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DAMAGE IN HIGHLY IRRADIATED 1100 ALUMINUM

By

J.M. Beeston, G.E. Korth, M.R. Martin, C.R. Brinkman

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ABSTRACT

Aluminum components in the ATR receive high irradiation fluences in relatively short times, so that material with high fluence is of interest in the evaluation of service life of reactor components.

An 1100 aluminum skirt plate which had been removed from the MTR after 18 years accumulation at 70°C of a peak fast fluence of 5.6×10^{22} n/cm² E > 1 MeV and thermal fluence of 6.5×10^{22} n/cm² was cut into specimens; these were tested to define the growth, mechanical properties, and compositional and structural changes.

The aluminum had retained sufficient ductility so that it was still serviceable. This report describes the tests and the changes observed.

SUMMARY

At the highest irradiation fast fluence of 5.6×10^{22} n/cm² $E > 1$ MeV a density decrease of 7.7% occurred in the 1100 aluminum MTR skirt plate. The density change could be represented by an equation $-\frac{\Delta\rho}{\rho} = a(\phi t)^3$. The growth was shown to be principally due to the formation of voids from the fast fluence, although the difference in density between the aluminum and the principal transmutation product, silicon, contributed about 10% of the change.

The yield and ultimate strengths of the as-irradiated aluminum increased 300 to 600% with the highest strengths occurring at intermediate fluences of 1.2 to 1.6×10^{22} n/cm² $E > 1$ MeV. The increase in strengths over nonirradiated aluminum standard data was retained after annealing for three hours at temperatures of 100, 200, and 300°C and testing at these temperatures. At 300°C the recovery of a principal hardening agent appeared to occur although the void density was little changed. The principal hardening agents are believed to be irradiation-produced silicon and point defects.

A reduction in ductility also occurred, although enough ductility was retained in the as-irradiated tests to consider the material still serviceable. A local ductility of greater than 45% reduction in area was retained. The least local ductility occurred in the high fluence specimens after the 300°C anneal and tests.

A comparison of calculated and measured transmutation products of silicon, helium, and hydrogen showed reasonable agreement except for hydrogen wherein the difference is thought to be due to pickup from in-reactor corrosion reactions and from proton recoil.

Transmission electron microscopy revealed that the void density at fast fluences less than 2×10^{22} n/cm² was 160 times less than at fluences of about 4×10^{22} n/cm². At ambient temperatures and 100°C the silicon is distributed so as to give no diffraction pattern. After a 300°C anneal for three hours the silicon particulate coarsened. An electron diffraction ring pattern gave positive evidence that the coarsening particulate was silicon. Deformation appeared to produce increased void formation especially in local regions, but the silicon distribution was not greatly affected. The density of the voids in high fluence specimens ($\approx 4 \times 10^{22}$ n/cm² $E > 1$ MeV) was not greatly affected by annealing at 300°C for three hours. A void-depleted region adjacent to the grain boundary was observed before and after annealing.

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DAMAGE IN HIGHLY IRRADIATED 1100 ALUMINUM

I. INTRODUCTION

Aluminum components in the newer test and high-flux reactors receive high irradiation fluences relatively fast in the service life of the reactor. Not much data are available at high fluences on which to base service life predictions for these components. Failures due to irradiation damage in aluminum have been noted^[1] in HFIR target rods. Data on 1100 aluminum to fluences of 1.2×10^{22} n/cm² (> 1 MeV) have been reported^[2] and some data for a 6061 aluminum target basket have recently become available^[3]. In the ATR the 6061 aluminum neck shim housing will reach fluences at which measurable growth occurs (about 10^{22} n/cm²) in one full power year. An 1100 aluminum skirt plate from the MTR had been removed after about 18 years irradiation next to the core at 70°C. The tests made from specimens cut from this piece to define growth, mechanical properties, and compositional and structural changes are the subject of this report.

