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DEVELOPMENT OF SILICON-MODIFIED 48 wt % U-Al ALLOYS FOR ALUMINUM PLATE-TYPE FUEL ELEMENTS

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OAK RIDGE NATIONAL LABORATORY
operated by
UNION CARBIDE CORPORATION
for the
U.S. ATOMIC ENERGY COMMISSION

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METALLURGY DIVISION

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SUMMARY

The 20% limitation on U-235 enrichment of uranium exported under the Atoms-for-Peace Program has created the need for development of 35 to 50 wt% U-Al alloys. These alloys, when modified with ternary additions of 3 wt% Si, Zr, Ge, Ti, or Sn, contain essentially no UAl_4, but instead, retain UAl_3 as the stable compound. This structural modification, which reduces the volume percentage of intermetallic compound, improves the quality of the fuel alloy in comparison with an unmodified alloy of the same uranium content. The present work discusses casting procedures, mechanical working characteristics, composite fuel plate fabrication, and fuel element assembly methods for a nominal 48 wt% U - 3 wt% Si-Al alloy.

It was found that localized clad thinning, resulting from "dogboning" of the fuel alloy is essentially eliminated in composite plates when this alloy is used in combination with 5154 aluminum frames, and 1100 aluminum covers. However, fuel component dimensions are difficult to maintain within commonly accepted tolerances in the brazing process because of the marked differences in the thermal expansion of the fuel alloy and the containment materials. The mechanical properties, irradiation studies, corrosion testing, pneumatic pressure testing, and chemical reprocessing of the silicon-modified alloy are also discussed.

INTRODUCTION

The stipulation that uranium exported under the Atoms-for-Peace Program be limited to 20% enrichment in the U-235 isotope has created a challenging need for the development of new fuel compositions for the aluminum research reactor fuel elements. As illustrated in Fig. 1 (Y-9250), this fuel element is of a plate-type design and generally contains 140 to 200 g of uranium, enriched in U-235 at the 93% level. Obviously, reducing the enrichment to 20% requires approximately a five-fold increase in contained uranium, which, in turn, increases the concentration of uranium in the fuel alloy.

The most expedient approach to the problem of obtaining increased uranium loadings is to modify and extend the well-developed procedures employed in the

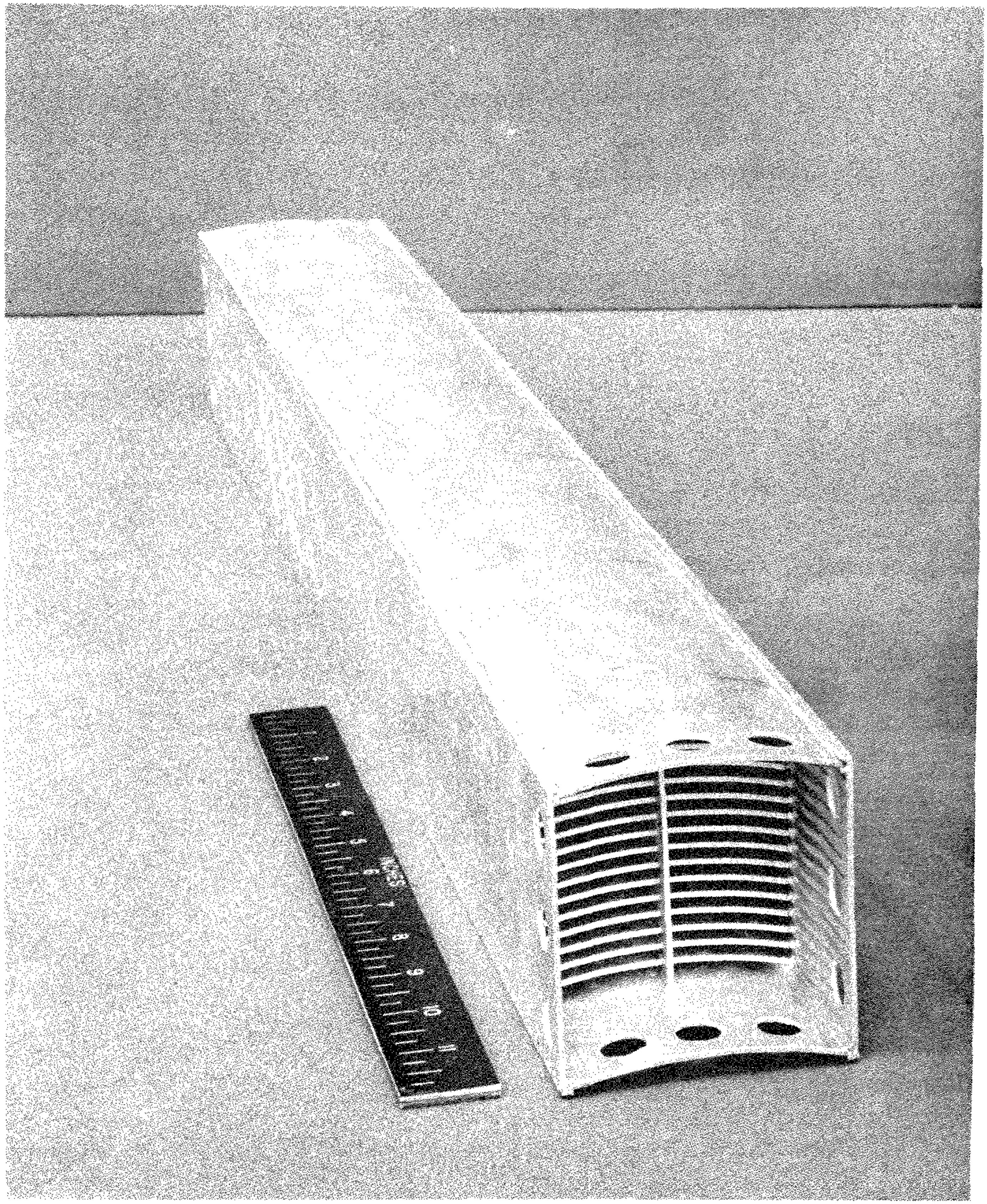

Fig. 1 (Y-9250) Typical MTR-Type Fuel Element

manufacture of MTR-type elements[1] to the fuel alloy compositions required for uranium limited to 20% enrichment. If a standard design composed of an array of 18 alclad, 60-mil-thick fuel plates integrally brazed to grooved side plates is selected as the reference element, it can be calculated that a 48 wt% U-Al alloy is required to yield the same loading of U-235 that normally can be obtained with an 18 wt% U-Al alloy containing highly enriched uranium.

The problems associated with melting and casting of a nominal 48 wt% U-Al alloy, as well as primary and composite plate rolling, were discussed previously.[2] It was pointed out that the casting problems encountered were primarily inhomogeneity and gas porosity stemming from the wide gradient between solidus and liquidus curves for this alloy. Breakdown rolling of the cast billet into plate stock for fuel cores also proved to be troublesome and an aluminum frame was necessary to laterally constrain the billet during rolling. The principal area of concern with the nominal 48 wt% U-Al alloy, however, was the deleterious end effect which occurred during the manufacture of composite fuel plates. This end effect, aptly termed a "dogbone" because of its peculiar shape, is characterized by localized thickening of the fuel alloy and thinning of the protective cladding at the extremities of the fuel core. It was demonstrated that "dogboning" is a result of the marked difference in plasticity between the high-uranium core alloy and the aluminum containment materials at the elevated temperature required for rolling. This defect could be somewhat ameliorated by the substitution of higher strength aluminum alloys for 1100 aluminum.[3] The "dogboning" was still considered serious, however. Subsequent studies have been directed toward modification of the fuel alloy by ternary additions which suppress the UAl_4 intermetallic compound, resulting in a substantial reduction in yield strength of the alloy. The present report describes the development of a nominal 48 wt% U - 3 wt% Si-Al alloy to fulfill this need for fuel plates of maximum integrity.

