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NEUTRON IRRADIATION EFFECTS ON THE MECHANICAL
PROPERTIES OF THORIUM AND THORIUM-CARBON ALLOY

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M. S. Thesis Submitted to Iowa State University

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Neutron irradiation effects on the mechanical
properties of thorium and thorium-carbon alloy

by

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ABSTRACT

The effects of neutron exposure to 3.0×10^{18} neutrons/cm² on the mechanical properties of thorium and thorium-carbon alloy are described. Tensile measurements were done at six different test temperatures from 4°K to 503°K and at two strain rates. Thorium and thorium-carbon alloy are shown to display typical radiation hardening like other face-centered cubic metals. The yield drop phenomenon of the thorium-carbon alloy is unchanged after irradiation. The variation of shear stress and effective shear stress with test temperature was fitted to Seeger's and Fleischer's equations for irradiated and unirradiated thorium and thorium-carbon alloy. Neutron irradiation apparently contributes an athermal component to the yield strength. However, some thermal component is detected in the low temperature range. Strain-rate parameter is increased and activation volume is decreased slightly for both kinds of metal after irradiation.

INTRODUCTION

The effects of neutron irradiation on the mechanical properties of metals were centered on the face-centered cubic (f.c.c.) metals in the earlier studies. The face-centered cubic metals like copper, nickel, gold, silver and aluminum were attainable in higher purities and their deformation processes were better understood. In the last few years, considerable experiments have been made and many different theoretical explanations have been built up. However, α -thorium (below transformation point 1345°C) is also face-centered cubic metal but the neutron irradiation effects on the mechanical properties are both fragmentary and conflicting. The primary purpose of this study is to obtain information concerning the effects of neutron irradiation upon the mechanical properties of pure thorium and thorium-carbon alloys and to compare the effects with those observed in other face-centered cubic metals.

Neutron Irradiation Effects on the
Mechanical Properties of Face-Centered Cubic Metal

The basic concepts of neutron irradiation effects on metals have been developed by many authors (1-6). Here only a brief review will be presented. When an energetic particle, say a fast neutron ($E > 1 \text{ MeV}$), traverses through matter, it interacts with matter by: (a) elastic collisions

with lattice atoms, and (b) electronic excitation and ionization of atoms through charge interaction. The first causes atoms to be displaced from their normal lattice positions resulting in vacancies, interstitials and replacements. The second is of little importance in metals, appearing only as heat. The vacancies and interstitials produced upon atom displacement may give rise to defect clusters, dislocation loops, voids, and bubbles, depending upon the neutron dose and irradiation temperature. All these kinds of defects can influence the mechanical properties of the metals.

There are a number of interesting changes in the mechanical properties of neutron irradiated metals and alloys, many of which may best be seen by observation of the stress-strain curves. Some of these properties are: yield stress (critical shear stress), work-hardening rate, ultimate tensile strength, uniform elongation prior to necking, and reduction in area at the fracture surface. Makin (7) observed the significant changes in the shear stress for single crystals of copper for low radiation doses (about 10^{16} neutrons/cm²). According to Greenfield and Wilsdorf (8), the critical shear stress of single crystals of copper increases eight times after irradiation to 10^{18} neutron/cm². Blewitt et al. (9) observed an increase in critical shear stress of copper single crystals

of a factor twenty-two after a room temperature irradiation up to 2×10^{19} neutrons/cm². The general nature of increase in yield stress of polycrystalline copper after irradiation is illustrated in Figure 1 after Makin (10). It can be seen that the nature of the yield is altered, the smooth yield of the unirradiated material being replaced by a small yield drop, and the yield stress is greatly increased by irradiation.

The rate of work-hardening of irradiated face-centered cubic metals such as copper and nickel is usually lower than for unirradiated metals, as is reported by Makin (11), and the increase in ultimate tensile strength due to irradiation is usually smaller than for the yield stress. The reduction of the work-hardening ability of irradiated samples is correlated with the development of necking and plastic instability at smaller strain than in unirradiated samples. This results in a reduction of uniform elongation despite the fact that the samples still fracture in the same manner as for the unirradiated metal. There is also a decrease in elongation to fracture upon irradiation. Makin (10) pointed out that the reduction in area at fracture is decreased by irradiation, but the decrease is very small in most face-centered cubic metals, indicating that even after 10^{22} neutrons/cm² dose there is no tendency towards brittle behavior in this crystal structure.

Figure 1. Stress-strain curves for polycrystalline copper tested at 20°C after various epithermal neutron doses at 80°C. Fast neutron doses are approximately 10% of the indicated values. After Makin (10)

The effects of tensile test temperature on irradiated single crystal and polycrystalline face-centered cubic metals have been studied. The critical shear stress of irradiated copper single crystals is very sensitive to the testing temperature, being much greater at low temperature than at high. A theoretical relationship between the critical shear stress, τ_c , and the absolute temperature, T , derived for a single type of dispersed obstacle is $\tau_c^{2/3} \propto T^{2/3}$, from Seeger (12). Makin (13) pointed out that upon extrapolation at the higher temperature the above relation is obeyed; for example, for copper crystals above $\sim 250^\circ\text{K}$. Below this temperature, the observed τ_c is greater than that predicted, and this has been attributed to the presence of a spectrum of obstacles of different strengths. Makin (11) studied the effects of neutron irradiation at 100°C with 5×10^{19} neutrons/cm² on the mechanical properties of annealed polycrystalline copper and nickel at a variety of testing temperatures between -195° and 200°C , as shown in Figure 2. The yield stress of both metals was increased by the irradiation and the increase was markedly temperature dependent, being very much greater at low temperature than at high. This temperature-dependent increase in yield stress was reversible and not due to thermal annealing. The temperature dependence of the yield strength of other irradiated face-centered cubic metals

Figure 2. Temperature dependence of the yield stress for unirradiated and irradiated copper and nickel. Neutron dose, 5×10^{19} neutrons/cm². Irradiation temperature about 100°C. Strain rate, 8.2×10^{-5} sec⁻¹. After Makin (11)

has also been shown to be considerably increased by irradiation. For example, Broomfield et al. (14) studied austenitic steel, in which the effect is the same.

The greater temperature dependence of the yield strength of metals after irradiation is consistent with an increased strain rate sensitivity, since this also depends upon the operation of thermally activated deformation processes. The increased strain rate sensitivity has been observed by Makin (13) and Koppenaal (15). Makin and Minter (16) pointed out that in irradiated copper and nickel the ultimate tensile strength is almost as temperature-dependent as the yield strength, and the elongation to fracture after irradiation is almost independent of testing temperature. The same results were also obtained by Broomfield et al. (14) for austenitic steel.