II. MATERIAL AND IRRADIATION HISTORY

The MTR 1100 aluminum west skirt plate irradiated at 70°C (Figure 1) was a rectangular piece $14\text{-}3/4 \times 16$ inches which formed the interface between the fuel elements and a horizontal beam hole, HT-1, and received an exposure of 177,187 MWd. It was removed at MTR Cycle 295 and subsequently cut into specimens for density and mechanical property measurements. The top of the skirt plate was located five inches below the reactor center line, and position flux maps from fuel and experimental holes were used to establish a flux grid across the face of the plate. The fast flux was estimated to be accurate within $\pm 20\%$; however, the thermal flux is not considered to be as accurate because of uncertainty in positions of flux peaking. An effective time was calculated for total operation from which the fluence values were computed. The thickness of the plate varied because of the convex surfaces at the interface with the sides of the fuel elements and because of the relief ribs for mounting, but was such as to accommodate specimen blanks $3/8 \times 1/4$ inch by four inches for mechanical properties and $1/4 \times 1/5 \times 1$ inch for the density specimens. The flux grid was

A. Fuel Element Side

B. Back Side

Fig. 1 Photograph of MTR skirt plate from which specimens were machined.

overlaid on the layout of the specimen cutout drawing to give individual specimen fluences.

III. DENSITY MEASUREMENTS

The growth of the aluminum was determined through immersion density measurements on eight specimens from various sections of the plate. The specimens were prepared by polishing all six surfaces on 600 grit paper to remove oxide and machining burrs. The 1MHD specimen (Table I) was further stripped in a solution of phosphoric and chromic acid followed by hot nitric acid. The immersion liquid was water at 23.5°C with photoflo as a wetting agent. A precision of $\pm 0.002 \text{ g/cm}^3$ is estimated. Control material for testing was not available so that the Aluminum Association nominal density for 1100 Al of 2.71 g/cm^3 was taken for the calculation of the density change or expansion.

TABLE I
DENSITY CHANGES IN IRRADIATED 1100 Al

Specimen	$\phi \text{ fast n/cm}^2$ $E > 1 \text{ MeV}$ $\times 10^{22}$	$\phi \text{ th n/cm}^2$ $\times 10^{22}$	Density g/cm^3	Density Change $(\frac{-\Delta\rho}{\rho})\%$
1MHD	5.6	6.5	2.501	7.7
D-1	4.1	5.0	2.5967	4.2
D-3	4.0	5.4	2.6088	3.7
D-5	2.8	5.1	2.6680	1.6
D-2	1.8	3.4	2.7039	0.2
D-4	1.8	3.9	2.7091	0.04
D-6	1.7	3.7	2.6977	0.4
D-7	0.9	2.6	2.7081	0.07

The density change is plotted in Figure 2 wherein the data are very well represented, except for scatter in the low density change values, by an equation of the form $-\frac{\Delta\rho}{\rho} = a (\phi t)^n$ where n is the growth exponent and is equal to 3. The fast fluence values used in all the figures are $E > 1 \text{ MeV}$.

Fig. 2 Density change in irradiated 1100 Al skirt plate specimens compared with 6061 Al.

IV. MECHANICAL PROPERTY CHANGES

Twenty-five specimen blanks for tensile testing were cut from the plate $3/8 \times 1/4$ inch by four inches. These blanks were finished into specimens on a Tensil-Kut machine by shaping a rectangular reduced width of about $1/4$ inch with a length of $1-1/2$ inches. The surface was finished on a milling machine such that the gage cross section for most specimens was approximately 0.05 in.². The uniformity over the gage section is estimated at better than ± 0.0004 inch for the width and thickness. Since control material was not available, comparison properties were taken for 1100 Al alloy from the "Aluminum Standards and Data", and are given in Table II. The data for elongation is for a 2 inch gage length. It is recognized [4] that specimen geometry has an effect on elongation, so that the equivalent ASTM round specimen to which a comparison can be made is estimated by using the concept that a constant elongation is obtained for a constant ratio of $\sqrt{\frac{A}{\ell}}$, where A is the cross section area and ℓ the gage length. Whereupon one finds the equivalent ASTM round specimen to be 0.25 inch diameter with a 1 inch gage length. The standard ASTM specimen is 0.505 inch diameter and for this reason the unirradiated ductility values are not plotted in the figures, and a qualitative comparison of changes in ductility are given.