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[1] J. E. Cunningham and E. J. Boyle, "MTR-Type Fuel Element," Proceedings of International Conference on Peaceful Uses of Atomic Energy, 9, 203 (1955).

[2] W. C. Thurber, J. H. Erwin, and R. J. Beaver, The Application of a Nominal 48 Wt% U-Al Alloy to Plate-Type Aluminum Research Reactor Fuel Elements, ORNL-2351 (February, 1958).

[3] The Aluminum Association standard designations for wrought aluminum alloys are used throughout this report.

Included in the discussion are alloying theory and practices, primary fabrication by hot rolling and extrusion, composite fuel plate manufacture with particular emphasis on minimization of "dogboning," component assembly, and mechanical and physical property evaluation. In many cases, the nominal 48 wt% U-Al binary alloy is used as a reference for evaluating the relative effects of ternary silicon additions.

CONCLUSIONS

1. Ternary additions of 3 wt% Si, Ge, Zr, Sn, or Ti result in virtually complete UAl_4 suppression in 48 wt% U-Al alloys with a decrease of ~ 18% in the volume of intermetallic compound.

2. The silicon addition is associated with aluminum in the UAl_3 lattice, forming compounds of the $U(Al,Si)_3$ type in which the lattice parameters are a function of the silicon content of the alloy.

3. No reversion of UAl_3 to UAl_4 occurs in the silicon-modified alloy after extensive thermal treatment at 605°C.

4. Silicon additions alter the macrostructure of the 48 wt% U-Al alloy castings from columnar to equi-axed and also eliminate hot tears in the cast billets.

5. The mode and magnitude of uranium segregation in 40 to 50 wt% U-Al alloys are not affected by the 3 wt% silicon addition.

6. The ternary alloys can be rolled into plate without framing, and casting-to-core yields of 40 to 50% can be realized.

7. Silicon additions to the fuel alloy decrease the magnitude of "dogboning" in composite fuel plates regardless of the type of aluminum considered for frames and covers.

8. "Dogboning" in composite fuel plates can be further reduced by using picture-frame material with higher yield strength than 1100 Al. The optimum materials combination consists of 48 wt% U - 3 wt% Si-Al cores, 1100 Al covers and alclad 5154 Al frames.

9. In the assembly of fuel elements by brazing, it is difficult to maintain plate spacings to within ± 10% of the nominal specification. If the specific reactor application requires these tolerances, it is recommended that mechanical assembly be considered as a desirable alternative.

10. The tensile and yield strengths of a nominal 48 wt% U-Al alloy are reduced with silicon additions because of the decreased quantity of dispersed intermetallic compound. Tin additions exert a similar effect.

11. Corrosion testing of unclad samples of the silicon-modified alloy in static, de-ionized water at 60 and 90°C indicates that, although general attack is very limited, some blisters form after three weeks exposure. The propensity to blister does not appear to increase with exposure time.

12. Pneumatic pressure testing of 18-plate fuel elements prepared with high-strength aluminum alloy frames and 48 wt% U - 3 wt% Si-Al alloy cores reveals that these elements are more resistant to deformation from pressure differentials than are standard Mark XI MTR-type fuel units.

DEVELOPMENT OF THE TERNARY ALLOY

It has been reported that the addition of 0.8 wt% Si to a 20 wt% U-Al alloy is suffucient to completely suppress the peritectic transformation of UAl_3 to UAl_4 occurring at 750°C,[4,5] allowing the primary nucleating compound, UAl_3, to be retained as the stable phase at room temperature. It can be predicted from the U-Al constitution diagram, shown in Fig. 2 (Y-20808), that a similar structural modification in a nominal 48 wt% U-Al alloy would decrease the volume fraction of intermetallic compound in the alloy by 18%. When working materials of marginal fabrication potential, such a change is obviously significant.

To establish the feasibility of UAl_4 suppression in the high-uranium alloys, a series of six castings containing amounts of silicon varying from 0 to 3 wt% were prepared by conventional open-air-induction melting techniques. Each alloy was cast into a graphite mold 1-1/2 in. in dia and 8 in. long. Samples were then taken from the extremities of each ingot for silicon analysis and determination of the alloy constituents by x-ray diffraction. Results of this study

[4] Metallurgy Division Quar. Prog. Rep. for period ending October 31, 1950, ORNL-910.

[5] M. L. Picklesimer, private communication.

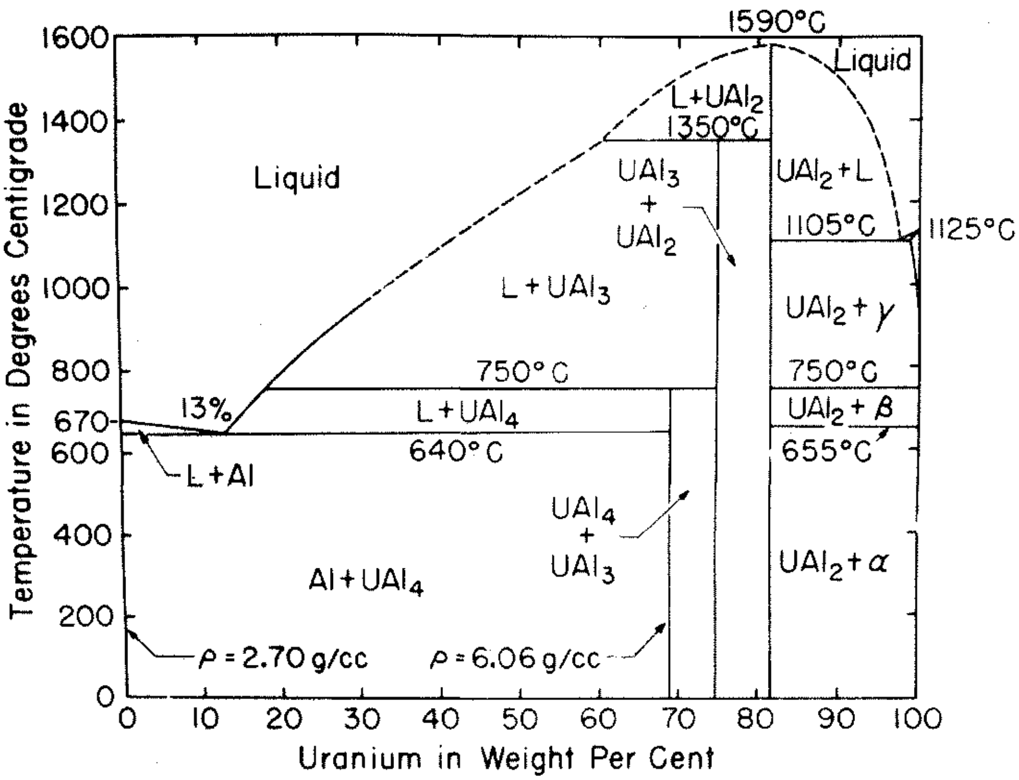

Fig. 2 (Y-20808) Aluminum-Uranium Phase Diagram.

are included in Table I. The results presented indicate that increasing UAl_4 suppression is achieved with increasing amounts of silicon and that no traces of UAl_4 are found in alloys containing more than 1.5 wt% Si.

