EXAMINATION OF THE IRRADIATED 6061 ALUMINUM HFIR TARGET HOLDER

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*On attachment from Australian Atomic Energy Commission Research Establishment, Lucas Heights, New South Wales.

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CONTENTS

	<u>Page</u>
Abstract	1
Introduction	1
Fabrication and Irradiation History	4
Specimen Preparation	7
Measurement of Density Changes	7
Mechanical Property Measurements	17
Postirradiation Annealing Studies	31
Discussion	33
Conclusions	42
Acknowledgements	43
Appendix 1	45
High Temperature Tests	45
Appendix 2	50
Examination of the Weld Seam	50
Appendix 3	55
Table 7	55
References	58

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ABSTRACT

A 6061 aluminum alloy target holder from the High Flux Isotope Reactor (HFIR), originally in a precipitation-hardened condition, was examined for radiation damage after exposure to a maximum fast neutron fluence of 9.2×10^{22} neutrons/cm² ($E > 0.1$ MeV) and a thermal fluence of 1.38×10^{23} neutrons/cm² ($E < 0.414$ eV) at $\sim 60^\circ\text{C}$. Voids and a precipitate of transmutation-produced silicon were found to cause about 1.1% internal swelling, and a surface oxide scale contributed additional dimensional change. Significant strength increases were measured at test temperatures in the range 25 to 200°C and are attributed to the silicon precipitate and to irradiation-induced dislocations. There was a corresponding loss of ductility which was particularly severe at 200°C for slow strain rate testing conditions. The fracture mode appeared to remain transgranular over the range of test temperatures. Specimens tested at temperatures between 200 and 500°C , at which the microstructure was unstable, showed considerable loss in strength but a concurrent gain in ductility. The strength and ductility of specimens containing 4043 aluminum weld metal were found to be similar to those of the 6061 aluminum-base metal.

INTRODUCTION

The High Flux Isotope Reactor (HFIR) target holder is a hollow cylinder of heat-treated 6061 aluminum alloy. Its position in the core of the reactor is shown in Fig. 1. The function of the holder is to aid in positioning the target assembly and to facilitate removal of the target. The target holder studied here was removed from the reactor

*On attachment from Australian Atomic Energy Commission Research Establishment, Lucas Heights, New South Wales.

Fig. 1. Schematic Diagram Showing Location of Target Holder in HFIR Core.

after 65,713 Mwd of service, equivalent to fluences of 9.2×10^{22} neutrons/cm² ($E > 0.1$ MeV) and 1.38×10^{23} neutrons/cm² ($E < 0.414$ eV). Despite this heavy irradiation it was in good physical condition and had performed its duties satisfactorily. This report details the examination of the target holder.

The 6061 aluminum alloy contains about 1% Mg and 0.5% Si which on suitable heat treatment react to form a finely divided precipitate of Mg₂Si. High strengths can be attained by thermal manipulation of this precipitate. Moreover, the alloy in its optimum age-hardened state is relatively ductile with over 17% total tensile elongation in the temperature range 100 to 300°C.¹

The effects of relatively low fluence neutron irradiation on the mechanical properties of this alloy have been investigated previously. McCoy and Weir² studied the effects of irradiation up to 10^{19} neutrons/cm² ($E > 2.9$ MeV) and 10^{20} neutrons/cm² (thermal), at temperatures ranging from 25 to 450°C, on the tensile properties of 6061 aluminum in the -T6 (optimum hardness), -O (annealed), and -O plus 20% cold-worked conditions. No irradiation-induced property changes were found for specimens in any condition irradiated at or above 150°C. Specimens irradiated at lower temperatures exhibited a small increase in strength and a reduction in ductility which were attributed to changes in the work-hardening coefficient. Steele and Wallace³ irradiated 6061 aluminum in the -O and -T6 conditions to 1×10^{20} fast neutrons/cm² and 1.26×10^{21} fission neutrons/cm² at about 65°C, and found larger strength increases in room temperature postirradiation tests than those reported by McCoy and Weir for their lower fluence irradiations. Steele and Wallace found the ductility of the 6061-O aluminum to be reduced by irradiation, whereas the ductility of 6061-T6 aluminum was unaffected. The strain hardening exponent decreased from 0.26 to 0.18 for the 6061-O specimens and remained unchanged for 6061-T6 specimens after irradiation. Gronbeck⁴ observed strength increases at room temperature in excess of 20,000 psi after exposure to 1.6×10^{22} neutrons/cm² ($E > 1$ MeV). His material was of unknown temper with an initial hardness of R_B^{44} compared with R_B^{56} for the 6061-T6 aluminum studied by McCoy and Weir. The strength increases were accompanied by a slight loss in ductility.

These data do not allow the extent of property changes at high neutron fluences to be forecast with confidence. Moreover, the irradiation-induced microstructural modifications responsible for the changes in mechanical properties have not been determined. Also, the effects of irradiation on weldments in the alloy have not been investigated. The irradiated HFIR target holder provided a unique opportunity to study the mechanical properties and microstructure of age-hardened 6061 aluminum alloy exposed to high fast and thermal neutron fluences.

FABRICATION AND IRRADIATION HISTORY

The HFIR Model "C" target holder is a perforated cylinder fabricated from 0.0625-in.-thick 6061 aluminum alloy sheet⁵ of chemical composition shown in Table 1. A blank of sheet material 10.82-in. wide x 29.5-in. long was rolled to a semicircular cross section with a 5.25-in. radius, annealed at 232°C, rolled to a 3.75-in. radius, annealed again, and finally rolled to a 2.1875-in. radius. This shape was annealed to the -O temper by standard heat treatments¹ and seam welded with 4043 Al alloy filler wire to form a right cylinder. The weldment was planished and the final heat treatment consisted of solution annealing at 515 to 527°C for 20 min, "quenching" in circulating air to room temperature, and then aging at 190^{+5}_{-25} °C for 7 hr. This heat treatment was designed to minimize thermal distortion and to produce a condition similar to the optimum age-hardened condition, the T6 temper. Perforations were then drilled into the cylinder to produce the final configuration shown in Fig. 2.

The holder we examined was irradiated in the HFIR for a total of 65,713 MWd. The neutron fluence values quoted in this report are derived from the calculated neutron flux distributions in the HFIR core.⁶ The neutron fluences for the holder at the reactor horizontal midplane were 9.2×10^{22} neutrons/cm² ($E > 0.1$ MeV) and 1.38×10^{23} neutrons/cm² ($E < 0.414$ eV). The neutron fluences at various other points along the vertical length of the holder were estimated from a symmetrical axial flux distribution for neutrons of energy greater than 0.183 MeV that is strictly applicable only along the vertical center line of the reactor

Table 1. Chemical Composition of the HFIR Target
Holder Sleeve before Irradiation

Element	Percent Present
Al	Balance
Cr	0.22
Cu	0.21
Fe	0.45
Mg	0.99
Mn	0.06
Ni	< 0.01
Pb	0.02
Si	0.65
Sr	0.02
Ti	0.05
Zn	0.06

Fig. 2. Composite Photograph of As-Irradiated Target Holder. Weld seam runs longitudinally. The lengths A and B have no meaning in this report.

at the start of a fuel cycle. The ratio of thermal to fast neutron fluxes was assumed to remain constant along the length of the cylinder. In reality, this ratio probably increases markedly at distances greater than ~ 10 in. from the reactor horizontal midplane.⁷

The operating temperature of the holder was slightly higher than than of the coolant water (55°C) with which it was in contact, but the exact temperature is related to the oxide film that formed on the surfaces during irradiation. Uncertainties in the time-dependent thickness and composition of this oxide film induce uncertainty in calculations of the true operating temperature of each region of the sleeve. The oxide film on sections from near the reactor midplane was determined by optical metallography to be approximately 0.0023-in. thick on each surface exposed to reactor coolant. Calculations of the steady-state temperature gradient across the oxide due to nuclear gamma-heating indicate an end-of-life operating temperature of 65°C for regions of the target holder near the reactor midplane. Thus, the irradiation temperature is estimated to have been between 55 and 65°C .