The effect of purity on radiation hardening is rather conflicting and the results probably depend rather critically on the nature of the impurity or alloying addition. In many cases it would appear that alloying has very little effect on the increase in yield stress on irradiation and this is in agreement with the results of Koppenaal (17) on the hardening of Cu-Al single crystals. Laidler (18) found that the yield strength of 99.5% nickel increased by three times after a dose of 4.5×10^{18} neutrons/cm², whereas the yield strength of 99.8% nickel

increased only twice after a dose of 1×10^{19} neutrons/cm². However, the influence of interstitial impurities on the mechanical properties of irradiated face-centered cubic metals has not been studied widely. These effects are very complicated and there is no quantitative theory of the hardening of impure metals.

The irradiation temperature is expected theoretically to be a considerably important parameter in radiation hardening. This follows from the fact that the radiation defects or defect clusters, which are thought to be primarily responsible for the hardening, can depend sensitively on irradiation temperature. As the irradiation temperature is raised, the minimum size of defect cluster expected to be stable would be expected to increase, and this should result in a lower sensitivity of yield stress to testing temperature. Bruch et al. (19) studied the effect of neutron bombardment (1×10^{17} neutrons/cm²) on the yield strength of polycrystalline copper at several irradiation temperatures. Two irradiations at 80°C and 150°C gave the same value of yield strength. However, after an irradiation at 300°C, the yield strength was approximately the same as for the unirradiated control sample. In general, it appears that less hardening is produced for higher irradiation temperatures. Shober and Murr (20) studied comprehensively type 347 and 348 stainless

steels. At neutron dose of less than 10^{20} neutrons/cm², the increase in yield strength after irradiation at 290°C is smaller than that produced by an equal dose at an irradiation temperature below 100°C.

The mechanism of radiation hardening of face-centered cubic metals has been studied by many people, but there is still no universally accepted theory. The basic idea seems to be well established that irradiation-produced defects are capable of pinning dislocations. Moreover, the dimensions of the defects that do the pinning must be small. There are two pieces of experimental evidence involving isolated interstitials. First, when neutron irradiation at 20°K is made, the dislocation pinning degree is far in excess of the amount to be expected, i.e., some defect is able to traverse large distances from the point of its generation, permitting pinning to take place as much as one hundred or more atomic spacings from the nearest hit. The only defect that appears to be capable of this amount of travel at low temperature is the interstitial. Lomer and Cottrell (21) suggested this result by assuming a crowdion configuration. Silsbee (22) suggested that the recoil atom, hitting on a close-packed row of atoms, might through a focusing effect permit the appearance of an interstitial atom at long distance from the point of initial impact. Second, in favor of the

interstitial being responsible for dislocation pinning is the fact when the temperature of irradiation is raised from 20°K to 35°K, a considerable increase in the amount of pinning is obtained, and it is believed that the mobile defect in this temperature range is the interstitial. Thus, the interstitial seems to play an important role. However, it is also known that vacancies can cause radiation hardening. This is demonstrated, for example, in the experiments of Maddin and Cottrell (23) on the hardening produced in aluminum upon quenching. Since the formation energy for interstitials is too high to permit retaining interstitials upon quenching, we must assume that the quench hardening is due to vacancies. It follows, therefore, that radiation-produced vacancies may also be responsible in part for radiation hardening.

Neutron Irradiation Effects on the Mechanical Properties of Thorium

One reason that thorium was selected for study is the potential application to nuclear reactor technology. Thorium is a fertile material which breeds the fissionable uranium isotope U^{233} by exposing thorium to a neutron flux. In addition, thorium is three times more abundant than uranium in the earth's crust. The early studies of the

properties of thorium were directed toward application in breeder reactors.

From a more fundamental point of view, there are some interesting differences between thorium and other face-centered cubic metals. Thorium shows a relatively high carbon solubility at room temperature. Michelson and Peterson (24) reported the solubility limit at room temperature to be 3500 wt ppm carbon. The lowest value reported is that of Smith and Honeycombe (25) who report the limit to be 2000 wt ppm at 350°C. Several reports have revealed observable influences of carbon on the mechanical properties of thorium, e.g., the reports by Skaggs (26) and Peterson and Skaggs (27). These investigators mention notably: (1) the appearance of a yield drop whose magnitude increases with carbon content; (2) a strain-rate sensitivity which increases linearly with increasing carbon content; (3) the thermally activated component increases linearly with carbon content; (4) the additions of carbon sharply increase the room temperature strength and the temperature dependence of the flow stress. All these effects are normally associated with body-centered cubic metals and are very atypical of the face-centered cubic metals. Zerwekh and Scott (28) and Peterson and Skaggs (27) have thus concluded that the carbon atoms in thorium must interact strongly with dislocations,

thus controlling the deformation behavior below room temperature according to Friedel's random solute barrier mechanism (29). The reason for the unusually strong interactions of carbon atoms with dislocations in thorium is not known, however. Other physical and mechanical properties of thorium have been investigated in the past twenty years, and reviewed by Smith, Carlson, Peterson and Scott (30).

Neutron irradiation effects on the mechanical properties of thorium have not been studied extensively. Bowman (31) reported slight increases in hardness and strength after irradiation. Feustel and Reynolds (32) observed a doubling in strength after 2×10^{19} neutrons/cm². Kittel (33) and Holden (34) reported that thorium is not embrittled by neutron irradiation. Hammond (35) tested the effects of neutron irradiation on the room temperature mechanical properties of alloys containing 1, 4 and 5.4 wt. % uranium. The irradiation temperatures were below 250°C and neutron doses were from 2.5×10^{19} neutrons/cm² to 4.9×10^{20} neutrons/cm². The results indicated that irradiation increases the yield strength substantially, increases the ultimate strength only moderately, and decreases the ductility.

Bement (36) investigated the effects of neutron exposure to 5×10^{21} neutrons/cm² (1.9 at. % burnup) on the tensile properties and hardness of two lots of thorium

differing in carbon content, i.e., 350 wt ppm and 500 wt ppm. The calculated maximum irradiation temperature, based on maximum core temperature, varied from 145°C for 0.1% burnup to 315°C for 2% burnup. Significant increases in strength and hardness resulted from an exposure of 2×10^{20} neutrons/cm². Thorium containing an average of 500 wt ppm carbon, which was initially stronger than the 350 wt ppm carbon lot, continued to have a higher strength for the different exposures employed, indicating that the initial nature of thorium can influence properties resulting from irradiation. This exposure also induced a pronounced yield drop and a greatly decreased strain hardening ability. Recovery of radiation damage could be obtained by annealing at 750°C; however, some nonrecoverable damage occurred upon higher exposure.