The six small ingots were sectioned along vertical center lines and drillings were taken at appropriate intervals for both uranium and silicon analyses. Analytical results from a typical casting (Si - 5 - C) are shown in Fig. 3 (ORNL-LR-DWG 30500). The distribution patterns for uranium and silicon are essentially identical, indicating that both of these elements are associated with the intermetallic compound $U(Si,Al)_3$. This would be expected since UAl_3 and USi_3 are isomorphous, having ordered structures of the $AuCu_3$ type.

Further evidence of the existence of silicon in the intermetallic compound was obtained by measuring the lattice parameters of the $U(Al,Si)_3$ in nominal 48 wt% U-Al alloys containing up to 10 wt% silicon. These data, which are presented in Fig. 4 (ORNL-LR-DWG 20534), clearly indicate the decrease in lattice constant of the $U(Si,Al)_3$ with increasing silicon content, providing additional evidence of the partitioning of the ternary addition to the intermetallic compound.

It was hypothesized that other elements which form compounds isomorphous with UAl_3 might also be effective in suppressing the UAl_4 phase. A literature survey[6] revealed that the compounds of all the Group III-B and Group IV-B elements in the Periodic Table including Al, Ga, In, Tl, Ge, Si, Sn, and Pb formed $AuCu_3$-type compounds. Castings of a 48 wt% U-Al alloy, 1-1/2 in. in dia by 8 in. long, were prepared with 3 wt% additions of these elements. Similar castings with ternary additions of Nb, Fe, Zr, and Ti were also prepared. As previously described, samples from each ingot were analyzed by x-ray techniques for the alloy components. It was found that in every case, some suppression of UAl_4 was affected and, as can be seen from Table II, virtually complete UAl_4 suppression was realized with 3 wt% additions of Si, Sn, Zr, Ge, and Ti. It is conceivable that complete UAl_4 suppression could be achieved with most ternary additions investigated at atomic percentage levels equivalent to silicon (i.e., 5 at%). Although any one of several ternary additions might

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[6] H. A. Saller and F. A. Rough, Compilation of U. S. and U. K. Uranium and Thorium Constitutional Diagrams, BMI-1000 (June 1955).

TABLE I

EFFECT OF SILICON ON THE SUPPRESSION OF UAl_4
IN CAST 48 Wt% U-Al ALLOYS

Ingot	Sample Location	Intended Silicon Content Wt%	Analyzed Silicon Content Wt%	X-Ray Spectrometer Trace		
				Al-%	UAl_3-%	UAl_4-%
Si-1-C	Top	< 0.2	0.12	50	--	50
	Bottom		0.15	50	--	50
Si-3-C	Top	0.5	0.90	40	30	30
	Bottom		0.88	30	50	20
Si-2-C	Top	1.0	1.57	45	50	5
	Bottom		1.03	25	70	5
Si-4-C	Top	1.5	2.20	30	70	--
	Bottom		1.82	30	70	--
Si-5-C	Top	2.0	2.89	40	60	--
	Bottom		2.01	30	70	--
Si-6-C	Top	3.0	----	35	65	--

TABLE II

EFFECT OF TERNARY ALLOY ADDITIONS ON THE SUPPRESSION OF UAl_4 IN 48 Wt% U-Al ALLOYS

Ternary Addition	Ternary Metal Content Wt%	Ternary Metal Content At.%	Spectrometer Trace % UAl_3	Spectrometer Trace % UAl_4	Spectrometer Trace % Al	Lattice Parameter UAl_3 A°	Lattice Parameter Al A°
Si	3	5.0	65	--	38	4.268	4.052
Ti	3	3.0	67	8	25	4.263	4.050
Ge	3	2.0	60	--	40	4.279	4.050
Zr	3	1.6	70	--	30	4.248	4.047
Sn	3	1.2	70	--	30	4.284	4.049
Pb	3	0.7	20	40	40	-----	-----
Ga	3	2.1	10	48	42	-----	-----
In	3	1.3	35	35	30	-----	-----
Tl	3	0.7	21	38	41	-----	-----
Fe	3	2.6	20	38	42	-----	-----
Nb	3	1.6	24	42	34	-----	-----
--*	-	---	--	55	45	-----	-----

*No ternary addition (48 U-Al)