The specimens were annealed for three hours at the temperature at which they were to be tested. Testing was done on an Instron machine at a cross head speed of 0.02 inch per minute with cross head readout. Difficulty was encountered on some specimens with grip slippage; the data from these specimens were discarded. The data are given in Table II.

The room temperature irradiated strength is seen to increase (Figure 3) to a maximum at a fluence of 1.2 to 1.6×10^{22} n/cm² E > 1 MeV; it then decreased. This decrease also occurs at 100 and 200°C but at 300°C it is obscured by data scatter (Figures 4, 5, and 6). The fracture strength also decreases at the higher fluence from 40 to 30 ksi. The decreased strength is due to void content.

The increase in strength from the unirradiated value is retained after the three-hour anneal at 100, 200, and 300°C. However, at about 300°C the recovery of a major hardening agent takes place as seen in Figure 7, wherein the general reduction in strength that occurs with increase in temperature is further accentuated at this temperature so that the shape of the curve as a function of fluence is changed. The hardening agent is unlikely to be dissolved silicon

TABLE II

MECHANICAL PROPERTIES OF IRRADIATED 1100 Al MTR SKIRT PLATE

Specimen No.	ϕ Fast E > 1 MeV $\times 10^{22}$	ϕ th $\times 10^{22}$	Test Temperature (°C)	U.T.S. (ksi)	Y.S. (ksi)	Uniform El. %	R.A.%	At Failure El. %	Cross Section Area, A_o , in. ² $\times 10^{-2}$	Fracture Stress (ksi)	Zero-Gage Length Ductility ^[c] e_o
A-1	4.5	5.6	Amb.	23.0	18.7	7.0	45.7	24	5.87	29.4	0.84
A-2	4.5	5.6	Amb.	19.5	13.8	9.5	53.2	16	5.42	31.2	1.14
B-2	2.7	4.1	Amb.	25.3	19.1	6.5	49.5	18	5.05	37.7	0.98
C-1	1.1	4.6	Amb.	32.8	28.9	8.0	48.5	25	5.40	40.5	0.94
C-2	1.0	4.2	Amb.	31.2	27.5	7.1	46.4	[d]	5.32	39.5	0.87
C-9	1.5	3.0	Amb.	33.6	28.5	9.0	55.9	21	5.08	48.1	1.26
Al Stn'd [a]	---	---	20	13.0	5.0	---	---	40 [b]			
A-3	4.3	6.3	100	25.4	22.9	9.0	41.7	13	3.51	33.5	0.72
A-4	4.3	6.3	100	24.1	23.1	2.7	41.1	14	3.56	31.0	0.70
B-3	2.7	4.1	100	24.6	22.0	5.5	53.5	15	5.63	36.7	1.16
B-4	1.9	4.7	100	31.3	28.0	6.5	54.4	23	5.08	45.3	1.20
C-3	1.1	4.6	100	27.8	25.7	5.5	44.8	12	5.38	31.6	0.81
C-4	0.7	3.2	100	30.3	26.9	6.6	46.0	18	5.02	36.5	0.85
Al Stn'd [a]	---	---	100	10.0	4.6	---	---	45 [b]			
A-6	3.9	3.8	200	12.5	10.5	9.6	52.5	[d]	3.55	16.8	1.10
B-5	1.9	4.7	200	30.4	26.6	8.5	52.1	8	5.40	38.7	1.09
C-5	0.7	3.2	200	21.8	20.2	11.9	52.3	[d]	5.03	28.1	1.10
C-6	0.38	2.3	200	30.7	23.0	12.6	48.2	18	5.15	---	0.93
Al Stn'd [a]	---	---	200	6	3.5	---	---	65 [b]			
A-7	3.9	3.8	300	14.0	13.3	3.6	32.6	13	3.61	---	0.48
A-8	3.9	3.8	300	6.1	5.6	5.0	27.3	[d]	4.11	---	0.38
B-7	1.7	2.1	300	11.2	9.5	6.5	66.4	[d]	5.05	---	1.98
B-8	1.7	2.1	300	13.0	12.5	1.4	72.3	[d]	5.26	---	2.60
Al Stn'd [a]	---	---	300	3.2	2.2	---	---	79 [b]			