SPECIMEN PREPARATION

The perforations in the target holder precluded the machining of standard tensile specimens. Consequently, specimens with a dogbone shape were machined as shown in Fig. 3. These specimens had no parallel gauge length. At their narrowest point in the gauge section they measured 0.15-in. $\times$ 0.0625-in. Longitudinal strips of specimens were cut; one strip contained the weld seam. Each longitudinal strip produced about 16 dogbone specimens. Control specimens were machined from an unirradiated cylinder with the same fabrication history.

MEASUREMENT OF DENSITY CHANGES

Irradiation-induced volume changes were studied by immersion density measurements on a set of untested dogbone tensile specimens. First, the density determinations were made on specimens with the oxide film as formed in the HFIR intact. The oxide was then stripped from the specimens in a solution of 900-ml H_3PO_4 and 100-ml HNO_3 at 60 to 80°C and

ORNL-DWG 72-3037

Fig. 3. Dimensions of the Dogbone-Shaped Tensile Specimen.

the densities were redetermined. The results of these measurements are summarized in Table 2 relative to the density of an oxide-free unirradiated control specimen, $2.7012 \pm 0.0005 \text{ g/cm}^3$.

Clearly, the oxide is less dense than the base metal, and the apparent density changes due to the oxide film are greater than those caused by defects within the metal. X-ray powder diffraction patterns of oxide layers on some other HFIR aluminum components suggest⁸ that the oxide is $\beta\text{-Al(OH)}_3$ (bayerite) and/or possibly $\alpha\text{-Al(OH)}_3$ (gibbsite) with theoretical densities of 2.53 and 2.42 g/cm^3 , respectively.⁹ Table 2 shows the total oxide thickness required to give the observed differences in density between the two sets of measurements, assuming the oxide is either bayerite or gibbsite. A total oxide thickness of 0.0047 in. was measured on specimen TH-136 which operated near the reactor midplane position. This is significantly lower than both of the calculated values. Metallographic observations of cracks and pores which are distributed throughout the oxide provide a qualitative explanation for this discrepancy, since the effective density of the porous oxide will be smaller than the theoretical density of the fully dense oxide.

MICROSTRUCTURAL EXAMINATION

The microstructure of the unirradiated HFIR target holder was examined by optical microscopy and by transmission electron microscopy. The structure contained many inclusions (Fig. 4). Electron microscopy revealed (Fig. 5) a fine needle-shaped precipitate of Mg_2Si which is responsible for age-hardening in the 6000 series aluminum alloys.¹⁰⁻¹³ This precipitate is coherent with the matrix,^{12,13} but when the needles are large, $> 1 \mu\text{m}$ in length, they become incoherent.¹³ Regions around the inclusions are free of the fine Mg_2Si precipitate. There are also precipitate-free zones at grain boundaries and around large Mg_2Si needles. These large Mg_2Si precipitates and the associated precipitate-free zones are not typical of a T6 temper and are probably a result of overaging or of slow cooling after solution treatment.

After irradiation, voids were observed in specimen TH-136 taken from the maximum flux region of the holder (Fig. 6). These voids were less

Table 2. Relative Change in Density of Dogbone Tensile Specimens Induced by Irradiation and Oxide Films, and Thickness of Oxide Layer Required to Cause Change

Specimen Number	Neutron Fluence ($\times 10^{22}$ n/cm ²)		Density Change ^a		Total Oxide Thickness ^d (in.) Required to Cause Observed Difference	
	(E < 0.414 eV)	(E > 0.1 MeV)	With Oxide ^b	Without Oxide ^c	Gibbsite	Bayerite
TH-144	?	1.2	-0.99	-0.02	0.006	0.010
TH-130	4.2	2.8	-0.99	-0.14	0.006	0.010
TH-142	6.7	4.5	-1.13	-0.18	0.006	0.011
TH-132	9.0	6.0	-1.61	-0.60	0.006	0.011
TH-140	10.9	7.3	-1.92	-0.76	0.008	0.014
TH-134	12.5	8.3	-2.22	-0.92	0.008	0.016
TH-138	13.3	8.9	-2.36	--	--	--
TH-136	13.8	9.2	-2.53	-1.06	0.01	0.018

^aCompared with the density of an unirradiated oxide-free specimen, 2.7012 g/cm³.

^bPrecision of $\pm 0.3\%$.

^cEstimated precision of $\pm 0.05\%$.

^dPrecision of ± 0.0015 in.; calculation is based on original metal thickness of 0.0625 ± 0.0015 in.

Fig. 4. As-Polished Sections Through an Unirradiated Target Holder Showing Inclusions. Top, longitudinal section; bottom, transverse section.

Fig. 5. Microstructure of 6061 Aluminum from an Unirradiated Target Holder. (a) Inclusions near grain boundaries which are denuded of the fine coherent Mg_2Si precipitate. Regions of the matrix imaged in absorption contrast reveal only large Mg_2Si precipitates (arrowed) and do not show the fine coherent Mg_2Si precipitate. (b) At higher magnification, denuding of the fine coherent Mg_2Si precipitate is clearly demonstrated around some of the large Mg_2Si precipitates as well as near the grain boundary.

Fig. 6. Microstructure of 6061 Aluminum Irradiated at $\sim 60^\circ\text{C}$ to Fluences of 9.2×10^{22} neutrons/cm² ($E > 0.1$ MeV) and 1.38×10^{23} neutrons/cm² (thermal) Showing Radiation-Induced Voids. (a) General view at low magnification. (b) Higher magnification showing clustering of voids. (c) Examples of voids undergoing coalescence within the void clusters.

Figure 6c

sharply faceted than those formed in other irradiated aluminum and aluminum alloys,¹⁴⁻¹⁷ and appeared to be located preferentially near large Mg_2Si precipitates and large inclusions. This apparent clustering of voids was confirmed by stereo microscopy and by the fact that there was a high incidence of clear cases of void coalescence within the clusters (see arrows in Fig. 6c), a feature not usually observed in metals where the mean void concentration is only $2 \times 10^{14}/\text{cm}^3$ as it was here.