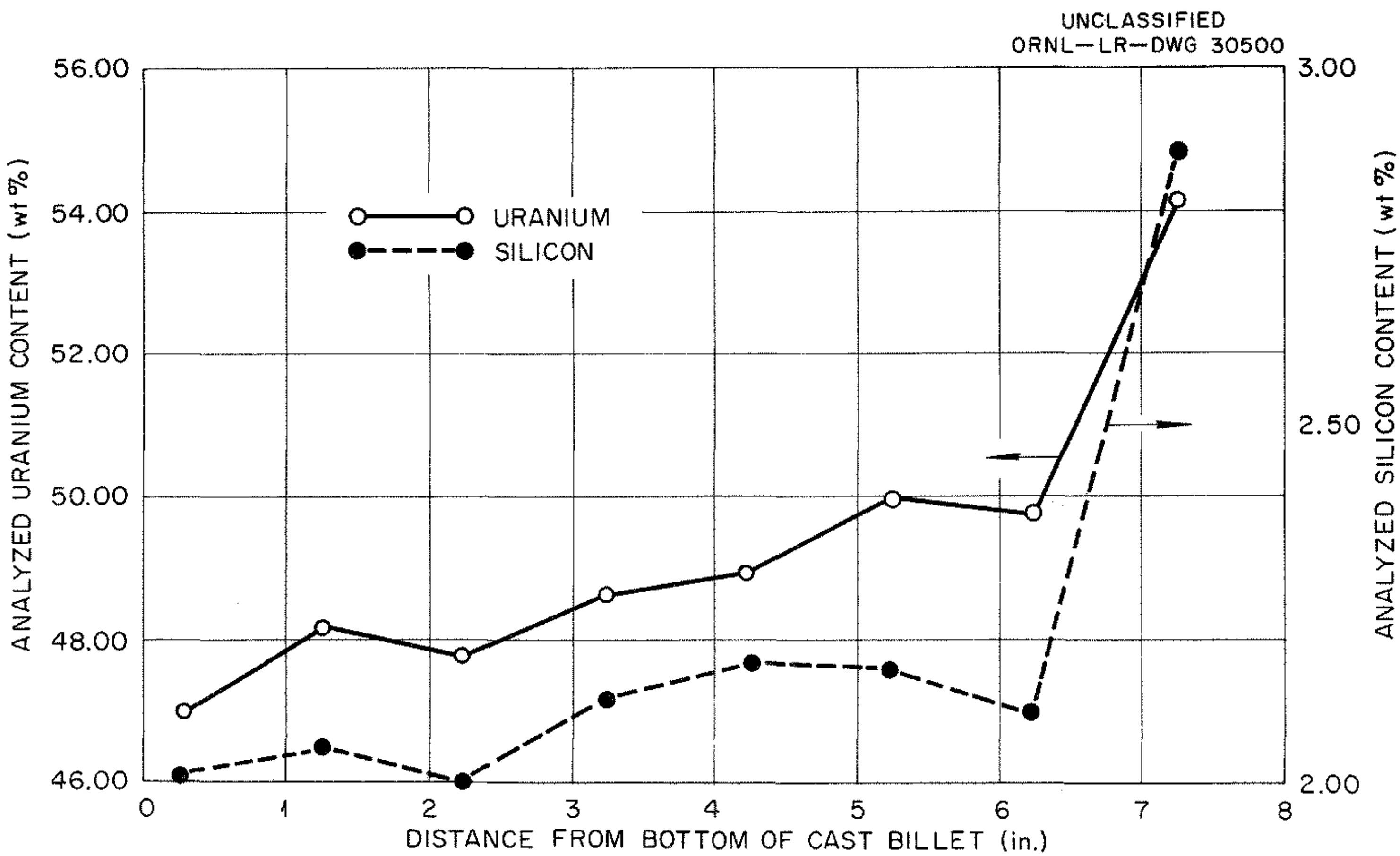
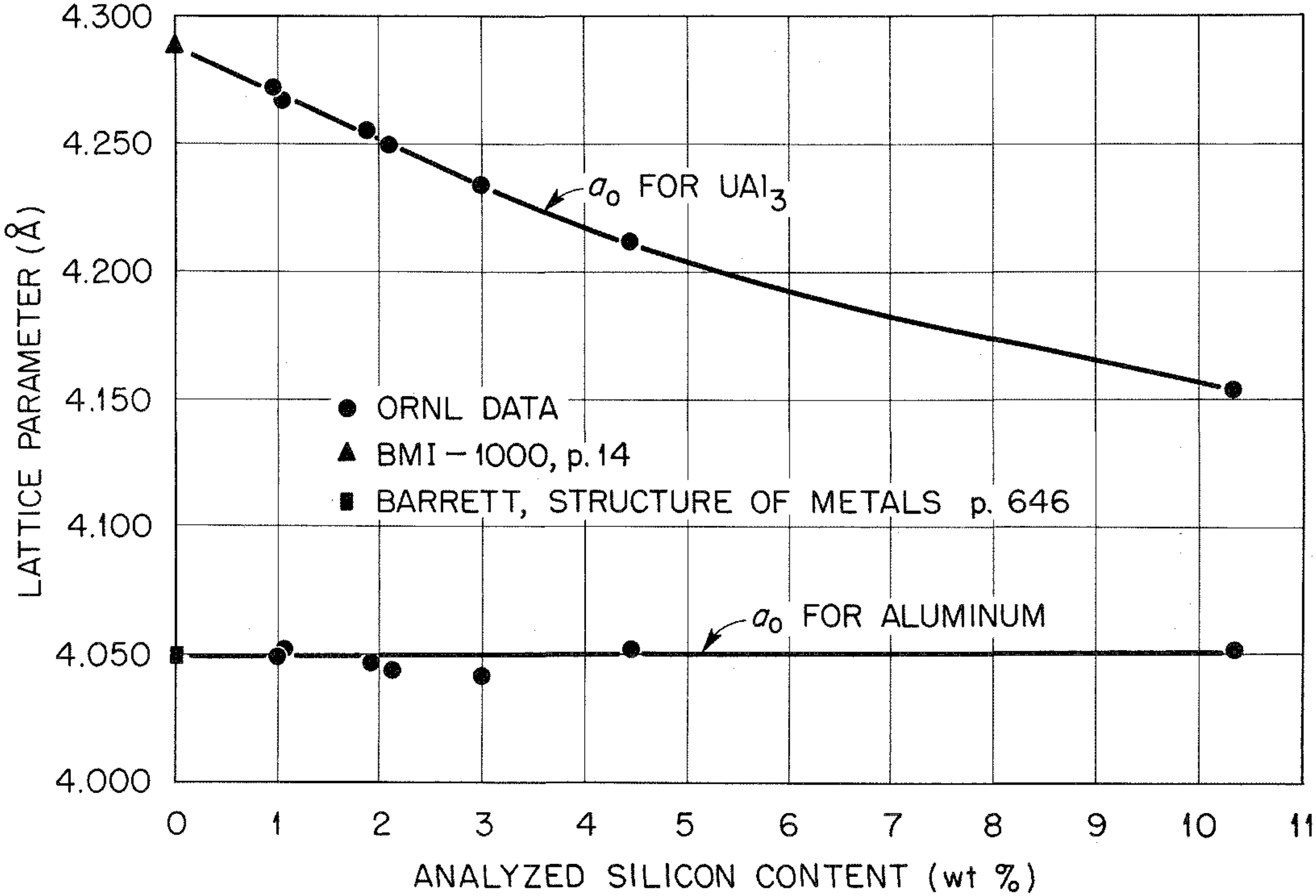


Fig. 3. Silicon and Uranium Distribution in Nominal 48 wt % U – 2.0 wt % Si-Al Alloy.



Effect of Silicon on the Lattice Parameters of UAl_3 and Aluminum.

Fig. 4

be used for modifying the 48 wt% U-Al alloy, only silicon was selected for extensive study. A limited amount of information, however, was obtained from a tin-bearing alloy.

Cooling curves were obtained from four different heats of the reference alloy and it was observed that, within the accuracy of measurement (± 5°C), 3 wt% silicon additions induced no change in the reported eutectic temperature of 640°C. No measurements of the liquidus temperature were attempted.

It was also found that in alloys containing less uranium than the peritectic composition (which would ordinarily nucleate UAl_4 rather than UAl_3 as the primary compound) UAl_4 suppression was also achieved. For example, a 1 wt% Si addition was made to a casting of the eutectic composition alloy (13 wt% U-Al) and x-ray examination of the material revealed the total absence of UAl_4.

To determine whether or not the UAl_3 in a 48 wt% U - 3 wt% Si-Al alloy would transform to UAl_4 after appreciable time at elevated temperatures, a sample of the alloy was held 312 hr at 605°C. No phase transformation was observed in this experiment.

MELTING AND CASTING

Melting and casting procedures for producing the 48 wt% U - 3 wt% Si-Al alloy were basically the same as those employed for producing the silicon-free alloy.[7] Standard practice involved conventional open-air-induction melting techniques. The silicon was added either as lump metal or as vacuum-melted 50 wt% Si - 50 wt% Al master alloy. When the vacuum-melted master alloy was used it was generally not necessary to presolidify the melts for degassification; instead, the as-melted charge could be immediately poured into an appropriate graphite mold. The melts were poured at 1175°C into molds heated to about 325°C.

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[7]Thurber, Erwin, and Beaver, op.cit.

Although graphite crucibles are used at ORNL, personnel of Babcock and Wilcox Company have indicated a preference for magnesia or alumina crucibles.[8] The use of oxide crucibles eliminates the possibility of carbon contamination in the melt and subsequent formation of harmful carbide inclusions during extensive scrap recycling associated with production practice. These carbides are believed to be sites for the nucleation of blisters and clad ruptures.

The addition of silicon to a 48 wt% U-Al alloy produces some rather striking changes in the cast billets. Figure 5 (Y-21833) illustrates the surface appearance of silicon-bearing and silicon-free alloys. It can be seen that the silicon addition completely eliminates hot tears at the juncture of the head and body of the rectangular billet. Although not visible, cracking along the edges is also eliminated. In Fig. 6 (Y-22051), the macrostructure of the cast billets at a section through the center is presented. The transition from a completely columnar to a completely equiaxed grain pattern is obvious. These alterations in the grain pattern and hot-tearing tendencies no doubt contribute to the improved fabrication qualities of the ternary material.