In the present experiments, the effect of neutron irradiation on mechanical properties was studied for thorium containing two different carbon concentrations. The experimental plan included tensile tests of the unirradiated and irradiated samples as a function of temperature from 4°K to 500°K. By changing the strain rate during the tensile tests, the strain-rate parameter as a function of test temperature was also studied. Only the room temperature tensile tests were continued to fracture such that complete stress-strain curves were obtained.

It is worth noting that the high radioactivity and serious contamination problems from irradiated thorium have been considered and solved in this experiment. Consequently, valuable experience was gained and many handling and testing techniques were established that will be useful for further investigations.

EXPERIMENTAL PROCEDURES

Specimen Preparation

The thorium used in the present investigation was prepared by the reduction of thorium tetrachloride by magnesium at the Ames Laboratory. This method has been described by Peterson et al. (37). All specimens used in this investigation were prepared from thorium from one batch, in order to avoid composition variations. Half of the thorium was taken out to prepare the thorium-carbon alloy by arc-melting the thorium with a weighed amount of carbon in the form of high purity graphite electrodes. Chemical analyses of the pure thorium (Group P) and the thorium-carbon alloy (Group C) after melting are given in Table 1. In these analyses, the major substitutional impurities were determined by emission spectroscopy and the interstitial impurities were determined by either the vacuum fusion technique for oxygen, nitrogen and hydrogen or a combustion technique for carbon. There was no large difference between the substitutional and gaseous impurities of the two groups of specimens. The carbon analyses gave 110 and 1050 ppm for pure thorium and thorium-carbon alloy, respectively.

The tensile test specimens were prepared by swaging annealed 3/8 inch diameter rods to 0.097 inch in five

Table 1. Chemical analysis of annealed thorium specimens
in ppm by weight

Elements	Pure Th (Group P)	Th-C Alloy (Group C)
C	110	1050
O	190	210
H	7	8
N	16	17
Al	36	30
Be	<20	<20
Ca	<20	<20
Cr	<20	<20
Fe	<20	<20
Mg	<20	<20
Mn	<20	<20
Ni	≤20	≤20
Si	60	30
Zr	<20	<20
Ta	N.D.	N.D.
Ti	N.D.	N.D.

passes. After being cut into two-inch sections, the rods' surface was cleaned by electropolishing. The center one-inch gage section of every specimen was electropolished to 0.094 ± 0.001 inch diameter. Electropolishing was done in a perchloric acid (6%) and methyl alcohol (94%) electrolyte at -70°C temperature and using a D.C. potential of about 50 volts.

All specimens were furnace annealed at 750°C for one hour under 1×10^{-6} torr and then stored in an evacuated desiccator to avoid oxidation. The grain size of both groups of specimens was approximately 1200 grains per square millimeter.

Irradiation Procedure

The neutron irradiations were conducted in the V-1 vertical thimble of the Ames Laboratory Research Reactor. Using small amounts of thorium, three preliminary irradiation experiments were carried out in order to obtain the following information in preparation for the irradiation of the final test specimens:

- (1) The radiactivity and the radiochemical analysis of the energy spectrum of irradiated thorium.
- (2) The design of the capsule in order to prevent heavy water from leaking into the capsule through any welded thermocouple seals.

- (3) The effectiveness of thorium sheet (20 mil thick) in shielding thermal neutrons. The thorium sheet was found not to be effective.
- (4) The proper total irradiation time that could yield observable radiation hardening effects on the specimens and still maintain sufficiently low radiactivity.

The final specimens were sealed in a capsule which was made of 16 mils thick stainless steel tube, lined with 30 mils thick cadmium sheet to absorb thermal neutrons and the capsule was filled with helium. The specimens were held by an aluminum stand in the capsule to separate them from each other in order to prevent the outer samples from shielding the inner ones. After testing by a helium sniffer to insure that there was no leakage of the capsule, the capsule was installed in the V-1 position of the Ames Laboratory Research Reactor for 48 hours of irradiation. During the irradiation, the temperature was determined by two copper-constantan thermocouples which were inserted into separate wells in the capsule. The average voltage of two thermocouples was 25.2 ± 3.8 mV, this value indicated an irradiation temperature around $468 \pm 46^\circ\text{C}$, which was much higher than expected. After irradiation, the capsule was transferred to a water pool for 20 days cooling and then the samples were de-encapsulated in the hot cell. A diagram of the irradiation capsule is given in Figure 3.

Figure 3. Diagram of irradiation capsule

Upon post-irradiation examination, the specimens appeared to be in good condition and the cadmium liner showed no sign of having melted. Since the melting point of cadmium is 320°C , we conclude the irradiation temperature was below this value. The recovery temperature of thorium is about 525°C after Boyle (38). He also pointed out that thorium is fully recrystallized by annealing at 650°C or higher for time exceeding 20 minutes. So it can be expected that no recovery or recrystallization occurred during irradiation. The high irradiation temperature measured from the thermocouples may have been due to the gamma-heating produced by the stainless steel wells. Three nickel wires were packed with the specimens for neutron flux determination. The resulting fast dose was found to be $3.0 \times 10^{18} \text{ n/cm}^2$ ($E > 1 \text{ MeV}$).

The radioactivity levels in the two groups of thorium were different. Twenty-eight specimens of Group P showed a radioactivity of approximately 17 R/hr. at a distance of six inches twenty days after irradiation. The same amount of Group C specimens showed 34 R/hr. in the same condition. The specimen surfaces were cleaned by dilute nitric acid and Na_2SiF_6 solutions and then transferred in a lead cask to the hot cell for tensile testing.

Tensile Test Experiments

All of the tensile tests were performed on a screw-driven Instron Model 1125 Universal Testing Machine. This instrument includes a loading frame and control console as separate basic assemblies. The cross head speed accuracy is ± 0.1 % of set speed and the load weighing accuracy is ± 0.5 % of indicated load. The cross head speed was set at 0.002 in./min. and increased cyclically by a factor of 10 during many tests. Cross head speeds corresponded to nominal strain rates of 3.3×10^{-5} per second and 3.3×10^{-4} per second. Strain rate changes were made at intervals of from 0.01 to 0.02 inches elongation.

Because of radioactivity of irradiated thorium specimens, the tensile tests were conducted on the loading frame in the hot cell (Figure 4) and operated with remote control through the window. A special jig was fabricated to hold the grips (Figure 5). By using manipulators, the specimen were inserted into the grips for a length of 1/4 inch at each end by a serrated four jaw collet, which in turn, was secured by a tapered compression nut. This arrangement prevented slipping in the grips and resulted in all deformation produced within the one inch gage length. Then, the specimen with the grips was transferred by the

Figure 4. Load frame of tensile machine in the hot cell

Figure 5. Jig for mounting irradiated specimens in tensile grips remotely in the hot cell

manipulators to a loading fixture fixed on the loading frame of the Instron Machine. The specimen was placed in a fixture attached to the bottom of the cross head and the upper end of the specimen was attached to a pull rod connected with the load cell. In this way, the tensile tests were carried out by the downward movement of the cross head, and this enabled the specimens to be completely immersed in a bath for the sub-ambient and above-ambient temperature tests.