The voids that are seemingly not associated with inclusions and large Mg_2Si precipitates in Fig. 6 are probably those associated with particles that were viewed end-on or which were located just outside the final foil section. Enhanced void concentrations were also evident near grain boundaries but no voids were observed on the grain boundaries. Comparison with the microstructures of the unirradiated control specimens in Fig. 5 suggests that the voids within the grains are clustered primarily in those regions of the unirradiated material that were originally denuded of the fine Mg_2Si precipitate.

Void size distributions for two fluence levels are shown in Fig. 7. Foil thicknesses were determined by stereo microscopy to be about 3000 Å. More than 1000 voids were measured for each fluence. The void concentrations and their average and maximum sizes increase with increasing neutron fluence. No voids were detected in foils from specimen TH-144 (1.2×10^{22} fast neutrons/ cm^2) and only a few were observed, near large precipitates only, in specimen TH-130 (2.8×10^{22} fast neutrons/ cm^2).

The precipitate structure in 6061 aluminum after neutron irradiation is more complex than that of the unirradiated alloy. A high density of small precipitate particles are visible in Figs. 6b and c which were recorded under absorption contrast conditions so as to enhance the visibility of voids and make the Mg_2Si precipitate invisible. We believe that the precipitate seen in Figs. 6b and c is transmutation-produced silicon as identified in other alloys in earlier studies.¹⁸ This silicon derives largely from a thermal neutron reaction, $^{27}\text{Al}(n,\gamma)^{28}\text{Al}$, followed by a β -decay to ^{28}Si . The high concentration of silicon particles resulted in image overlapping which made counting difficult. Also many of the particles were too small and displayed too little contrast for us

ORNL-DWG 72-3039

Fig. 7. Size Distributions of Voids Measured for Two Fast Neutron Fluences. $\bar{N}$ is the average concentration of voids.

to observe and measure them with confidence. Their concentration was at least $1 \times 10^{16}/\text{cm}^3$ and was apparently independent of neutron fluence over the range of fluences examined. The average size of the particles seemed to increase with fluence. In specimens irradiated to the highest fluence the silicon particle sizes ranged from $< 40 \text{ \AA}$ diam to a maximum of $\sim 320 \text{ \AA}$, with a most frequent size of $\sim 150 \text{ \AA}$.

The high strength of the unirradiated alloy is due to the coherent Mg_2Si precipitate and thus it is important to determine the effect of neutron irradiation on the stability of these coherent precipitates. Unfortunately, the irradiated alloy contained a high density of silicon particles and image overlap rendered contrast analyses extremely difficult. Nevertheless, the large Mg_2Si needles appeared to be unaltered by irradiation, and some of the fine Mg_2Si precipitate was detected in thin regions of irradiated foils where image overlap was minimized. Here the characteristic streaking of precipitate reflections in selected area diffraction patterns appeared similar to the unirradiated alloy, implying that no major irradiation-induced changes occurred in the morphology of the coherent precipitate. Problems of image overlap in diffraction contrast also interfered with attempts to study the dislocation structure in the irradiated alloy. The control material contained about 2×10^9 dislocations/ cm^2 , which increased to about 2 to $3 \times 10^{10}/\text{cm}^2$ in specimens irradiated to fast fluences in the range 3 to 9.2×10^{22} neutrons/ cm^2 . These dislocations (Fig. 8) were irregular segments which may be parts of loops larger than the foil thickness. No small dislocation loops were observed. A summary of quantitative measurements of the irradiation-induced defects is presented in Table 3.

MECHANICAL PROPERTY MEASUREMENTS

The dogbone specimens were tested in air in an Instron tensile testing machine using special grips to accommodate the specimen shape. For all practical purposes, the test specimen may be considered as a sheet tensile specimen with zero gage length and, hence, strain and strain rate cannot be defined. Nevertheless, the curves for applied load versus the Instron crosshead displacement (Fig. 9) were of similar

Fig. 8. Irradiated Specimen Examined Under Diffraction Contrast to Reveal the Dislocation Structure and a High Concentration of Fine Precipitate Particles.

Table 3. Microstructural Parameters in As-Irradiated 6061 Aluminum Target Holder

Neutron Fluence ($\times 10^{22}$ n/cm ²)		Concentration of Voids ($\times 10^{14}$ /cm ³)	Mean Void Diameter (Å)	Volume Fraction of Silicon Precipitate	Typical Silicon Particle Size (Å)	Dislocation Density ($\times 10^{10}$ lines/cm ²)
(E < 0.414 eV)	(E > 0.1 MeV)					
0	0	0	0	0	0	0.2
4.2	2.8	0.1	200	0.011	~100	2.3
9.0	6.0	1.4	310	0.024	~125	2.7
13.8	9.2	1.9	370	0.037	~150	2 to 3

Fig. 9. Typical Load-Extension Curves Obtained from the Dogbone Specimens. The stress and strain parameters are indicated. CHS is the Instron crosshead speed.

form to those normally obtained from standard tensile specimens. Engineering stress values were computed from the pertinent load values and the minimum original cross-sectional areas of the individual specimens. The actual cross-sectional area included the oxide film which fractured readily at low strains during testing and probably did not bear any load at the ultimate load point. The yield stress was arbitrarily defined at a crosshead displacement of 0.00002 in. from the apparent elastic region. The magnitudes of the crosshead displacement between the yield point and the maximum load, and between the yield point and the point of fracture, were arbitrarily chosen as measures of the uniform and total plastic extensions, respectively, as illustrated in Fig. 9. The former is more accurately characterized as the crosshead travel before large-scale plastic instability occurred. The use of these stress and strain parameters is believed to provide an internally consistent comparison between irradiated and unirradiated specimens, but the data are not from standard tests and hence are not suitable for engineering design purposes.

Because of the reactor-oriented nature of this investigation the tests were performed at temperatures ranging between room temperature and 200°C, the anticipated maximum range of operating temperatures for this alloy in the HFIR. A period of 20 min. was allowed to heat the specimens to the test temperature. Strain rate effects were measured by changing the crosshead speed. A few specimens were tested at temperatures between 200 and 500°C but at such high temperatures the microstructure was found to be unstable and quite different from that at lower temperatures. These high temperature tests are therefore considered as a separate item under Appendix 1. Tests on specimens containing the seam weld are discussed in Appendix 2. A compilation of all the test data is given in Appendix 3.

The effects of neutron fluence and test temperature at several crosshead speeds on the yield stress and ultimate stress are shown in Figs. 10 and 11, respectively. The data for the higher fluence specimens as a function of test temperature are presented in Fig. 12. There are significant irradiation-induced increases in strength, particularly at

Fig. 10. Variation of Yield Stress of Dogbone Specimens with Neutron Fluence and Test Temperature at Two Crosshead Speeds.

Fig. 11. Variation of Ultimate Stress of Dogbone Specimens with Neutron Fluence and Test Temperature at Two Crosshead Speeds.

Fig. 12. The Yield and Ultimate Strengths of the High Fluence Specimens as Functions of Test Temperature and Crosshead Speed.

the lower test temperatures, but the strengths appear to be approaching saturation (i.e., rate of change of strength with neutron fluence is negligible) at fluences above about 4×10^{22} neutrons/cm². There was no discernable effect of crosshead speed at test temperatures below 150°C. At 200°C, tests at the slowest crosshead speed showed no irradiation-induced strength increase but an appreciable increase was noted at the highest crosshead speed.