An identical casting poured from 1350°C rather than the reference temperature of 1175°C also had an equiaxed structure, indicating that the sources of crystal nucleation producing the equiaxed grains are not eliminated with additional superheat. It can also be seen from Fig. 6 that the porosity in the head of the silicon-free castings resulting from intercrystalline shrinkage is drastically reduced by the 3 wt% silicon addition.

Uranium segregation in as-cast silicon-modified alloys is inverse in nature, resulting from interdendritic feeding of the eutectic liquid away from the last-to-freeze regions of the casting. This mode of segregation produces a uranium-rich zone at the thermal center of the billet. The same type of segregation is observed in silicon-free alloys, indicating that silicon additions do not affect uranium homogeneity. Figure 7 (ORNL-LR-DWG 30501) illustrates the uranium distribution in billets poured from the reference temperature of 1175°C as well as from 1350°C. Since little difference in segregation pattern

[8] N. K. Jesson and E. L. Lee, Babcock and Wilcox Company, private communication.

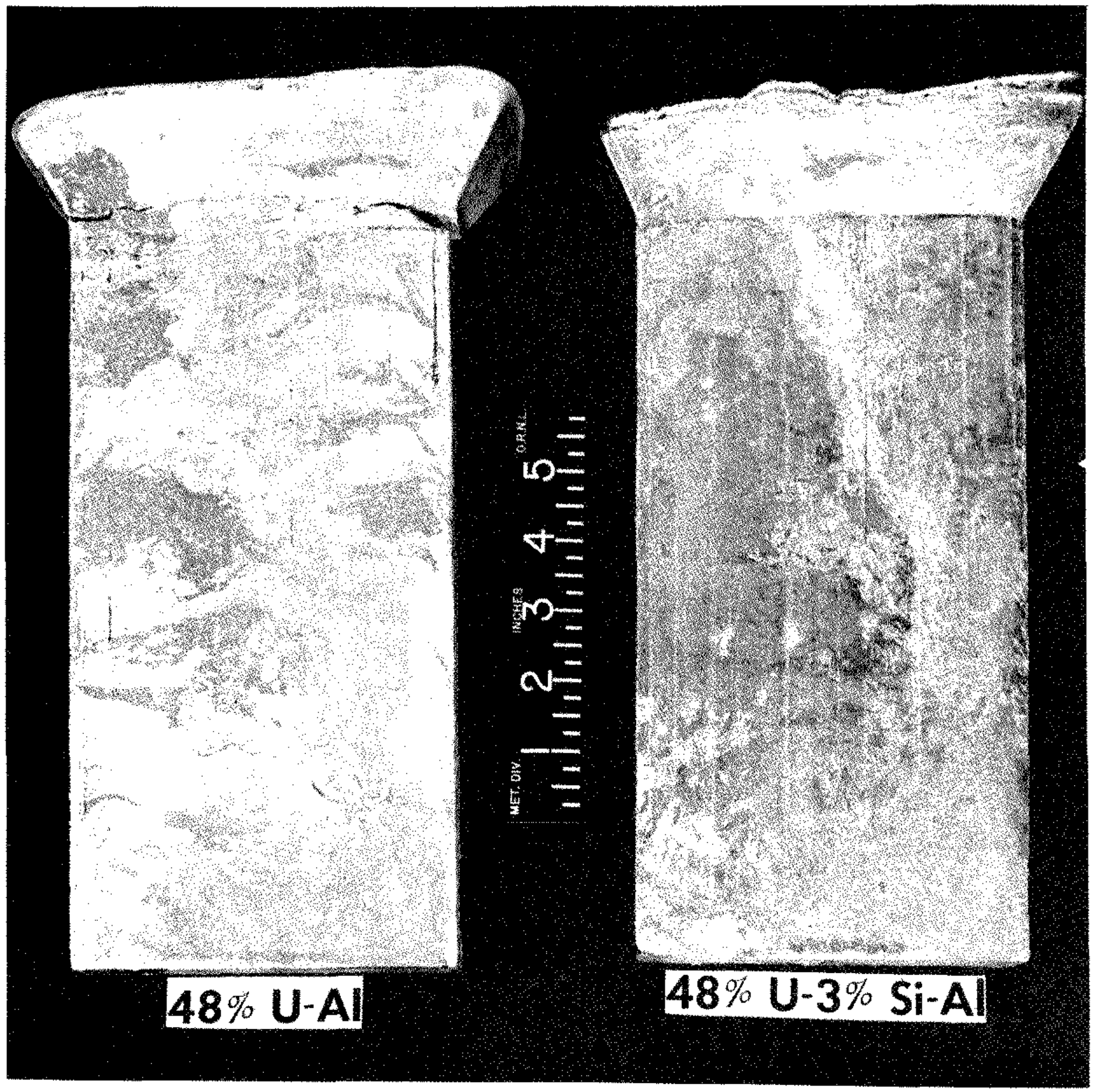
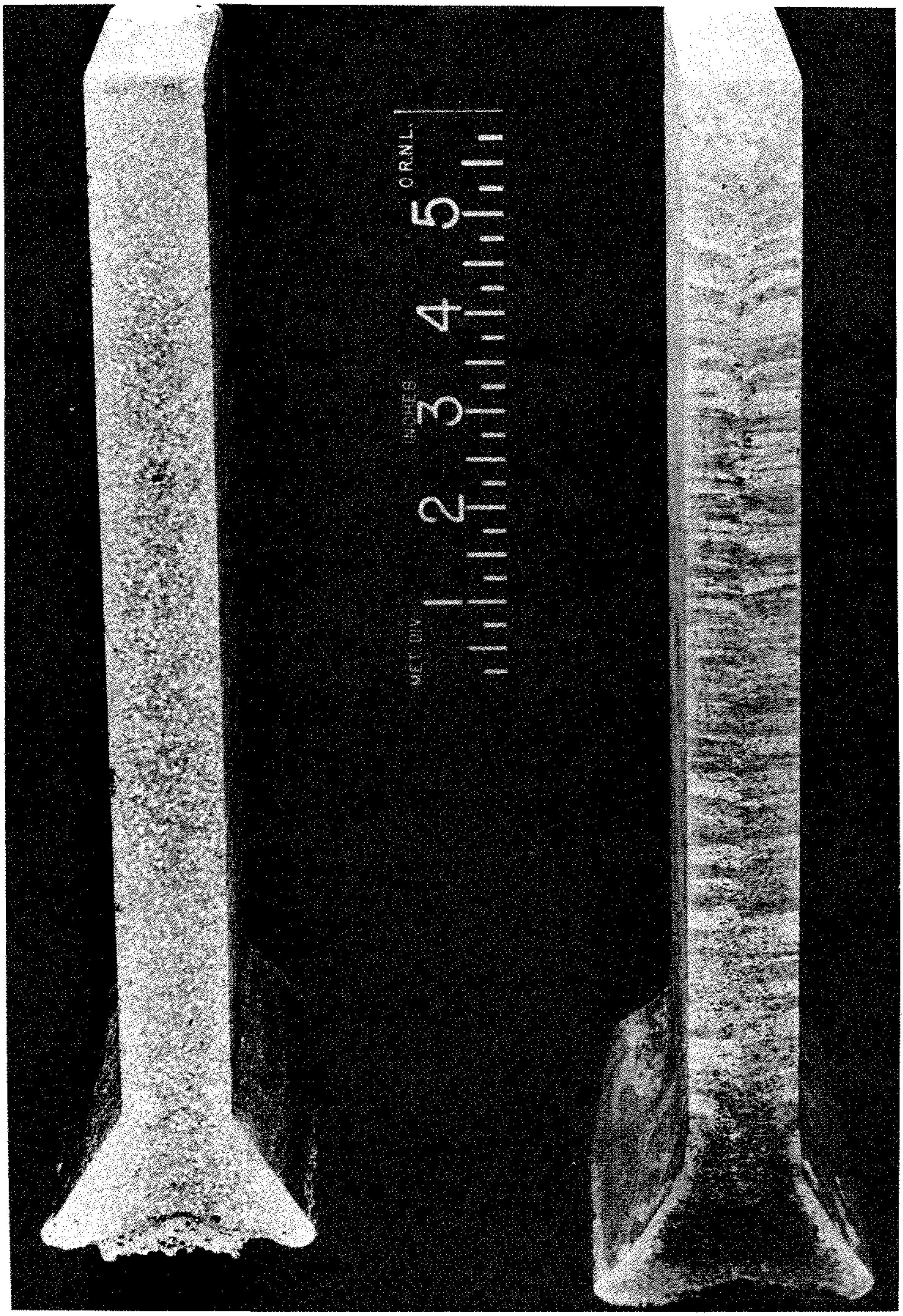