Tensile measurements were done at six different test temperatures. These involved liquid helium (4°K), liquid nitrogen (78°K), solid carbon dioxide and alcohol (196°K), heated silicon oil (423°K and 503°K) and room temperature (298°K). Liquid helium was transferred by a double wall tube through the hot cell wall into a double Dewar in which the specimen was immersed. The liquid helium was added continuously to compensate for losses while testing. The same method was used for liquid nitrogen, except that a single Dewar was used instead of a double Dewar. Silicon oil was heated by an immersion heating coil with an adjustable transformer to get stable temperatures and the oil was circulated by a stirrer. All test temperatures were controlled to within $\pm 2^{\circ}\text{K}$. A compressed air lifting frame, remotely controlled, was installed on the loading

frame for raising the Dewar or oil bath in order to immerse the specimens.

Two tensile tests were performed for each sample type. The specimen extension was determined by measuring the cross head movement shown on the Extension Measurement Unit on the control console, and rechecked with the testing time taken from the recorder chart.

EXPERIMENTAL RESULTS

Tensile specimens of irradiated and unirradiated pure thorium (Group P) were tested at room temperature and the complete stress-strain curves are plotted in Figure 6. The 0.1 percent offset yield stress was increased from 7.13 kpsi to 9.93 kpsi, i.e., it was increased about 39.3 percent due to irradiation. The ultimate tensile strength was increased from 17.56 kpsi to 18.04 kpsi, i.e., only 2.73 percent. It is observable that the work hardening rate of irradiated pure thorium specimens is much lower than for unirradiated ones. The apparent uniform elongation decreased from 35 to 27.5 percent upon irradiation, and the total elongation decreased from 47.8 to 42.3 percent. As can be seen in Figure 6, the unirradiated and irradiated samples fractured at about the same stress level (about 7 kpsi), indicating that the reduction in area was not greatly affected by the irradiation.

The thorium-carbon alloy (Group C) specimens were also tested at room temperature. The stress-strain curves are shown in Figure 7. The 0.1 percent offset yield stress was increased from 28.28 kpsi for the unirradiated specimen to 33.83 kpsi for the irradiated specimen, indicating an increase of 19.6 percent. The ultimate tensile strength was increased from 34.22 kpsi to 38.15

Figure 6. Engineering stress-strain curves for pure thorium (Group P) after irradiation to 3.0×10^{18} neutron/cm² ($E > 1$ MeV). Irradiation temperature, below 320°C

Figure 7. Engineering stress-strain curves for thorium-carbon alloy (Group C) after irradiation to 3.0×10^{18} neutron/cm² ($E > 1$ MeV). Irradiation temperature, below 320°C

kpsi, i.e., it was increased about 11.5 percent. This indicated again that the work hardening capacity is reduced upon irradiation. The uniform elongation was decreased from 23.8 to 17.5 percent; the corresponding decrease in total elongation was from 40.3 to 31.7 percent upon irradiation. From the fracture stresses shown in Figure 7, the irradiated specimen seems to indicate a smaller reduction in area than the unirradiated specimen. There are the yield drops also shown in Figure 7 for both irradiated and unirradiated specimens. These yield drop phenomena were only observed for thorium-carbon alloy at room temperature during present experiment.

The effect of irradiation on the yield stress of pure thorium and thorium-carbon alloys was investigated over the range of temperatures from 4° to 500°K. The variation of the 0.1 percent offset yield stress (for values see Appendix I) with temperature for irradiated and unirradiated thorium and thorium-carbon alloy specimens is shown in Fig. 8. A similar plot of 0.2 percent offset yield stress is shown in Figure 9. The strength of thorium was found to be a function of the carbon content and testing temperature. For a given carbon concentration the yield stress decreases approximately linearly with temperature from 4°K to about 300°K and then approaches a nearly temperature-independent value at approximately 500°K.

Figure 8. Temperature dependence of 0.1% offset yield stress for unirradiated and irradiated pure thorium and thorium-carbon alloy. Neutron dose, 3.0×10^{18} neutron/cm² ($E > 1$ MeV). Irradiation temperature, below 320°C

Figure 9. Temperature dependence of 0.2% offset yield stress for unirradiated and irradiated pure thorium and thorium-carbon alloy. Neutron dose, 3.0×10^{18} neutron/cm² ($E > 1$ MeV). Irradiation temperature below 320°C

As was shown earlier by Peterson and Skaggs (27), the temperature dependence of yielding increases with increasing carbon concentration. At the low temperatures, some of the stress-strain curves showed discontinuous drops; for example, the irradiated and unirradiated thorium-carbon alloy and irradiated pure thorium specimens at 4°K, and the irradiated thorium-carbon alloy material at 78°K. This may be due to the formation of twins, but would need to be verified by further tests.

After irradiation, however, the yield stress was increased for both thorium and thorium-carbon alloy specimens. The variation of the increase in 0.1% offset yield stress and the corresponding error bars with temperature are shown in Figure 10. For thorium-carbon alloy, the increase of yield stress is greater in the low temperature range. With increasing temperature above about 100°K, this value is decreased and then appears to be temperature independent at 400°K and above. The increase in yield stress due to irradiation for pure thorium specimens is also shown in Figure 10. The value is low at low temperatures and then increases with increasing temperature to the maximum at 300°K and then appears temperature-independent. The increase in yield stress due to irradiation for thorium and thorium-carbon

Figure 10. Temperature dependence of the increase in yield stress due to irradiation of pure thorium and thorium-carbon alloy. Neutron dose, 3.0×10^{18} neutron/cm² ($E > 1$ MeV). Irradiation temperature, below 320°C

alloy, by considering the error bars, seems to have the same values at about 300°K.

For comparison between these experimental results and those of Peterson and Skaggs (27), the variation of 0.1% offset yield stress with temperature for unirradiated thorium with four concentrations of carbon is shown in Figure 11. The curves from this experiment appear to be very similar to those of Peterson and Skaggs (27). At high temperatures, the yield stresses increase consistently with increasing carbon concentration. At low temperature, however, the yield stresses observed for the thorium - 1050 wt. ppm carbon alloy in the present investigation appear lower than those observed by Peterson and Skaggs (27) for thorium - 1000 wt ppm carbon. As is discussed below, this may be due to differences in grain size.

