τ^* due to neutron irradiation. Also, it has been convincingly shown that small additions of Ti "scavenge" the V matrix, i.e. the Ti removes the impurity interstitials from random solid solution. (4) Therefore, it should be expected that for low Ti content alloys there should be a large increase in τ^* due to neutron irradiation, but again a similar complication as with Mo is observed, i.e. there was a reduction of τ^* (Fig.6). The decrease in τ^* due to neutron irradiation is only confined to the low Ti content alloys, (0.75 and 1.6% Ti) whereas the higher Ti content alloys (4 and 20% Ti) there was no change in τ^* due to neutron irradiation. (8) Also there is no change in the activation parameters of the high Ti alloys due to the neutron irradiation. The lack of a change can be explained by at least two concepts. First, the neutron produced defects could be saturated with O_2 atoms due to the higher binding energy between the defect than between Ti. For it has been shown in the case of unalloyed V the presence of impurity interstitials reduces and eliminates any increase in τ^* due to neutron irradiation. The second concept is based on the assumption that if the neutron produced defect leads to an activation energy of deformation at zero τ^* of .65 eV. When this is compared to a value of 1.5 eV, it has been shown that the neutron produced defects will not have an effect on τ^* .

The decrease in τ^* of low Ti content alloys as a result of irradiation is not as readily explained. There are several possibilities, however, the most likely explanation is the same explanation as that used to explain the decrease observed in τ^* in Mo due to neutron irradiation. This explanation is based on the fact that it is possible for the neutron produced defect to aid in the function of a double kink, thereby reducing τ^* .

Conclusions

The effect of neutron damage on the thermal component of the flow stress (τ^*) of b.c.c. metals is strongly influenced by the amount of impurity interstitials in the metal. A general conclusion, which is independent of the specific b.c.c. metal, is that the presence of a high interstitial concentration results in no change in τ^* as result of neutron irradiation. In the intermediate interstitial range (100 - 300 wt ppm) there is a possibility that a reduction in τ^* can be observed do to trapping of interstitials by the neutron produced defects. However, for this to occur, i.e. a reduction in τ^* , the interstitials have to be rate controlling barrier to dislocation motion. Therefore, this can only occur in certain b.c.c. metals e.g. V but not in others e.g. Mo or Fe. In b.c.c. metals such as Mo where interstitials are not the rate controlling barriers the intermediate interstitial case is the same as the high interstitial case i.e. no change in τ^* due to neutron irradiation. If there is a low interstitial concentration then it is possible that the neutron damage will result in an increase in τ^* . However, this statement needs to be qualified, an increase will only be observed in those b.c.c. metals which have a low activation energy of deformation, i.e. a small Peierls' stress, e.g. Fe, Nb, V. In such b.c.c. metals as Mo where activation

energy of slip is high, neutron irradiation again will not result in an increase in τ^* at low temperatures.

These general observations become obscured by alloy additions, and reasons for this are readily understood.

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FIG.1: A plot of the difference in the effective stress between irradiated and annealed single crystal and polycrystalline vanadium. The comparison is made between samples of the same interstitial concentration.

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FIG. 2: A plot of activation energy vs. effective stress of vanadium single crystals of two different oxygen concentrations, and of neutron irradiated single crystals.

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ACTIVATION ENTHALPY H (eV)

FIG. 2: A plot of activation energy vs. effective stress of vanadium single crystals of two different oxygen concentrations, and of neutron irradiated single crystals of low and intermediate oxygen content.

FIG. 3: The yield stress vs. temperature of vanadium containing 11 wt-ppm H_2 .

FIG. 4: Stress-strain curves and activation volume vs. strain for annealed and irradiated

FIG. 4: Stress-strain curves and activation volume vs. strain for annealed and irradiated vanadium single crystals at room temperature.

SHEAR STRESS, τ_y (kg/mm²)

FIG. 5: The yield stress vs. temperature of annealed and neutron irradiated Mo single crystals

FIG. 6: The yield stress vs. temperature of annealed and neutron irradiated single and polycrystalline vanadium and V-Ti alloys.

6: The yield stress vs. temperature of annealed and neutron irradiated single and polycrystalline vanadium and V-Ti alloys.