[a] Values from Aluminum Standards and Data, 2nd Edition, December 1969, The Aluminum Association, N.Y., N.Y.

[b] Elongation in 2 inch gage length.

[c] Zero-gage-length ductility, $e_o = \frac{q}{1-q}$, where q = Reduction-in-Area.

[d] Specimen broke outside gage marks.

Fig. 3. Room temperature tensile properties of irradiated 1100 Al.

Fig. 4 Tensile properties after anneal and 100°C tests.

Fig. 5 Tensile properties after anneal and 200°C tests.

Fig. 6 Tensile properties after anneal and 300°C tests.

since the solubility at these temperatures is likely to be less than 0.06 wt% and the transmutation-produced silicon at the peak strengths is about ten times greater. The principal hardening agent is also unlikely to be voids, since as will be shown later in the high fluence specimens at 300°C the density of the voids appears little changed. It is thought that the irradiation-produced silicon and point defects are the principal hardening agents and annealing of these produce the differences in etching behavior illustrated in Figures 8 and 9. The irradiation-produced silicon distributed uniformly is believed to be a principal defect which influences the strength and ductility and is agglomerated at 300°C as will be shown later by electron diffraction.

Fig. 7 Ultimate strength of irradiated 1100 Al after annealing and testing at temperatures indicated.

The reduction in ductility that normally occurs with an increase in strength also occurs in 1100 aluminum upon irradiation (Figures 3 through 6) as measured by uniform and total elongation to failure. However, as measured by reduction in area, an increase to a maximum and then a reduction occurs in the room temperature and 100°C tests (Figures 3 and 4). The reduction in area was measured at the minimum cross section (Figure 10) by means of a Filer eyepiece. The measurements were made by direct reading of the width and thickness (rectangular section) of the two specimen halves, and the reduction in area was computed from the average of the two measurements. The measurement of ductility by elongation is recognized as composed of a local necking extension and a uniform extension[4]. The local necking extension is a function of the cross sectional area as is the reduction-in-area. The uniform extension is a primary function of the metallurgical condition, while the local necking is primarily a function of the state of stress (specimen size and shape). To reduce the contribution from the neck, a zero-gage-length elongation, $e_o = q/(1-q)$ where q = the reduction-in-area can be calculated as given in Table II. It is seen that at intermediate

(a) Spec. A-2, $\phi = 4.5 \times 10^{22}$ 84X

(b) A-2 Room Temperature 500X

(c) Spec. A-4, $\phi = 4.3 \times 10^{22}$ 84X

(d) A-4 100°C 500X

Fig. 8 Longitudinal sections showing islands of deformed regions after room temperature (a) and (b), and after 100°C (c) and (d) tests.

(a) Spec. A-6, $\Phi = 3.9 \times 10^{22}$ 84X

(b) A-6 200°C 500X

(c) Spec. A-8, $\Phi = 3.9 \times 10^{22}$ 84X

(d) A-8 300°C 500X

Fig. 9 Longitudinal sections showing apparent increase in coalescing cavity size with increase in temperature (a) and (b) at 200°C and (c) and (d) at 300°C.