The ductility of the specimens is shown in Figs. 13 and 14. Under all conditions of fluence, test temperature and crosshead speed the plastic extension to the ultimate stress was reduced slightly by the irradiation. On the other hand, the total plastic extension was markedly reduced, particularly at a test temperature of 200°C and the slowest crosshead speed.

EXAMINATION OF FRACTURED SPECIMENS

Specimens tested at the slowest crosshead speed were examined to determine the mode of fracture. In Fig. 15 it can be seen that for test temperatures of 25 and 200°C the unirradiated specimens necked more extensively near the fractures than irradiated specimens. In both irradiated and unirradiated specimens at all test temperatures, holes formed at large inclusions near the fracture face, either by cracking of the inclusion or by tearing of the matrix at the inclusion (Fig. 16). Final fracture appeared to occur by coalescence of such holes. Scanning electron microscopy of the fracture surfaces revealed only dimpled surfaces, characteristic of transgranular tearing and shearing (Fig. 17). Because of the large holes it was not possible to obtain useful transmission electron microscopy specimens from regions near the fracture to determine whether deformation had changed the microstructure. Specimens taken from regions away from the fractures showed that the short times that the specimens spent at the test temperatures did not modify the microstructures.

Fig. 13. Effect of Neutron Fluence and Test Temperature on Plastic Extensions of Dogbone Specimens at Two Crosshead Speeds.

Fig. 14. Plastic Extensions of the High Fluence Specimens as Functions of Test Temperature and Crosshead Speed.

Fig. 15. Sections Through the Fracture Regions of Dogbone Specimens Showing Lack of Necking in Specimens Irradiated to 8.9×10^{22} neutrons/cm² ($E > 0.1$ MeV) and Tested at Crosshead Speed of 0.0002 in./min. Top, specimens tested at room temperature; left, unirradiated, right, irradiated. Bottom, specimens tested at 200°C; left, unirradiated, right, irradiated. All specimens as-polished.

Fig. 16. Holes Opened at Inclusions Near the Fracture Region of Dogbone Specimen. Unirradiated, tested at 0.0002 in./min. at 25°C.

Fig. 17. Scanning Electron Micrographs of Fracture Surfaces of Specimens Tested at 200°C at a Crosshead Speed of 0.0002 in./min. Top, unirradiated. Bottom, irradiated to 8.9×10^{22} neutrons/cm² ($E > 0.1$ MeV).

POSTIRRADIATION ANNEALING STUDIES

To confirm that the microstructures were stable at the test temperatures and to observe the thermal response of the voids, we did some postirradiation annealing studies. First we subjected an oxide-free section of the fractured tensile specimen TH-134 (8.3×10^{22} fast neutrons/cm²) to a series of cumulative anneals in air for 1-hr periods at progressively higher temperatures, measuring its density after each annealing treatment. The findings are listed in Table 4. These data were only slightly influenced by the small fracture region. Despite the cumulative nature of the anneals, temperatures of 350 to 400°C were required to remove the bulk of the voids. At 500°C shrinkage was reversed and the specimen swelled to more than that shown in its unheated condition. This swelling is probably due to growth of gas bubbles, as observed in other irradiated and annealed aluminum alloys.^{14,17,19} A comparison of the present observations with void annealing data for pure aluminum,^{14,20} 1100 aluminum¹⁹⁻²⁰ and 8001 alloy¹⁷ suggests that voids in the 6061 aluminum alloy disappear more slowly than those in other materials.

Relatively slow void annealing was substantiated in a second annealing experiment. A piece of specimen TH-132 (6×10^{22} fast neutrons/cm²) was thinned and examined by transmission electron microscopy. It was then annealed in air and several selected regions were reexamined to observe any microstructural changes, then annealed further. We found that the precipitate structure remained stable at 200°C, at least for a total time of 1 hr, but coarsened rapidly in 5 min. at 275°C and thereafter coarsened more with further increases in temperature to 300°C. The voids were more stable than the precipitate and dislocation structures. Only the smaller voids disappeared in 1 hr at 200°C. By comparison, the voids in a thinned foil of high purity aluminum, irradiated to give roughly the same concentration and sizes of voids as in the 6061 specimen, all disappeared in 1 hr at 200°C. In the 6061 aluminum foil many of the larger voids were still present after an accumulative anneal including one of 30 min. at 300°C. After this treatment the regions near grain boundaries that were originally free of voids had broadened considerably. No gas bubbles were found on grain boundaries. Thinned foil annealing was discontinued at

Table 4. Effect of Postirradiation Annealing on Swelling in a Portion of Dogbone Specimen TH-134. Cumulative Anneals.

Condition	Density (g/cm ³ at 20°C)	Total ^a Swelling (%)	Swelling from ^b Voids	Percentage of Void Volume Annealed Out
Unheated	2.6776	0.874	0.414	0
1 hr 200°C	2.6812	0.74	0.28	32.4
1 hr 250°C	2.6805	0.765	0.305	26.4
1 hr 300°C	2.6850	0.60	0.14	66.1
1 hr 350°C	2.6881	0.485	0.025	94.0
1 hr 400°C	2.6885	0.47	0.01	97.5
1 hr 500°C	2.6745	0.99 ^c	----	----

^aThe total swelling is $\frac{100(\rho_0 - \rho_{\text{irradiated}})}{\rho_0}$, where ρ_0 is the density of an unirradiated specimen, 2.7012 g/cm³.

^bDefined as total swelling minus that due to the silicon precipitate. The swelling contribution from the silicon is estimated to be 0.46%, which is assumed to remain constant at all annealing temperatures.

^cAt this high annealing temperature gas bubbles are expected to develop and contribute to swelling.

300°C because of interference from the formation of surface oxide films.

DISCUSSION

The observed density changes can be separated into three contributions, one from the oxide films, one from radiation-induced voids, and the third from the transmutation-produced silicon precipitate. Table 2 shows that the oxide films formed in contact with the reactor coolant caused the major part of the overall density change of the target holder. The relative importance of oxide films on the overall density change compared with changes from the other two sources is a function of the component cross section or, rather, its surface-to-volume ratio, assuming that a constant thickness of oxide forms under given conditions of flux and time in the reactor. The results in Table 2 suggest that the oxide thickness must increase with increasing neutron fluence or flux. Some factors that may influence oxide formation are gamma-heating temperature gradients, changes in the degree of radiation-induced water dissociation, and enhanced diffusion due to irradiation-induced defects in the metal and oxide along the length of the holder.