The strain rate parameter, $\Delta\sigma/\Delta\ln\dot{\epsilon}$, was evaluated at 1-2 percent strain by changing the strain rate from 3.3×10^{-5} per second to 3.3×10^{-4} per second at six different temperatures for irradiated and unirradiated thorium-carbon alloy and pure thorium specimens. This parameter shows a maximum at approximately 250°K for thorium-carbon alloy and 200°K for pure thorium specimens. The effect of temperature on the strain rate parameter is shown in Figure 12. There is only a slight increase in the strain rate parameter upon irradiation for the pure

Figure 11. Temperature dependence of 0.1% offset yield stress of high-purity unirradiated thorium for several concentrations of carbon. PS and CS are after Skaggs (26) and Peterson and Skaggs (27)

Figure 12. Variation of the strain rate parameter with temperature for irradiated and unirradiated thorium-carbon alloys. Neutron dose, 3.0×10^{18} neutron/cm² ($E > 1$ MeV). Irradiation temperature, below 320°C. PS and CS are after Skaggs (26) and Peterson and Skaggs (27)

thorium. However, for the thorium-carbon alloy below 320°K, the strain-rate parameter is significantly increased upon irradiation. Figure 12 also shows the temperature dependence of the strain rate parameter from the work of Peterson and Skaggs (27) for unirradiated thorium - 112 wt ppm carbon (PS) and thorium - 1000 wt. ppm carbon (CS). The curve for Peterson and Skaggs' 1000 wt ppm carbon material agrees very well with our curve for 1050 wt ppm carbon in the unirradiated condition (CU), but their curve for 112 wt ppm carbon (PS) lies significantly above our curve for unirradiated 100 wt ppm carbon (PU) below 400°K. Possible reasons for the discrepancy are discussed below.

From the observed changes in flow stress resulting from change in strain rate we can determine the activation volume, V^* , as defined by the expression (26),

$$V^* = KT \left(\frac{\partial \ln \dot{\epsilon}}{\partial \tau} \right)_{T, \epsilon} \approx KT \left(\frac{\Delta \ln \dot{\epsilon}}{\Delta \tau} \right)_{T, \epsilon} \quad (1)$$

By using the relation between the shear stress, τ , and the axial stress, σ ,

$$\tau = \frac{\sigma}{3.1} \quad (2)$$

and the lattice parameter of 5.086 Å (25°C), both from Skaggs (26), we can calculate V^*/b^3 as a function of test temperature for irradiated and unirradiated specimens,

as shown in Figure 13. The activation volumes for pure thorium and thorium-carbon alloy specimens are decreased slightly due to irradiation. The activation volume of unirradiated thorium with 112, 1000 and 1500 wt ppm carbon from Skaggs (26) are also shown in Figure 13. The difference of activation volume between 1000 and 1050 wt ppm carbon unirradiated thorium is not very large. But the 112 wt ppm carbon from Skaggs (26) is shown to have a much lower activation volume than the 110 wt ppm carbon thorium tested in this investigation.

Figure 13. Variation of the activation volume with temperature for irradiated and unirradiated thorium-carbon alloys. Neutron dose, 3.0×10^{18} neutron/cm² ($E > 1$ MeV). Irradiation temperature, below 320°C. PS, CSI and CS2 are after Skaggs (26) and Peterson and Skaggs (27)

DISCUSSION

The general effects of moderate neutron irradiation upon the stress-strain behavior of the pure and carbon-doped thorium at room temperature are illustrated in Figures 6 and 7, respectively. In both cases, there is an increase in yield and flow stress, a decrease in the rate of work hardening, and a decrease in uniform elongation. The increase in 0.1% offset yield stress is 2.8 kpsi for the pure thorium (Figure 6). The increase in 0.1% offset yield stress is 5.5 kpsi for the thorium-carbon alloy (Figure 7). The corresponding increases in lower and upper yield stresses for the thorium-carbon alloy are 6.3 and 6.1 kpsi, respectively. However, as is discussed below, the relative magnitudes of these increases in yield stress depend upon the test temperature. In general, the radiation hardening depends on the strength, density, and size distribution of the radiation-produced defect clusters in their action as barriers to slip dislocation motion. Also, the relative effectiveness of these barriers changes with deformation temperature.

One measure of the rate of work hardening is the difference between the ultimate and yield stresses. For the pure thorium this difference is 10.5 kpsi and 8.1 kpsi for unirradiated and irradiated material, respectively. For the thorium-carbon alloy, using the lower yield stress,

this difference is 7.4 and 5.2 kpsi for unirradiated and irradiated material, respectively. Thus, according to this criterion, the irradiation-induced decrease in the rate of work hardening at room temperature is about the same in the two materials.

The uniform elongation is usually designated as the strain corresponding to the ultimate tensile stress on the stress-strain curve. The fractional decreases in uniform elongation due to irradiation are 21 and 26 percent for pure thorium from Figure 6 and thorium-carbon alloy from Figure 7, respectively. The total elongation or the fracture strain is also shown in those two figures. The decrease in total elongation due to irradiation is 12 and 21 percent for pure thorium and thorium-carbon alloy material, respectively. These results thus indicate that the radiation-induced decrease in room temperature uniform elongation and total elongation is larger for thorium-carbon alloy than for pure thorium material. This will be discussed later in connection with the increase in yield stress upon irradiation for pure thorium and thorium-carbon alloy material.

Another measure of ductility is the reduction in area at fracture, which is defined as

$$R = 1 - \frac{A_f}{A_o} \quad (3)$$

where A_o is the initial cross sectional area and A_f is the cross sectional area at fracture. It may be possible to obtain an approximate value for R from Figure 6 and 7 as follows. Assuming constancy of volume up to the point of plastic instability, we may write

$$A_o \ell_o = A_u \ell_u$$

where ℓ_o is the initial gauge length and ℓ_u and A_u are the length and cross sectional area corresponding to the uniform strain. Then

$$\begin{aligned} A_o &= \frac{\ell_u}{\ell_o} A_u \\ &= (1 + \epsilon_u) A_u \end{aligned} \quad (4)$$

where ϵ_u is the uniform strain. Equation (4) in (3) gives

$$R = 1 - \frac{A_f}{(1 + \epsilon_u) A_u} \quad (5)$$

Now let us assume that the true stress $\bar{\sigma}$ remains constant at the neck during necking in order to maintain a constant strain rate. Also

$$\bar{\sigma} = \frac{P}{A}$$

where P is the load. Therefore, if $\bar{\sigma}$ remains constant during necking, we may write

$$\frac{A_f}{A_u} = \frac{P_f}{P_u}$$

and equation (5) becomes

$$R = 1 - \frac{P_f/P_u}{1 + \epsilon_u} \quad (6)$$

Since the stress shown in Figures 6 and 7 is engineering stress, we may determine P_f/P_u from the ratio of the stress at fracture to the maximum stress or ultimate tensile stress. Table 2 shows that the reduction in area is essentially unchanged upon irradiation for pure thorium. This result is similar to that for other face-centered cubic metals, as described by Makin (10). However, Table 2 indicates that the thorium-carbon alloy behaves quite differently in that the reduction in area is significantly decreased as a result of the irradiation.