A-1 (4.5×10^{22}) RmT., %R.A. = 46

C-1 (1.1×10^{22}) RmT., %R.A. = 49

A-4 (4.3×10^{22}), 100°C, %R.A. = 41

C-3 (1.1×10^{22}), 100°C, %R.A. = 45

A-6 (3.9×10^{22}), 200°C, %R.A. = 53

C-6 (0.38×10^{22}), 200°C, %R.A. = 48

A-7 (3.9×10^{22}), 300°C, %R.A. = 33

B-7 (1.7×10^{22}), 300°C, %R.A. = 66

Fig. 10 Fracture surfaces generally show greater local necking at lower fluence.

fluence ($1.5 - 2.7 \times 10^{22}$ n/cm²) the zero-gage-length ductility is generally greater than one while at higher or lower fluences it is generally less than one (most of the exceptions occur at 200°C where the defect-silicon is expected to be starting to agglomerate). It will be seen in the section on microscopy that the density of the voids has not been affected greatly by the three hour anneal and testing temperature, although the defects influencing the etching behavior are affected (Figures 8 and 9). It is thus thought that the maximum in reduction-in-area is due principally to a critical distribution of the defect-silicon and not to the specimen size and shape.

The hardness decrease and density of the highest fluence specimens (5.6×10^{22} n/cm² $E > 1$ MeV) were investigated by annealing samples for the times and temperatures indicated in Table III.

TABLE III
HARDNESS AND DENSITY CHANGES AFTER ANNEALING

Condition	Density Decrease (%)	Hardness (Vickers) (DPH)
As irradiated	7.7	61
1 hr at 220°C	7.0	51
1 hr at 400°C	1.1	40
20 hr at 540°C	5.2	38
Unirradiated 20 hr at 540°C	---	25

In agreement with the decrease in strength of the higher fluence specimens upon annealing (Figure 7), a decrease in room temperature hardness also occurs. The hardness after 20 hours at 540°C approaches but does not drop to that of the unirradiated 1100 Al heated equivalently. This difference is not due to the silicon in solution but to the coarsened silicon precipitate generated from nuclear transmutation.

The density upon annealing does not change much at 220°C, which is consistent with the lack of much change in void density or size distribution at about this temperature, as will be shown subsequently. The density increase at 400°C indicates the voids collapse at about this temperature. The subsequent decreased density at 540°C is presumed due to gas bubble growth.

V. COMPOSITION CHANGES

The transmutation products from long-term irradiation can build to significant concentrations which affect the properties of the irradiated material. The formation of silicon in aluminum alloys from the nuclear reactions involving the absorption of thermal neutrons with subsequent radioactive decay to silicon is the major constituent. Other products that can affect the properties, especially at temperatures above one-half the absolute melting temperature ($T_m/2$), are the gaseous products helium and hydrogen from nuclear reactions with the fast neutrons.

The major contribution of the silicon comes from the reaction

since the Al^{27} isotope comprises 100% of the isotopic composition of the natural aluminum and the Al^{28} isotope decays with a 2.3 minute half-life. Using this assumption and a cross section of 0.23 barn for thermal neutron absorption, the calculated conversion to silicon for a sample from the highest flux region is given in Table IV and compared to the wet chemical analysis. If the 0.1% silicon normally present in 1100 Al is added to the calculated conversion, the result is similar to the wet chemical analysis. However, this result is better than expected because of the uncertainty in the thermal flux values previously discussed.

The hydrogen and helium production were calculated by the assumption that the principal contribution comes from (n, p) and (n, α) reactions with aluminum. The fission-spectrum-integrated reaction cross sections used were 3.1 mb for hydrogen and 0.65 mb for helium production^[5]. The values calculated are compared in Table IV with the quantity measured by gas analysis. The helium production agrees fairly well, while the hydrogen found by gas analysis is three to four times higher than that calculated. Since the experimental cross section is considered acceptable to within a factor of two, the difference in hydrogen content is thought to be due to pickup during in-reactor corrosion reactions and from proton recoil.