The density changes, or swelling, measured on specimens from which the oxide was removed were caused by the formation of voids and transmutation-produced silicon precipitate. Swelling due to voids alone was measured on several specimens by quantitative electron microscopy and is shown in Table 5. The contribution from silicon is the difference between these values and the measured density changes on the oxide-stripped specimens. Alternatively, we can compute the theoretical contribution from silicon from the known amount of transmutation-produced silicon. At the irradiation temperature of approximately 60°C the silicon is virtually insoluble in aluminum. Assuming that all of the transmutation-produced silicon is precipitated as elemental silicon with a density²¹ of 2.3293 g/cm³, and using the measured density of 2.7012 g/cm³ for the unirradiated 6061 alloy, we estimate that each 1 wt % Si in the form of precipitate occupies 0.0116 net volume fraction and causes the

Table 5. Swelling Contributions from Voids and from Silicon Precipitate
in Irradiated 6061 Alloy Target Holder

Specimen Number	Neutron Fluence ($\times 10^{22}$ neutrons/cm ²)		Measured Swelling from Voids (%)	Estimated Swelling from Silicon Precipitate (%)	Swelling from Voids and Silicon (%)	Measured Density Decrease (%)
	< 0.414 eV	> 0.1 MeV				
TH-130	4.2	2.8	< 0.01	0.15	0.15	0.14
TH-132	9.0	6.0	0.24	0.33	0.57	0.60
TH-136	13.8	9.2	0.68	0.51	1.19	1.06

alloy to swell by 0.16%. Measurements of the production rate of silicon in high purity aluminum in the HFIR²² show that each 1×10^{22} neutrons/cm² ($E < 0.414$ eV) produces ~ 0.23 wt % Si. In excellent agreement with this, chemical analysis of specimen TH-132 (9×10^{22} thermal neutrons/cm²) revealed that irradiation had increased the silicon content by 2.05 wt %. When the swelling contributions from the silicon precipitate (Col. 5, Table 5) and from voids are summed (Col. 6) there is good agreement with the measured density changes (Col. 7) in the oxide-stripped alloy. Over the narrow range of fluences shown in Table 5, the degree of swelling from the silicon precipitate is roughly the same as that from the voids. But the rate of increase of swelling with fluence is faster for voids than for the silicon precipitate.

It is emphasized at this point that these swelling data apply strictly to the particular overaged 6061-T6 alloy irradiated under the thermal to fast flux ratio characteristic of the target region of the HFIR. If this ratio is varied or if the sleeve is in a different condition of heat treatment there may be significant departures from the observed swelling behavior. The effects of microstructure and chemical composition on swelling response seem to be very important. At a constant thermal to fast flux ratio, the swelling in the 6061 alloy is smaller than that measured²³ in any other aluminum alloy subjected to high neutron fluences, as shown in Fig. 18. Note that a fast fluence of only $\sim 5 \times 10^{21}$ neutrons/cm² at approximately 60°C causes more than 2% swelling in high purity aluminum. A fluence of $\sim 3 \times 10^{22}$ neutrons/cm² gives 1.5% swelling in annealed 1100 Al, 3% in cold-drawn 8001 alloy, but only about 0.1% in the hardened 6061 alloy. The 6101 alloy in Fig. 18 is basically similar to 6061 alloy in that it contains alloying additions of magnesium and silicon and can be hardened by Mg₂Si precipitate. But in the annealed 6101 alloy the coherent Mg₂Si precipitate is absent. Silicon is generated in all of these materials by irradiation so the difference in swelling behavior is caused by their relative resistances to void formation. The one feature present in the irradiated 6061 aluminum and absent in the other irradiated materials is the coherent Mg₂Si precipitate, and we believe that the 6061 aluminum owes its

Fig. 18. Irradiation-Induced Swelling in Aluminum and Its Alloys. All materials were irradiated in contact with the cooling water in the HFIR except the 1100 Al which was irradiated in contact with the cooling water in the Oak Ridge Research Reactor. These data include the swelling contribution from the transmutation-produced silicon precipitate.

superior resistance to void formation to the presence of this precipitate. The observations that voids occurred only in those regions free of the fine Mg_2Si precipitate and that swelling was produced much more readily in the annealed 6101 alloy in which there was no fine MgSi precipitate support this view.

The heat treatment given the target holder was chosen to strike a balance between good strength and good ductility with a minimum of warpage from thermal quenching stresses. This treatment resulted in overaging and the formation of precipitate-free zones around inclusions in the microstructure. Precipitate-free zones would presumably be minimized in material in the proper T6 condition and less swelling should occur from voids. Recent studies²⁴ of the 6061-T6 Al HFIR hydraulic tube agree with this conclusion; the microstructure of the tube consists of a uniform distribution of fine Mg_2Si precipitate and is essentially free of voids after irradiation to fast fluences $> 1 \times 10^{23}$ neutrons/cm². However, the fine Mg_2Si precipitate does not alleviate swelling from the transmutation-produced silicon. This silicon will develop in a thermal neutron flux regardless of the heat treatment of the aluminum and at high fluences will cause significant swelling, e.g., 1.1% swelling at 3×10^{23} thermal neutrons/cm².

The effect of the Mg_2Si precipitate on suppressing void formation is in qualitative agreement with a contemporary observation²⁵ that swelling is resisted in irradiated Nimonic PE16 alloy when a high concentration of coherent γ' precipitate particles are present. The mechanism by which coherent particles prevent void formation is not clear. Bullough and Perrin²⁶ have suggested that the coherent interface may trap displaced atoms and vacancies during irradiation and thus promote recombination. On the other hand, the high dislocation density stabilized by coherent precipitate particles may promote recombination by one of the mechanisms proposed by Bullough and Perrin²⁶ and by Harkness and Li.²⁷ There is evidence,¹⁶ too, that the transmutation-produced gases, helium and hydrogen, enhance void nucleation in aluminum, and if these gases are trapped by precipitates their effectiveness as void nuclei might be diminished.

The postirradiation annealing experiments demonstrate that the general microstructure of the irradiated 6061 aluminum was stable at temperatures up to 200°C for test periods of 1 hr or so, but changed during annealing at temperatures of 275°C or more. The irradiation-induced voids annealed out more slowly than those in 1100 grade aluminum or in high purity aluminum. The higher gas content in the relatively heavily irradiated 6061 aluminum on perhaps differences in the interaction of solute atoms and vacancies in the various alloys may affect this annealing rate.

The radiation-induced increases in yield and ultimate stresses can be attributed primarily to the transmutation-produced silicon precipitate and to the increase in dislocation density during irradiation. The increase in yield stress from silicon particles alone, $\Delta\sigma_{Si}$, can be calculated using Ashby's²⁸ modification of the Orowan model of strengthening by hard particles:

$$\Delta\sigma_{Si} = \frac{1.04 Gb}{\pi\ell} \log_e \left(\frac{d}{r_o} \right), \quad (1)$$

where the shear modulus $G = 2.7 \times 10^5$ kg/cm² at room temperature, the Burgers vector $b = 2.86 \times 10^{-8}$ cm, d is the mean particle diameter, the dislocation cut-off radius r_o is usually assumed equal to $4b$, and ℓ is the mean planar particle separation, given by

$$\ell = \frac{d}{2} \left(\frac{2\pi}{3f} \right)^{1/2}, \quad (2)$$

where f is the volume fraction of silicon precipitate. For a specimen irradiated to 9.2×10^{22} neutrons/cm² (> 0.1 MeV) and 1.38×10^{23} neutrons/cm² (< 0.414 eV), f is 0.037 and d is $\sim 1.5 \times 10^{-6}$ cm. Consequently, $\ell = 5.7 \times 10^{-6}$ cm and from Eq. (1), $\Delta\sigma_{Si} = 16,500$ psi. The yield stress increase from dislocations alone follows the relation²⁹

$$\Delta\sigma_d = \alpha Gb (\lambda)^{1/2}, \quad (3)$$

where α is about 1.0 and λ is the dislocation density. For $\lambda = 2.5 \times 10^{10}$ /cm², $\Delta\sigma_d$ is 17,400 psi. Voids also contribute to hardening:³⁰

$$\Delta\sigma_v = 2 \frac{Gb}{L} \quad (4)$$

where $L = (Nd)^{-1/2}$ and N is the concentration of voids. Consequently, for $N = 1.92 \times 10^{14}/\text{cm}^3$ and $d = 3.5 \times 10^{-6}$ cm, $\Delta\sigma_v = 5,700$ psi.