Fig. 5 (Y-21833) Surface Appearance of Cast Billets Showing Elimination of Hot Tears with Silicon Addition.



48 wt% U-3 wt% Si-Al 48 wt% U-Al

Fig. 6 (Y-22051) Refinement of Macroscopic Grain Size with Silicon Additions in Cast 48 Wt% U-Al Alloys.

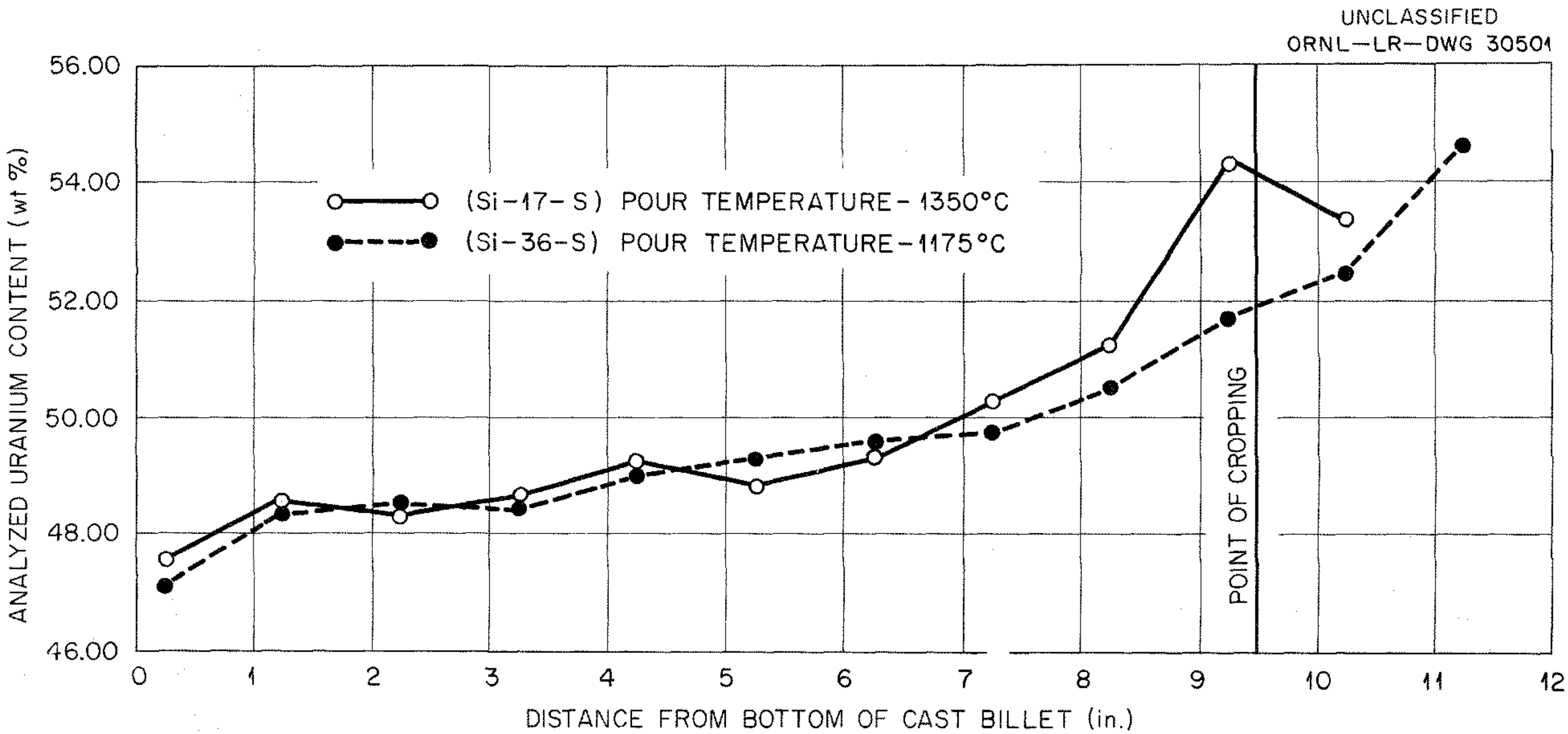


Fig. 7. Uranium Distribution in Nominal 48 wt % U - 3wt % Si-Al Alloys.

is noted with the two pouring temperatures, it is recommended that the selected pour temperature be 1175°C and that the head of the casting be cropped as shown in Fig. 7. A more detailed discussion of segregation and its effect on uranium accountability can be found in ORNL-2476.[9]

PRIMARY ALLOY FABRICATION

Fabrication by Hot Rolling

In the breakdown rolling of 1-in.-thick cast billets of nominal 48 wt% U-Al alloy into 1/4-in.-thick plates, it was necessary to laterally constrain the fuel material in an aluminum alloy frame and also envelope it with thin (0.025-in.) aluminum cover sheets to minimize edge and surface cracking. The component hardware for this process at various stages of completion is illustrated in Fig. 8 (Y-21018). The disadvantages of this type of hot-rolling are increased costs of machining the frame and fuel alloy, and the problem of accurate uranium accountability resulting from the thin, integrally bonded aluminum cover sheet on the fuel cores. With the 3 wt% silicon addition, however, the as-cast billets can be reduced directly into core plate after merely cropping the head from the casting. Casting-to-core material yields, only slightly lower than those obtained with the framed billets, were achieved when a cross-rolling pass was introduced into the rolling schedule. Yields of 40 – 50% were routinely obtained with the principal material losses being the casting head and punching scrap. However, if desired, the scrap can be recovered by recycling in subsequent melts. The appearance of unmodified alloy and silicon-modified alloy after hot rolling is shown in Fig. 9 (Y-22362). The binary alloy cracked and virtually disintegrated after less than 10% reduction in thickness at 600°C; whereas, the ternary alloy was successfully reduced 75% to 1/4-in.-thick plate.