TABLE IV

TRANSMUTATION PRODUCTS IN IRRADIATED ALUMINUM

Sample	Si Calculated Conversion wt%	Si Wet Analysis wt%	He Calculated Conversion ppm wt	He Gas Analysis ppm wt	H ₂ Calculated Conversion ppm wt	H ₂ Gas Analysis ppm wt	ϕ Fast E > 1 MeV	ϕ th
1 MHD	1.55	1.6	5.6	4.5	6.5	25.8	5.6×10^{22}	6.5×10^{22}
D-3	1.3	---	4.0	3.8	4.6	13.7	4.0×10^{22}	5.4×10^{22}
D-4	0.9	---	1.8	1.1	2.1	7.1	1.8×10^{22}	3.9×10^{22}

VI. MICROSTRUCTURE

The cross sections for preparation of the photomicrographs were taken parallel to the tensile axis and close to the fracture surface. The photomicrographs show lineup of voids and impurities as the result of the tensile deformation and reduction in cross section in the as-irradiated structure (Figure 8) and the additional effect of heating at 100, 200, and 300°C (Figures 8 and 9). These cross sections were all taken from the higher fluence (3.9 to 4.5×10^{22} n/cm²) specimens; hence they contained numerous voids after irradiation. Voids are compared at a fluence of 1.8×10^{22} n/cm² with that of 5.6×10^{22} n/cm² in Figure 11. The micrographs show elongated islands of deformed material interspersed with cavities and impurities which tend to align and elongate in the island boundary. The effect of temperature is to produce recovery, ie, loss of the displacement damage and coalescence of the cavities and the silicon (Figures 8 and 9).

The microstructure was further investigated by transmission electron microscopy at 175 kV maximum for any changes due to the deformation, the annealing and test temperature, and the effects of fluence on the voids, dislocations, and silicon precipitates. Figures 11 through 16 illustrate these changes. As seen in Figure 11, voids appear to be just measurable at the fluence of 1.8×10^{22} n/cm², which corresponds to a density decrease that is also just significant (Figure 2).

From stereomicrographs the thickness of the foils was determined and the void density calculated as given in Table V, wherein the void density increased over 160 times with the increase in fluence. The calculated volume change from voids is less than the measured volume change in Table I, which is attributed principally to the transmutation of aluminum to silicon with density change from 2.7 to 2.33 g/cm³. If one assumes that the conversion of aluminum atoms to silicon atoms occurs with precipitation, then the change in density would be the change in density of the lattice cells per atom converted. Table V shows the volume change adjusted for silicon production. The adjusted volume change (calculated from voids and silicon production) is less than the measured density decrease in Table I, except for the low fluence specimen, D-2, but is in reasonable agreement.

Spec. D-2 $\Phi = 1.8 \times 10^{22}$

15,000X

Spec. 1 MHD $\Phi = 5.6 \times 10^{22}$

15,000X

Fig. 11 Transmission electron micrographs showing increase in void density of as-irradiated aluminum with increase in fluence.

Fig. 12 Increase in void density in deformed regions of specimens at fluence of $1.9 \times 10^{22} \text{ n/cm}^2$.

Fig. 13 Silicon precipitates distributed throughout while the voids are depleted at the grain boundary.

A. Spec. B-5 Gage, No. 653 200°C
 $\Phi = 1.9 \times 10^{22} \text{ n/cm}^2$

B. Spec. A-5 Gage, No. 635 200°C
 $\Phi = 4.3 \times 10^{22} \text{ n/cm}^2$

C. Spec. B-5 Head, No. 615 200°C

D. Spec. A-5 Head, No. 637 200°C

Fig. 14-Deformed regions in gage develop some subcell formation but less than at 100°C.

C. Spec. A-7 Head, No. 628 300°C

D. Spec. A-7 Head, No. 630 300°C

Fig. 15 Voids present after 300°C anneal for three hours and testing.

Fig. 16 Voids present in clusters after deformation at 300°C in gage section.