These calculations show the relative strengthening effects of the various irradiation-induced defects. These individual contributions must be superimposed to give the overall hardening increment due to the irradiation. However, there is no generally accepted method of combining the individual contributions since much depends on whether the particular obstacles responsible for the hardening are weak or strong and of high or low concentrations. The practical limits of the possible combinations may be defined on the one hand by a simple summation of the strength contributions, and on the other by the square root of the sum of the squares of the individual components.³¹ Real cases probably lie between these limits. The situation in the 6061 aluminum alloy is further complicated by the original large concentrations of Mg_2Si particles and dislocations that hinder dislocation motion. We estimate the strengthening contribution from these original obstacles as follows: The yield stress of unirradiated 6061 aluminum alloy, σ_u , at room temperature is about 30,000 psi. Annealed 1100 grade aluminum which has similar solid solution alloying elements but contains neither the Mg_2Si precipitate nor the many dislocations has a yield stress, σ_o , of 5,000 psi.¹ Thus the yield stress of the unirradiated alloy is given by

$$\sigma_u = \sigma_o + \Delta\sigma(\text{Mg}_2\text{Si} + d), \quad (5)$$

where the contribution of Mg_2Si precipitates and dislocations, $\Delta\sigma(\text{Mg}_2\text{Si} + d)$, is about 25,000 psi. The yield stress of the irradiated material, σ_i , lies somewhere between

$$\sigma_i = \sigma_o + \Delta\sigma(\text{Mg}_2\text{Si} + d) + \Delta\sigma_{\text{Si}} + \Delta\sigma_d + \Delta\sigma_v \quad (6)$$

and

$$\sigma_i = \sigma_o + \left\{ [\Delta\sigma(\text{Mg}_2\text{Si} + d)]^2 + (\Delta\sigma_{\text{Si}})^2 + (\Delta\sigma_d)^2 + (\Delta\sigma_v)^2 \right\}^{1/2}. \quad (7)$$

Predicted values of room temperature yield stress derived from Eqs. (6) and (7) using our microstructural measurements are listed in Table 6 as functions of neutron fluence. The measured values of σ_i fall between the two calculated values.

Table 6. Comparison of the Measured Room Temperature Yield Strengths with the Calculated Values Derived from Eqs. (6) and (7)
Using the Microstructural Parameters from Table 3

Neutron Fluence ($\times 10^{22}$ n/cm ²) (E > 0.1 MeV)	Measured* Yield Stress ($\times 1000$ psi)	Calculated Yield Stress ($\times 1000$ psi)	
		Eq. (6)	Eq. (7)
0	30.0	30.0	30.0
2.8	43.0	58.5	36.8
6.0	48.0	67.0	39.0
9.2	51.0	69.6	40.0

*Measured yield stresses are taken from the smooth-line fit of the experimental data in Fig. 10.

From these calculations we conclude that the major sources of irradiation-induced strengthening in the 6061 aluminum are the dislocation structure and the transmutation-produced silicon precipitate. The effect of voids is relatively small. There was no observable change in the original Mg₂Si precipitate that might have influenced the strength. The gases generated from (n, α) and (n,p) reactions, estimated to be 40 at. ppm helium (we measured 41 at. ppm) and 270 at. ppm hydrogen for the highest fluence, did not form visible bubbles at low temperatures. If such gases were present as submicroscopic bubbles < 20 Å diam their hardening contribution would be negligible, according to Kroupa and Hirsch.³² The decrease in the rate of strengthening at fast fluences above about 3×10^{22} neutrons/cm² (Fig. 10) is also attributed to the silicon precipitate and the dislocation structure whose concentrations appear to saturate above this threshold fluence. At high fluences the

increase of strength with increasing fluence depends only on the voids and the increasing size of the silicon particles, and thus the strength is less sensitive to neutron fluence than it is at lower fluences where the concentrations of dislocations and silicon particles are below saturation.

One reason for the decrease in mechanical strength with increasing test temperature is common to irradiated and unirradiated material: the elastic modulus of the aluminum matrix decreases with increasing temperature. The total strength decrease up to 200°C is greater in the irradiated alloy but no microstructural changes that could account for this difference were found.

Absolute ductility changes are difficult to evaluate for the zero-gage length dogbone specimens. Plastic deformation occurs over increasingly larger volumes of the specimen as work-hardening occurs and stresses increase during tensile testing. Moreover, the extent of this effect may change with irradiation and with test temperature and strain rate. Despite these considerations, however, the general trends of the apparent ductility changes are similar to those observed in irradiated 8001 aluminum¹⁷ and in irradiated 1100 aluminum.^{33,34} In all cases there is considerable ductility loss at low test temperatures and maximum loss at temperatures between ~ 100 and ~ 300°C. The ductility loss in 1100 aluminum³³ has been analyzed most thoroughly and most of the loss in this temperature range results from irradiation-induced hardening. The defect structure makes dislocation motion more difficult and the work-hardening coefficient n from the equation $\sigma = k\epsilon^n$, where σ and ϵ are true stress and true strain, is severely reduced. Since significant defect hardening occurs in the 6061 aluminum, this presumably causes at least some of the observed ductility loss and is probably responsible for the reduction in the plastic extension to the ultimate stress.

Irradiated 1100 aluminum specimens usually fail by ductile tearing at low test temperatures but at 200°C fail intergranularly in slow tensile tests.³³ Intergranular fracture is frequently stimulated by the presence of helium³²⁻³⁵ and is enhanced by slow strain rates. However, all irradiated and unirradiated 6061 aluminum dogbone specimens appeared to fail by tearing under all test conditions even at 200°C under the

slowest crosshead speed where the ductility of irradiated specimens reached a minimum. Tearing involves dimple formation during fracture and is characteristic of a rapid strain rate process where high local strains open up holes at obstacles such as inclusions. There were many such obstacles in the dogbone 6061 aluminum specimens, most of them surrounded by zones free of fine Mg_2Si precipitate. The local strain rates in plastically deforming regions were high because of the short gage length of the specimens. Test times ranged from less than 1 min. to 4 hr. These are fast tests in relation to the creep-type tests under which intergranular fracture is predominant.

We picture the fracture process as one involving tearing or dimple formation in the Mg_2Si precipitate-free zones around inclusions and at the large Mg_2Si rods in both the irradiated and the unirradiated specimens. Irradiation modifies the fracture process to the extent that it hardens the matrix and the regions around inclusions and large Mg_2Si particles, and introduces clusters of voids around the particles. During testing these voids may serve as pre-existing ductile dimple embryos for local fractures that propagate through the hardened matrix at lower strains than those in the unirradiated material. The role of gases, other than to stabilize voids, remains obscure.