The uniform strain, ϵ_u , and the reduction in area, R , for unirradiated thorium and thorium-carbon alloy were also determined by direct examination of the specimens. Table 2 (ϵ_u values in brackets) shows that the first visual observation of necking occurred at strains about 4% greater for both thorium and thorium-carbon material than the strains corresponding to the maximum engineering stress. This may be due to the inability to detect visually the earliest onset of necking.

Table 2. Reduction in area for unirradiated and irradiated pure Th and Th-C alloy at room temperature

	σ_u , ultimate tensile stress (kpsi)	σ_f , stress at fracture (kpsi)	$\frac{P_f}{P_u} = \frac{\sigma_f}{\sigma_u}$	ϵ_u , uniform strain	R, reduction in area
Pure Th, Un- irradiated, PU	17.6	7.0	0.398	0.35 [0.39] ^a	0.71 [0.84] ^a
Pure Th, Ir- radiated, PI	18.0	7.1	0.394	0.28	0.69
Th-C alloy, Un- irradiated, CU	34.2	11.4	0.333	0.17 [0.21] ^a	0.72 [0.77] ^a
Th-C alloy, Ir- radiated, CI	38.3	20.6	0.538	0.24	0.57

^aValues determined by direct visual examination.

For the direct determination of reduction in area at fracture, R , photographs were taken of the fracture surface, the surface areas were measured, and R was calculated according to Eq. (3). These values are shown in brackets in Table 2. It is seen that they are higher than the R values deduced from Eq. (6). This suggests that the assumption of constant true stress at the neck during necking is not quite correct. In fact, using the directly determined values of R we conclude that σ_f/σ_u has values of 1.8 and 1.2 for PU and CU material, respectively. The increase in true flow stress in the strain region past the uniform strain may be due to a combination of work hardening and strain-rate dependence.

The metal thorium was shown to possess a yield drop by Milko et al. (39) and Peterson and Skaggs (27). Peterson and Skaggs pointed out that the magnitude of the yield drop is a maximum at approximately 250°K, and diminishes with both increasing and decreasing temperature. The magnitude of the yield drop was also observed to increase with increasing carbon content and it was much larger in fine-grained than in coarse-grained specimens. In the present work, a yield drop was observed only for the thorium-carbon alloy at room temperature. At two other test temperatures, 196°K and 423°K, there were shown some slight yield drops by Skaggs (26) and Peterson and Skaggs (27), but this was not seen in this study. The carbon content, 1050 wt ppm, for our thorium-carbon alloy is quite similar to the 1000 wt ppm

carbon used by Skaggs (26). However, the grain sizes are quite different; 1200 grains/mm² for this study as compared to 3200 grains/mm² reported by Skaggs (26). The coarser-grained specimens are expected to exhibit a smaller yield drop and this seems to be a reasonable explanation for the different observations in the two studies. After irradiation the yield drop also appeared in the thorium-carbon alloy material at room temperature only (shown in Figure 7). Therefore, we can conclude that the yield drop in thorium is unaffected by moderate irradiation at the temperature studied. Bement (36) found no yield drop at room temperature for unirradiated thorium containing 350 and 500 wt ppm carbon alloys. However, a sharp yield drop at room temperature was shown for thorium-carbon alloys irradiated at 145°C to 2.0×10^{20} neutrons/cm². This dose was considerably higher than that used in this study. He observed that the development of a yield drop due to irradiation was more pronounced for the thorium-500 wt ppm carbon alloy than that for the thorium-350 wt ppm carbon alloy. Bement found also that the yield drop became less prominent and eventually disappeared at greater exposures ($>2.0 \times 10^{20}$ neutrons/cm²). He interpreted this to be due to fission products which increase the impedance to dislocation glide.

The shapes of the curves of yield stress as a function of test temperature (Figures 8 and 9) are analyzed according to several expressions. The model generally employed to

explain radiation hardening is the dispersed barrier model in which the radiation-produced obstacles to dislocation motion are distributed randomly on the slip plane. The temperature dependence of the yield stress in irradiated face-centered cubic metals is thought to depend upon the overcoming of these obstacles as a result of thermal motion. Several theories of the dispersed barrier hardening have been developed on the basis of thermally aided overcoming of dispersed obstacles. Hardening is proportional to the inverse of the average distance between the obstacles on slip planes and is generally considered to be proportional to the square root of the concentration of the obstacles (40). This rate-theory approach is useful in analyzing various deformation processes in terms of dislocation mechanisms. The ones most frequently cited are those due to Seeger (12)

$$\left(\frac{\tau}{\tau_0}\right)^{2/3} = 1 - \left(\frac{T}{T_0}\right)^{2/3} \quad (7)$$

and to Fleischer (41)

$$\left(\frac{\tau}{\tau_0}\right)^{1/2} = 1 - \left(\frac{T}{T_0}\right)^{1/2} \quad (8)$$

The expression by Seeger, Equation (7), is based on empirical interaction energy between the slip dislocation and the radiation-produced defect of the form

$$U(x) = U_0 \frac{1}{1 + \exp(-x/x_0)}$$

where $U(x)$ is the interaction energy at distance x and U_0 and x_0 are constant parameters. The applied shear stress τ is assumed to bring the dislocation part way past the barrier and the remaining energy to surmount the barrier, $H(\tau)$, is supplied by thermal activation. Equation (7) then follows from the assumption that the strain rate is given by a thermal-activation expression of the form

$$\dot{\epsilon} = \dot{\epsilon}_0 \exp\left(-\frac{H(\tau)}{kT}\right)$$

Fleischer's expression, Equation (8), is based on a similar analysis but the interaction energy between the slip dislocation and the barrier is taken to be appropriate to the case where the barrier is the center of a tetragonal lattice distortion. Dislocation motion is then envisioned as a series of thermally activated dislocation jumps through the tetragonal stress fields produced by individual solute atoms.

The shear stresses of irradiated and unirradiated thorium and thorium-carbon alloy material are calculated from the 0.1% offset yield stress by using equation (2), and then fitted to equations (7) and (8). The results are shown in Figures 14 and 15. The constants, T_0 and τ_0 , for equations (7) and (8) were calculated for least-squares fit and are listed in Table 3. The standard deviation, Δ_y , for the fits is also listed in Table 3.