This marked improvement in fabricability of the silicon-modified alloys results both from the equiaxed grain structure, previously discussed, and the

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[9] W. C. Thurber and R. J. Beaver, Segregation in Uranium-Aluminum Alloys and Its Effect on the Fuel Loading of Aluminum-Base Fuel Elements, ORNL-2476 (September, 1958).

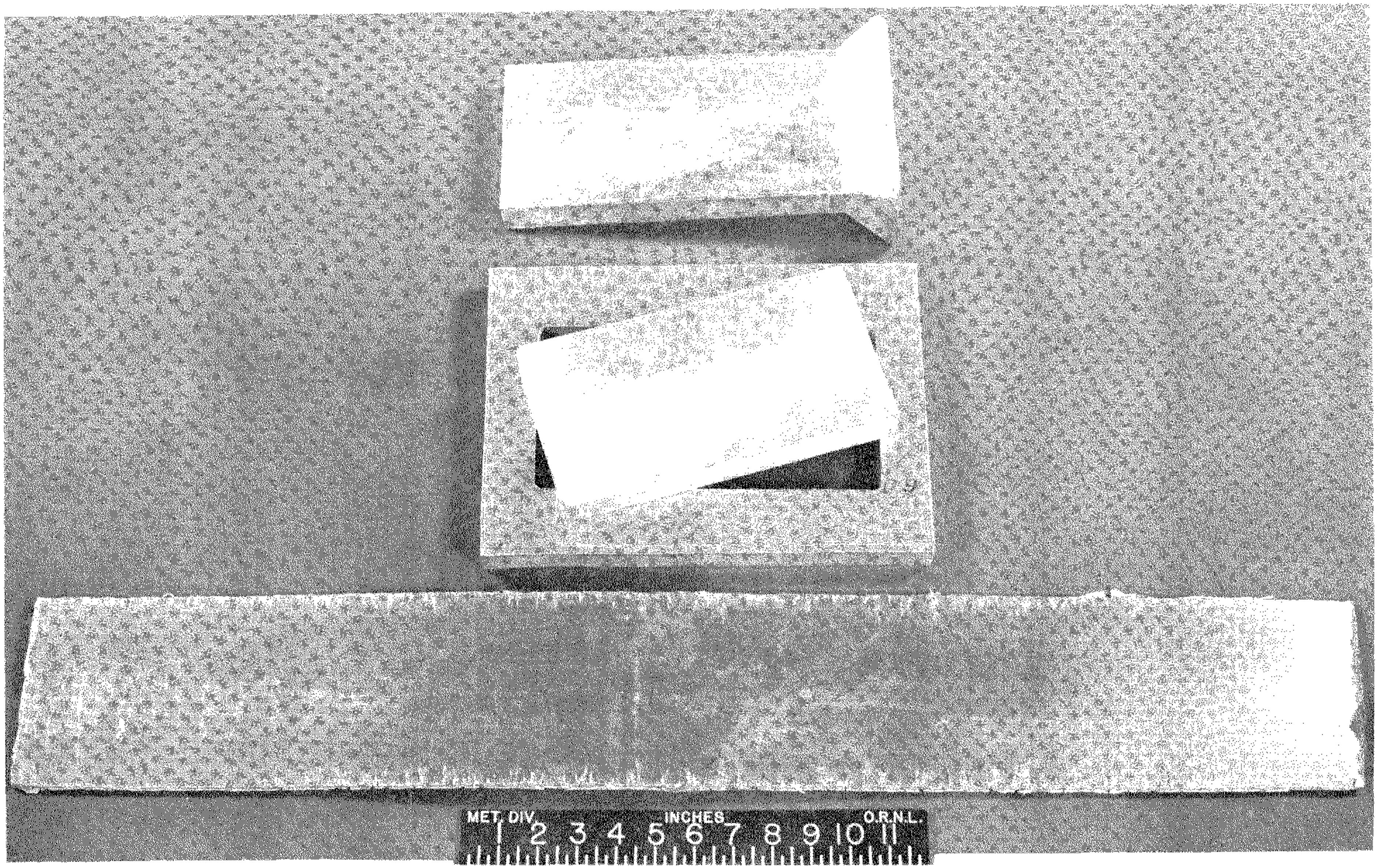
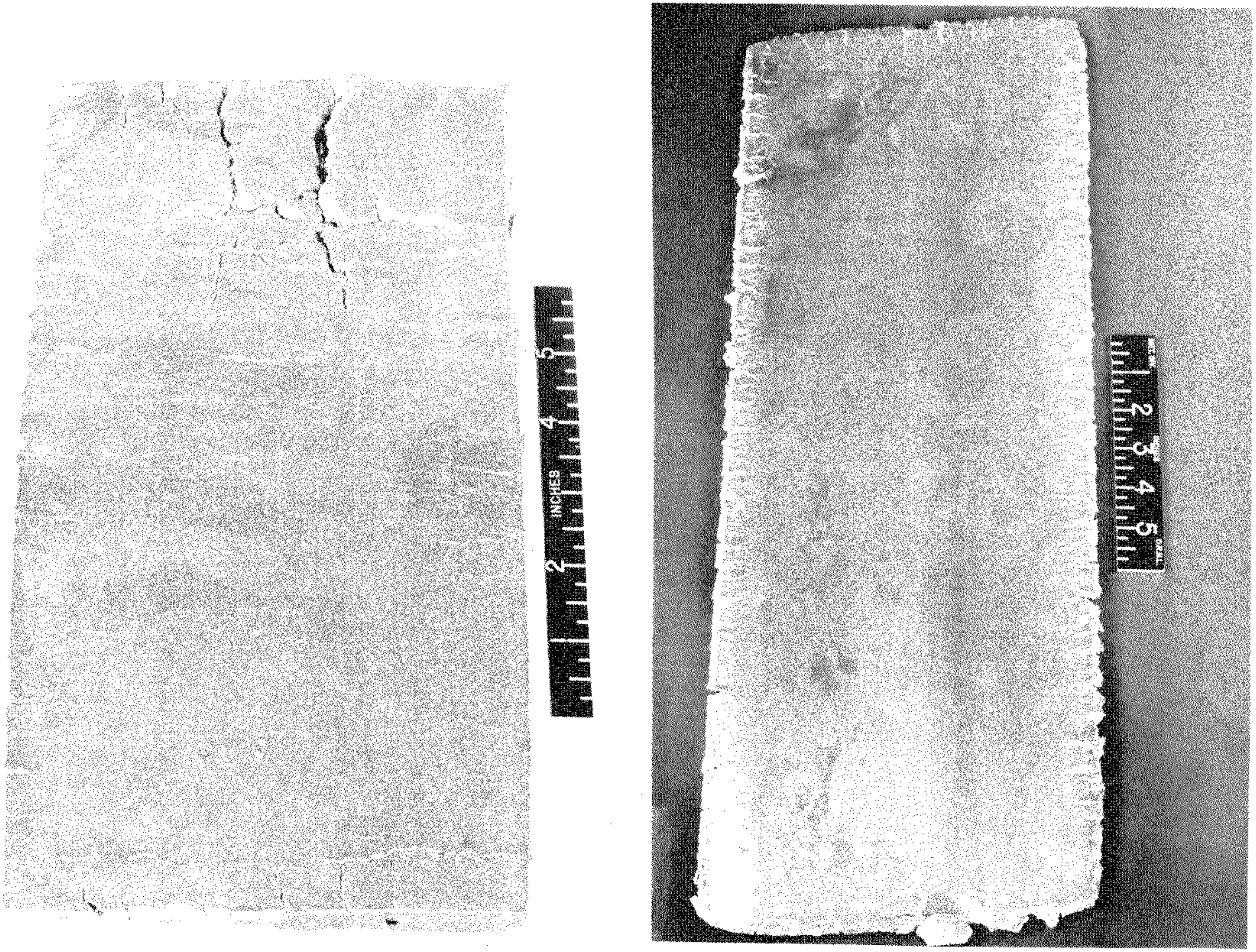


Fig. 8 (Y-21018) Procedure for Rolling 48 wt% U-Al Alloy Castings into Plate.
Top: Cast alloy - 1 in. thick.
Middle: Cropped and machined alloy with aluminum picture frame.
Bottom: Rolled alloy with frame removed - 1/4 in. thick.