The distribution of silicon in an as-irradiated high fluence specimen and specimen A-7 which had been heated for three hours at 300°C was examined by electron diffraction. The as-irradiated specimen did not indicate the presence of silicon. However, the dark spots in specimen A-7 (Figure 16) were identified by electron diffraction as being silicon. Selected area diffraction, Figure 17, revealed a ring pattern in addition to the diffraction spots of aluminum. The ring pattern corresponded to the hkl planes for silicon of (111), (220), and (311) in the ASTM card index. Verification of the particles producing the ring pattern was made through dark field transmission techniques.

Examination of TEM photographs (typical ones are shown in Figures 11 through 15) revealed the following:

- (1) At ambient temperature and 100°C the silicon is distributed mostly as very fine particulate.
- (2) At fluences less than $2 \times 10^{22} \text{ n/cm}^2$ $E > 1 \text{ MeV}$ the void density is noticeably less than at high fluences.
- (3) After deformation the voids in the gage section of tensile specimens were found in patches, but the silicon distribution was not greatly affected.

TABLE V

VOID DENSITY AND CALCULATED VOLUME CHANGE IN IRRADIATED ALUMINUM

Sample No.	Fluence n/cm^2 $E > 1 \text{ MeV}$	Void Density $\frac{\text{voids}}{\text{cm}^3}$	Void Size Range ($\text{\AA}$)	Average	Calc. $\Delta V/V\%$	$\frac{\Delta V}{V}$ adjusted for silicon Production (%)
1 MDH	5.6×10^{22}	2.3×10^{14}	240-1200	800	6.2	6.8
D1	4.1×10^{22}	2.5×10^{14}	240-880	640	3.4	3.9
D2	1.8×10^{22}	1.4×10^{12}	350-1400	1100	0.1	0.4

- (4) The density of the voids in high fluence specimens (3.9 to 4.3×10^{22} n/cm² $E > 1$ MeV) was not greatly affected by temperatures of 200 to 300°C for three hours.
- (5) A void depleted region adjacent to the grain boundary was observed.

The lack of much change in void density in the high fluence specimens at 300°C is an interesting result and explains the decrease in strength and ductility of these specimens when compared with the intermediate fluence specimens. An effect of high fluence on void annealing temperature was found by comparing with the results of Jostons et al[6] on 1100 aluminum irradiated to a fluence of about one third that of the present specimens wherein recovery of density at about 300°C after 1 and 20 hours annealing occurred. Transmission electron microscopy of specimens heated at 400°C for one hour was required for void collapse in our high fluence specimen. Most regions were depleted of voids, while a few regions in the foil had voids with a larger size distribution (Figure 18). The collapse in treating times of one or three hours evidently occurs very sharply, since one hour at 400°C produced recovery of the immersion density while three hours at 300°C produced little change in the void density or size distribution. The present results indicate that at the higher fluence a higher annealing temperature is required for void collapse.

Fig. 17 Ring pattern of silicon imposed on diffraction spots of aluminum -- Specimen A-7.

Fig. 18 Voids after 400°C anneal.

VII. CONCLUSIONS

- (1) The mechanical properties of 1100 Al after irradiation at 70°C to a fast fluence of 4.5×10^{22} n/cm² are considered to be acceptable for continued reactor service. The uniform elongation was 7 to 10% with local ductility of 46% reduction-in-area.
- (2) A 7.7% reduction in density occurred at 5.6×10^{22} n/cm² principally due to formation of voids.
- (3) Annealing at 300°C for three hours produced a reduction in strength and ductility as recovery and redistribution of the principal hardening agents, silicon and point defects, occurred.
- (4) Comparison of calculated and measured transmutation products of silicon, helium, and hydrogen showed reasonable agreement except for hydrogen where pickup from corrosion or proton recoil is thought to occur.
- (5) Irradiation to higher fluence required a higher annealing temperature for void collapse.

VIII. REFERENCES

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