Figure 14. The temperature dependence of the shear stress for unirradiated and irradiated pure thorium and thorium-carbon alloy plotted according to equation (7). Neutron dose, 3.0×10^{18} neutron/cm² ($E > 1$ MeV). Irradiation temperature, below 320°C.

Figure 12. The temperature dependence of the shear stress for unirradiated and irradiated pure thorium and thorium-carbon alloy plotted according to equation (8). Neutron dose, 3.0×10^{18} neutron/cm² ($E > 1$ MeV). Irradiation temperature, below 320°C

Table 3. The constants T_1 and τ_0 , and the standard deviation, Δ_y , for equations (7), (8) and (9) for the total shear stress τ

	Equation (7)			Equation (8)			Equation (9)		
	T_0 °K	τ_0 kpsi	Δ_y^a (kpsi) ^{2/3}	T_0 °K	τ_0 kpsi	Δ_y^a (kpsi) ^{1/2}	T_0 °K	τ_0 kpsi	Δ_y^a (kpsi)
Pure Th, Un-irradiated, PU	1484.6	8.57	0.057	874.5	7.53	0.139	288.8	1.19	0.335
Pure Th, Irradiated, PI	2095.8	9.41	0.034	1240.8	7.78	0.088	143.3	2.26	0.392
Th-C alloy, Unirradiated, CU	862.0	62.41	0.194	653.4	44.19	0.519	-793.1	-5.98	0.732
Th-C alloy Irradiated, CI	882.7	67.24	0.217	703.5	42.93	0.534	-855.6	-5.94	0.707

^a

$$\Delta_y = \left[\frac{\sum_{i=1}^n (y_i - \bar{y}_i)^2}{n-1} \right]^{1/2}$$

y_i : experimental values

$\bar{y}_i$: calculated values

n : number of points

In the analysis of the thermally activated aspects of plastic deformation, the applied yield stress is considered to consist of an athermal component, τ_g , and a thermal or effective component, τ^* . Since τ_g plays no role in the process of thermal activation, we can use τ^* in equations (7) and (8), instead of the full applied stress, τ . The athermal stress may be approximately chosen as the value at 503°K in Figure 8. When these τ_g 's are subtracted from the applied stress to yield τ^* as a function of temperature, the plots according to equations (7) and (8) shown in Figures 16 and 17, respectively, are obtained. The constants T_0 and τ_0 , and the standard deviation, Δ_y , in equations (7) and (8) for the effective shear stress are listed in Table 4.

Another empirical fit of the present results can be made to the simple expression discussed by Fisher (42) and Louat (43), in connection with the interaction between interstitial impurities and dislocations in steel, viz.

$$\frac{\tau}{\tau_0} = 1 + \frac{T_0}{T} \quad (9)$$

The plot according to this relation is shown in Figure 18. Of course, since the ordinate is linear in shear stress τ , the degree of fit to effective shear stress, τ^* , versus temperature would be the same as that in Figure 18. The

Figure 16. The temperature dependence of the effective shear stress for unirradiated and irradiated pure thorium and thorium-carbon alloy plotted according to equation (7). Neutron dose, 3.0×10^{17} neutron/cm² ($E > 1$ MeV). Irradiation temperature, below 320°C

Figure 17. The temperature dependence of the effective shear stress for unirradiated and irradiated pure thorium and thorium-carbon alloy plotted according to equation (8). Neutron dose, 3.0×10^{18} neutron/cm² ($E > 1$ MeV). Irradiation temperature, below 320°C

Table 4. The constants T_0 and τ_0 , and the standard deviation, Δ_y , for equations (7) and (8) for the effective shear stress τ^*

	Equation (7)			Equation (8)		
	$T_0, ^\circ\text{K}$	τ_0, kpsi	$\Delta_y (\text{kpsi})^{2/3}$	$T_0, ^\circ\text{K}$	τ_0, kpsi	$\Delta_y (\text{kpsi})^{1/2}$
Pure Th, Un-irradiated, PU	546.6	8.27	0.195	471.0	6.24	0.086
Pure Th, Ir-radiated, PI	558.8	10.34	0.130	512.5	6.08	0.050
Th-C alloy, Un-irradiated, CU	521.2	76.84	0.258	481.4	44.04	0.382
Th-C alloy, Ir-radiated, CI	500.9	108.31	0.038	480.8	46.60	0.436

Figure 18. The temperature dependence of the shear stress for unirradiated and irradiated pure thorium and thorium-carbon alloy plotted according to equation (9). Neutron dose, 3.0×10^{18} neutron/cm² ($E > 1$ MeV). Irradiation temperature, below 320°C

constant T_0 and τ_0 and the standard deviation, Δ_y , are also listed in Table 3.

Another interesting result indicated in Figures 8 and 9 is that the irradiated and unirradiated yield stress curves appear almost parallel for both pure thorium and the thorium-carbon alloy metals. This means that the increase in the yield stress due to irradiation appears to be mostly athermal in nature. Here, the results compare favorably with those of Arsenault (44) in the sense that the irradiation induced change in flow stress appears to be temperature independent. Wechsler et al. (45) found that the increase in shear stress is relatively insensitive to test temperature in niobium single crystals irradiated to 2.8×10^{17} neutrons/cm². Smolik and Chen (46) also reported that neutron irradiation makes a contribution to only the athermal component of the yield stress in vanadium. Wechsler et al. (47) studied ASTM A 212, Grade B, pressure vessel steel irradiated to 1×10^{19} neutrons/cm² ($E > 1$ MeV) at 60°C. They carried out the tensile tests at temperatures from 20° to 273°K and found that the yield stress increased rapidly with decreasing test temperature and the radiation-produced increment in yield stress was largely independent of test temperature.

However, on a much finer scale, as shown in Figure 10, it appears that the amount of the increase in yield stress

due to irradiation is varying slightly with test temperature, especially at lower temperatures. The dependence with temperature is different for pure thorium and thorium-carbon alloy, i.e., the increase in yield stress for pure thorium is decreasing and for thorium-carbon alloy is increasing with decreasing temperature. But at 300°K and above, this tendency does not exist. Here, the difference in the yield stress increment due to irradiation is not very different for the two kinds of metals, and, considering the error bars shown in Figure 10, the small changes in the increase in yield stress for both of the alloys may not have a very significant meaning.

Previously, it was convenient to divide the stress into two components, one due to a long-range or athermal stress and another stress due to short-range obstacles or thermal barriers where thermal fluctuations can assist dislocations to surmount these barriers, and thus contribute to a temperature dependent stress. In considering the increase in yield stress with decreasing test temperature for unirradiated thorium, Peterson and Skaggs (27) concluded that the carbon increases the strength of thorium by introducing a large number of short-range obstacles to dislocation motion. The carbon atoms are in a nearly random solid solution, and these solute atoms are considered not able to diffuse rapidly up to 500°K. Furthermore,

thermal activation can assist the applied stress in moving dislocation past the solute atom obstacles.