CONCLUSIONS

Age-hardened 6061 aluminum containing inclusions and a coarse and fine precipitate of Mg_2Si shows good resistance to void formation, much better than 1100 aluminum and 8001 alloy. Voids are not detected in the 6061 aluminum at fluences below 2.8×10^{22} neutrons/cm² (> 0.1 MeV) and account for about 0.7% swelling after exposure to 9.2×10^{22} neutrons/cm² (> 0.1 MeV). Additional swelling arises from the formation of a precipitate of transmutation-produced silicon and from the growth of a surface oxide film. Resistance to swelling from voids can be improved by heat treating to ensure that the 6061 aluminum contains a uniform distribution of fine coherent Mg_2Si precipitates before irradiation.

Radiation-induced strengthening in precipitation-hardened 6061 aluminum is caused by the transmutation-produced silicon precipitate, by

an increase in the concentration of dislocations, and by voids. Irradiation also reduces the ductility of the alloy, the loss being most severe at a test temperature of 200°C and a relatively slow testing speed. The fracture process remains transgranular and is characterized by dimpled fracture surfaces.

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APPENDIX 1

High Temperature Tests

In anticipation of the question "What would happen to the mechanical properties of the 6061 aluminum target sleeve if it were subjected to temperatures higher than those that might normally be experienced in the HFIR?", we tested some specimens at temperatures between 200 and 500°C. Since at these temperatures the Mg_2Si precipitate is unstable and much of the irradiation damage is annealed, the microstructures of both irradiated and unirradiated specimens differ from those at test temperatures below 200°C. The mechanical properties will change accordingly.

The effect of these high temperatures on the strength properties is shown in Fig. 19. The irradiated specimens remained stronger than the unirradiated ones at all test temperatures, no doubt due to the transmutation-produced silicon. Irradiated specimens tested at the higher crosshead speed were the stronger. This effect of crosshead speed is probably connected with the time the specimens spent at the test temperature, the longer time permitting greater change of the original microstructure and also allowing recovery of plastic deformation induced during the test. Beyond temperatures of 400°C the 6061 alloy in both irradiated and unirradiated conditions has very low strength at the lower crosshead speed.

The ductility of the specimens (Fig. 20) was a minimum at 200°C above which it increased with increasing test temperature, but irradiated specimens remained consistently less ductile than the unirradiated specimens. A metallographic examination of an irradiated specimen fractured at 500°C showed extensive internal tearing and cracking (Fig. 21) and scanning electron microscopy of the fracture surfaces indicated failure by tearing (Fig. 22).

These data suggest that embrittlement of the irradiated 6061 aluminum alloy is likely to be a maximum at temperatures in the region of 200°C under slow strain rate conditions. At higher temperatures much structural strength is lost but a margin of safety from sudden catastrophic failure is promised by a concomitant increase in ductility.

Fig. 19. Tensile Strengths of Dogbone Specimens Tested at Elevated Temperatures.

Fig. 20. Ductility of Dogbone Specimens Tested at Elevated Temperatures.

Fig. 21. Fracture in Specimen Irradiated to 8.3×10^{22} neutrons/cm² ($E > 0.1$ MeV) and Tested at 500°C at a Crosshead Speed of 0.0002 in./min.

Fig. 22. Scanning Electron Micrographs of Fracture Surfaces of Specimens Tested at 500°C at a Crosshead Speed of 0.0002 in./min. Top, unirradiated. Bottom, irradiated to 8.3×10^{22} neutrons/cm² ($E > 0.1$ MeV).

APPENDIX 2

Examination of the Weld Seam

Only a limited number of specimens contained the weld seam. This seam lay in the direction of tensile testing and the narrow sections of the specimens were comprised entirely of weld metal, the original welds being 0.20 to 0.30-in. wide while the specimens were only 0.15-in. wide at their narrowest sections. These welds were made before the sleeve was heat treated. A section through the base metal-weld metal interface after heat treatment is shown in Fig. 23. There are large pores at the weld metal-base metal interface.

The main purposes of testing these specimens were to determine differences in properties between the base and weld metals and to assess whether the weldment was safe for use in-reactor after exposure to high neutron fluences. Therefore, the specimens that had received the highest fluences were tested under the conditions most likely to occur during normal reactor operation, that is, at test temperatures below 200°C.

The strength and ductility of the weld seam are plotted as functions of neutron fluence and test temperature in Figs. 24, 25, and 26. The curves in these figures represent the data for the 6061 aluminum-base metal; tests on the weld metal are shown as individual points with vertical lines connecting them to the appropriate base metal curve. Within the experimental scatter the strength properties of irradiated weld material do not differ markedly from those of the irradiated base material. But at the low crosshead speed the ductility of the weld specimens was only about half that of the base material.

Unirradiated weld sections were not tested. Moreover, because of the method of cutting samples from the sleeve, the base metal-weld metal interface with its apparent porosity (Fig. 23) was also not tested. This is unfortunate, particularly since these tests provide the only publically known information on an aluminum weldment exposed to very high fast neutron fluences. However, the weld did not fail in service, and the results of our examination, though meager, do not discourage in-reactor use of this type of weldment. But the poor ductility should not be overlooked.

Fig. 23. Photomicrograph of a Polished but Unetched Section of Base Metal-Weld Metal Interface in the Unirradiated Target Holder.

Fig. 24. Effect of Neutron Fluence and Test Temperature on the Yield Stress of Specimens Containing Weldment. Curves are for 6061 alloy base metal from Fig. 10.

Fig. 25. Effect of Neutron Fluence and Test Temperature on the Ultimate Stress of Specimens Containing Weldment. Curves are for 6061 alloy base metal from Fig. 11.

ORNL - DWG 70-15323A

Fig. 26. Ductility of Specimens Containing Weldments. Curves are for 6061 alloy base metal from Fig. 13.

APPENDIX 3

Table 7. Specimen Identification, Test Conditions and Tensile Data for Dogbone-Shaped Specimens of 6061 Aluminum from HFIR Target Holders. Specimens were tested in air at the indicated crosshead speeds in an Instron tensile machine

CHS = 0.0002 in./min.