For the irradiated thorium-carbon alloy metal, the increase in yield stress due to irradiation, in Figure 10, varies with the test temperature below 300°K. This variation can be interpreted as follows. During irradiation, the impurity carbon atoms, which are dissolved in the metal, may serve as nucleation centers to which the radiation produced defects aggregate and form defect clusters. Bajaj and Wechsler (48) studied vanadium with different oxygen levels irradiated to 1.4×10^{19} neutrons/cm² ($E > 1$ MeV). From analysis by transmission electron microscopy, the defect cluster density increased with increasing oxygen concentration and the average defect cluster size decreased. They interpreted this increase of density in terms of the nucleation of defect clusters at isolated oxygen atoms, small oxygen aggregates, or oxygen-defect complexes. They then attributed the decrease in defect cluster size to the conservation of radiation-produced point defects. In the present experiment, this model (48) implies that carbon atoms act as the nucleation centers. Thus, the defect cluster density would increase with increasing carbon concentration and the average defect cluster size would decrease. This kind of defect cluster could act as short-range obstacles. Thus, as the

temperature is decreased, thermal fluctuations would diminish and could not assist dislocations to surmount these barriers, which would then increase the yield stress.

Because of the low concentration of carbon (110 wt ppm) in the pure thorium metal, the carbon atoms could combine with the point defects during irradiation, and thus decrease the total number of solute carbon atoms in solution. Usually, the solute carbon atoms are considered to be thermal barriers in the thorium metal, e.g., by Peterson and Skaggs (27). A decrease of solute atoms upon irradiation would thus decrease the number of the thermal barriers. This would produce a decrease in the radiation hardening, which could explain why the increase in yield stress is decreasing with decreasing test temperature, i.e., below 300°K, as shown in Figure 10.

The grain boundaries usually can increase the yield and flow stresses of metals, and this phenomenon is generally described by the Hall-Petch relation (49,50) between the yield stress and the grain size. Peterson and Skaggs (27) studied the contribution of grain boundaries to the total strength of thorium-carbon alloy by testing a series of extremely coarse-grained and fine-grained specimens. By heating their fine-grained material (3200 grains/mm²) at 1350°C for 45 hours, the specimens contained 1 to 3 grains per cross section (about 0.2 ~ 0.6 grains/mm²). The 2

percent offset strength of those coarse-grained specimens at 4.2°K was about 10 percent lower than that for the corresponding fine-grained material. They concluded that the contribution of the grain boundaries to the strength of thorium was small. The results, Figure 11, show that the 0.1% offset yield stress for the thorium-1050 wt ppm carbon alloy with 1200 grains/mm² is about 10 percent lower than that for the thorium-1000 wt ppm carbon alloy with 3200 grains/mm² at 4.2°K (Peterson and Skaggs (27)).

It is shown in Figure 12 that the strain-rate parameter is increased slightly due to irradiation. Rühle (51) also found an increase in the strain-rate parameter for copper single crystals irradiated to 4×10^{17} neutrons/cm². This is attributed to short-range obstacles produced by irradiation and the idea that overcoming these obstacles is a strong function of the strain rate. Thus, we have another evidence that some of the irradiation-produced defect clusters, which increase the strain-rate parameter, may be short-range obstacles. For the thorium-carbon alloy below 300°K, as shown in Figure 12, the strain-rate parameter is significantly increased upon irradiation. The increase in yield stress due to irradiation for thorium-carbon alloy is larger in the same temperature range, as was shown before. These two results agree with the short-range obstacle concept. The curve, from Peterson and Skaggs' result (27),

for the 112 wt ppm carbon (PS) as shown in Figure 12, lies significantly above our curve for unirradiated 100 wt ppm carbon (PU). The reasons may be due to the difference in grain size, the change of the strain rate, and the strain at which the strain rate was changed.

The activation volume is decreased somewhat upon irradiation for both thorium and thorium-carbon alloy, as shown in Figure 13. Arsenault and Pink (52) reported the activation volume decreased for the single crystal vanadium irradiated to 1.6×10^{19} neutrons/cm². The activation volume represents the average volume of dislocation structure involved in the deformation process. Because the activation volume was calculated from the strain-rate parameter by using equations (1) and (2), the difference between the PU and PS material as shown in Figure 13 may be due to similar reasons as discussed for the difference of the strain-rate parameter of PU and PS shown in Figure 12.

The twinning that occurred at low temperatures for thorium and thorium-carbon alloy upon irradiation is not usually observed in other face-centered cubic metals, and is more typical of body-centered cubic metals. This observation requires further investigations before a reasonable explanation can be suggested.

SUMMARY

The effects of neutron irradiation upon the mechanical properties of thorium and thorium-carbon alloy may be summarized as follows:

- a) The yield and flow stresses are increased and the rate of work-hardening and the uniform and total elongation are decreased due to irradiation for both pure thorium and thorium-carbon alloy metal.
- b) The decrease in reduction in area upon irradiation is larger for thorium-carbon alloy than for pure thorium. The yield drop phenomenon of the thorium-carbon alloy is unchanged after irradiation.
- c) The variation of shear stress and effective shear stress with test temperature can be fitted to Seeger's and Fleischer's equations for irradiated and unirradiated thorium and thorium-carbon alloy.
- d) Irradiation apparently contributes mostly an athermal component to the yield strength. However, some thermal component is detected in the low temperature range.
- e) Thorium-carbon alloy shows some increase in radiation hardening with decreasing test

temperature. This is attributed to a distribution of small sized radiation-produced defect clusters.

- f) Strain-rate parameter is increased and activation volume is decreased slightly for both kinds of metal upon irradiation..

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Appendix I. The measured values of the 0,1% offset yield stress at the various test temperatures

	4°K	78°K	196°K	298°K	423°K	503°K
Pure Th, Unirradiated, PU	23.17 23.48	16.82 16.82	10.51 10.59	8.45 7.49	5.16 5.22	5.22 5.05
Pure Th, Irradiated, PI	24.05 24.33	18.72 19.80	13.36 14.66	10.20 11.14	8.99 8.54	7.98 7.78
Th-C alloy, Unirradiated, CU	114.31 115.78	93.42 94.86	58.87 57.0	27.79 28.38	14.95 15.51	12.91 13.20
Th-C alloy, Irradiated, CI	120.21 122.31	99.22 101.37	63.61 62.70	31.47 32.75	18.00 17.11	16.15 14.51