Specimen Number	Distance Above (A) or Below (B) Horizontal Center Line (in.)	Neutron Fluence ($\times 10^{22}$ neutrons/cm ²)			Test Temp. (°C)	Yield Stress (psi)	Ultimate Stress (psi)	Plastic Extension to Ultimate (in.)	Plastic Extension to Failure (in.)
		(E<0.414 eV)	(E>0.1 MeV)	(E>0.82 MeV)					
69-10	--	0	0	0	RT	28,100	35,700	0.0105	0.0333
69-19	--	0	0	0	50	27,700	35,700	0.009	0.033
69-14	--	0	0	0	100	30,200	34,000	0.0107	0.040 ₃
69-2	--	0	0	0	150	27,400	29,400	0.0081	0.0440
69-6	--	0	0	0	200	23,200	25,300	0.0053	0.0441
69-9	--	0	0	0	300	8,006	8,108	0.0027	0.0578
69-16	--	0	0	0	400	1,840	2,056	0.0066	0.140
69-11	--	0	0	0	500	440	606	0.0109	0.220
30	10.4A	4.3	2.84	1.7	RT	42,000	48,800	0.0066	0.0192
9	5.6B	11.0	7.3	4.3	RT	45,200	53,700	0.0068	0.0186
19	2.4A	13.3	8.9	5.24	RT	50,245	56,257	0.00695	0.0171
8	7.2B	9.0	6.0	3.5	50	41,900	47,900	0.0053	0.0185
16	0.8B	13.8	9.2	5.4	50	39,900	49,900	0.0053	0.0156
28	8.8A	6.7	4.5	2.65	100	38,200	48,300	0.005	0.015
20	2.4A	13.3	8.9	5.24	100	33,100	40,700	0.0046	0.0137
43	4.0A	12.5	8.3	4.9	125	40,800	46,200	0.0043	0.0120
42	2.4A	13.3	8.9	5.24	125	45,000	48,100	0.0063	0.0128
26	7.2A	9.0	6.0	3.5	150	33,600	48,000	0.00475	0.0090
12	4.0B	12.5	8.3	4.9	150	32,300	35,800	0.00275	0.0045
41	0.8B	13.8	9.2	5.4	150	28,400	40,200	0.0055	0.00830
24	5.6A	11.0	7.3	4.3	200	21,400	27,600	0.0048	0.0138
14	2.4B	13.3	8.9	5.24	200	22,200	25,300	0.0008	0.0022
39	2.4A	13.3	8.9	5.24	200	32,900	33,600	0.00055	0.00095
55	2.4A	13.3	8.9	5.24	300	9,690	10,900	0.0017	0.0164

Table 7 (continued)--

Specimen Number	Distance Above (A) or Below (B) Horizontal Center Line (in.)	Neutron Fluence ($\times 10^{22}$ neutrons/cm ²)			Test Temp. (°C)	Yield Stress (psi)	Ultimate Stress (psi)	Plastic Extension to Ultimate (in.)	Plastic Extension to Failure (in.)
		(E<0.414 eV)	(E>0.1 MeV)	(E>0.82 MeV)					
56	0.8A	13.8	9.2	5.4	400	9,270	10,920	0.0043	0.0224
54	4.0A	12.5	8.3	4.9	500	412	618	0.0080	0.0600
61-40	0	13.8	9.2	5.4	RT	41,200	50,200	0.0093	0.0106 Weld
61-41	1.6A	13.5	9.0	5.3	50	41,600	50,600	0.0053	0.0101 Weld
61-39	1.6B	13.5	9.0	5.3	100	38,500	44,500	0.0046	0.0110 Weld
61-35	8.0B	8.0	5.3	3.1	150	31,100	35,700	0.0032	0.0117 Weld
61-38	3.2B	12.9	8.6	5.1	150	27,800	39,100	0.0026	0.0112 Weld
61-48	12.8B	?	0.6	0.37	200	26,698	32,365	0.0028	0.0128 Weld

CHS = 0.002 in./min.

69-3	--	0	0	0	RT	28,100	35,700	0.013	0.035
69-13	--	0	0	0	50	27,700	35,700	0.014	0.036
69-23	--	0	0	0	100	30,200	34,000	0.010	0.037
69-21	--	0	0	0	150	27,400	29,400	0.0085	0.041
69-1	--	0	0	0	200	23,200	25,300	0.0075	0.050
61-74	2.4B	13.3	8.9	5.24	100	39,400	44,700	0.0029	0.013 ₁
61-73	0.8B	13.8	9.2	5.4	150	37,300	42,800	0.0018 ₅	0.005
61-72	0.8A	13.8	9.2	5.4	200	34,300	38,000	0.004 ₆	0.0138

TH-144	12.0B	?	1.2	0.7	} Untested
TH-130	10.4A	4.2	2.8	1.7	
TH-142	8.8B	6.7	4.5	2.65	
TH-132	7.2A	9.0	6.0	3.54	
TH-140	5.6B	10.9	7.3	4.3	
TH-134	4.0A	12.5	8.3	4.86	
TH-138	2.4B	13.3	8.9	5.24	
TH-136	0.8A	13.8	9.2	5.4	

Table 7 (continued)--
CSH = 0.2 in./min.

Specimen Number	Distance Above (A) or Below (B) Horizontal Center Line (in.)	Neutron Fluence ($\times 10^{22}$ neutrons/cm ²)			Test Temp. (°C)	Yield Stress (psi)	Ultimate Stress (psi)	Plastic Extension	Plastic Extension
		(E<0.414 eV)	(E>0.1 MeV)	(E>0.82 MeV)				to Ultimate (in.)	to Failure (in.)
69-4	--	0	0	0	RT	31,900	37,100	0.013	0.0355
69-5	--	0	0	0	50	28,200	33,800	0.0125	0.0375
69-20	--	0	0	0	100	27,500	33,100	0.0120	0.0280
69-7	--	0	0	0	150	27,800	33,100	0.0085	0.0345
69-22	--	0	0	0	200	26,000	29,000	0.009	0.040
1	12.0B	?	1.2	0.7	RT	48,400	57,600	0.009 ₀	0.021 ₀
29	10.4A	4.3	2.8	1.7	RT	41,100	45,000	0.009 ₅	0.019 ₁
18	0.8A	13.8	9.2	5.4	RT	54,100	54,500	0.010 ₀	0.019 ₀
31	12.0B	?	1.2	0.7	50	44,800	48,600	0.009 ₀	0.018 ₅
5	8.8B	6.7	4.5	2.65	50	47,400	52,200	0.009 ₀	0.020
15	0.8B	13.8	9.2	5.4	50	47,200	54,000	0.009 ₅	0.016
22	4.0A	12.5	8.3	4.9	100	37,900	43,000	0.009 ₅	0.019 ₃
13	2.4B	13.3	8.9	5.24	100	44,500	47,100	0.008 ₅	0.017 ₅
27	8.8A	6.7	4.5	2.65	150	39,500	42,300	0.006	0.020
11	4.0B	12.5	8.3	4.9	150	37,200	43,500	0.011	0.024
23	5.6A	11.0	7.3	4.3	200	34,700	36,700	0.005	0.0165
21	4.0A	12.5	8.3	4.9	200	39,000	40,700	0.006	0.0135
59	4.0B	12.5	8.3	4.9	300	23,400	26,200	0.003	0.0020
58	2.4B	13.3	8.9	5.24	400	11,500	12,900	0.004	0.038
57	0.8B	13.8	9.2	5.4	500	12,100	13,600	0.0015	0.029
60	5.6B	7.3	7.3	4.3	500	9,800	11,000	0.002	0.046
61-45	8.0A	8.0	5.3	3.1	RT	46,900	56,000	0.0085	0.019 ₀ Weld
61-34	9.6B	5.5	3.7	2.2	50	38,500	46,800	0.008 ₀	0.020 Weld
61-47	11.2A	?	2.0	1.2	100	31,400	44,300	0.006	0.023 Weld
61-43	4.8A	11.7	7.8	4.6	150	37,900	44,400	0.004 ₅	0.013 Weld

CHS is the Instron Crosshead Speed.

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