48 wt.% U-Al

48 wt.% U - 3 wt.% Si - Al

Fig. 9 (Y-22362) Unmodified 48 wt% U-Al Alloy (left) Shows Virtual Disintegration After a Reduction in Thickness of Less than 10% . Silicon-modified 48wt% U-Al Alloy (right) successfully rolled to a 75% reduction in thickness with only limited edge cracking.

relatively "clean" microstructure with a reduced volume percentage of intermetallic compound. The microstructures shown in Fig. 10 (Y-22090, -22091) clearly reveal the improvements associated with silicon additions.

Several usable hot-rolling schedules were evolved during the development programs, each requiring some cross rolling to improve the material yield. One of the most satisfactory rolling schedules is listed below:

1st pass - 15% reduction
2nd pass - 20% reduction, cross roll
3rd pass - 20% reduction
4th pass - 20% reduction
5th pass - 20% reduction
6th pass - 20% reduction
Finish to 0.225

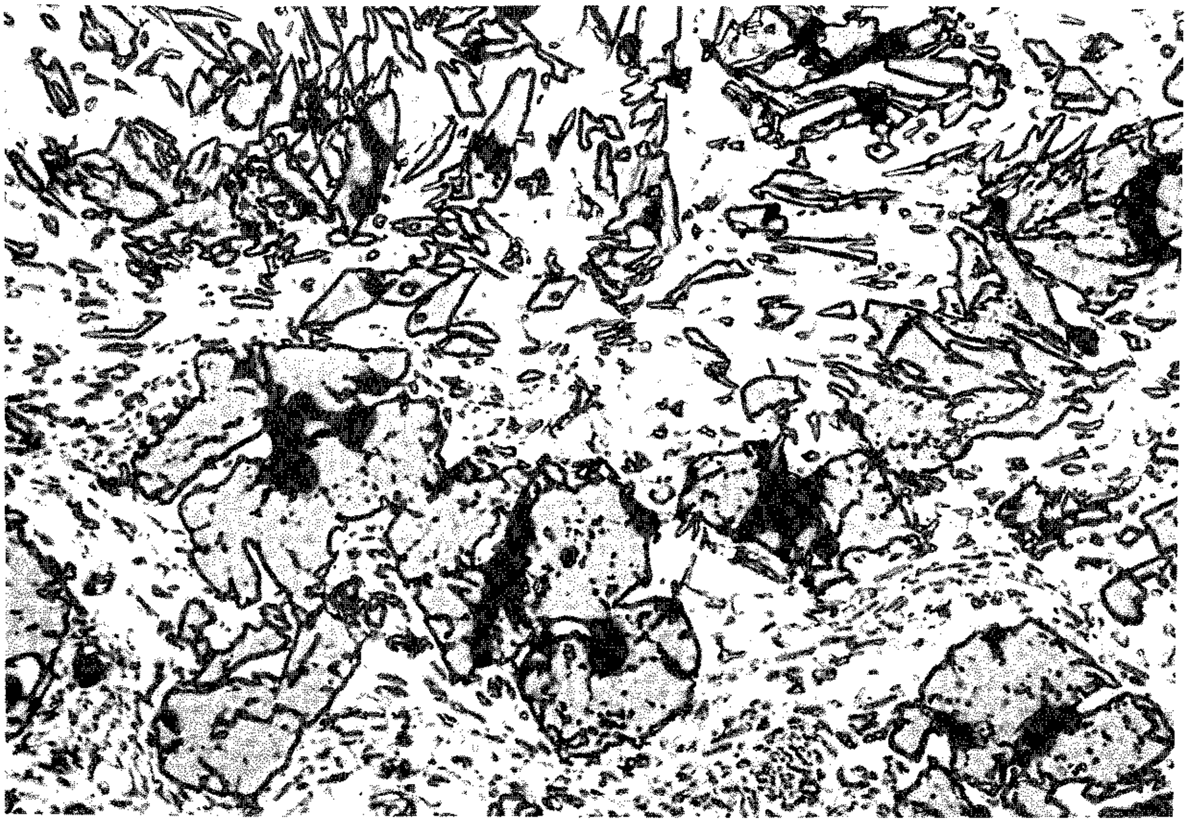
Fabrication by Hot Extrusion

During the course of this development program, a very cursory investigation of extrusion as a primary fabrication technique was conducted. It was found that cast cylindrical billets could be readily extruded into rods and comparison of the silicon-bearing and silicon-free alloys indicated that the initial pressures were reduced by about 30% when extruding the ternary alloy from a 3-in.-dia billet into a 3/4-in.-dia rod. Typical extrusion conditions are listed as follows:

Capacity of extrusion press	- 700 tons
Billet compositon	- 48 wt% U - 3 wt% Si-Al
Billet size	- 3 in. dia x 6 in long
Billet temperature	- 600°C
Lubricant	- Fused PbO and Necrolene
Die size	- 3/4 in. dia with 45 deg core
Die temperature	- 300°C
Container temperature	- 300°C
Dummy block	- Brass at room temperature
Starting pressure	- 55 tsi
Running pressure	- 54 tsi

The mechanical properties of an alloy rod prepared by this technique are subsequently discussed in "Property Evaluation" of this report.

Efforts to extrude 3-in.-dia billets into strips 1/2 in. thick and 2-1/2 in. wide were not successful. However, it is felt that with more extensive die development and a higher capacity press, strips can be extruded from which fuel cores for composite plate fabrication can be directly blanked.



Y-22090 48% U-Al Alloy

Y-22091 48% U-3% Si-Al Alloy

Fig. 10 Microstructures of Unmodified and Silicon Modified 48% U-Al Alloys after 75% Reduction in Thickness by Hot Rolling at 600°C

COMPOSITE FUEL PLATE FABRICATION

One of the major problems encountered with high-uranium-content alloys is the undesirable end effects in composite fuel plates fabricated by the conventional picture-frame technique. These end effects result from an appreciable difference in the yield strength between the fuel alloy and the aluminum canning materials at the rolling temperature. A typical example of these "dogbone" end effects, shown in Fig. 11 (Y-16359), reveals the significant clad thinning at the ends of the fuel core.

To determine the materials combination which would be most compatible at temperatures encountered during hot rolling, tensile tests were conducted at 550°C on aluminum alloys 6951, 1100, 5050, 5052, 5154, and 6061 as well as 48 wt% U-Al alloys containing 0, 1, 2, and 3 wt% Si. Results of these studies are graphically documented in Fig. 12 (ORNL-LR-DWG 27090). It can be seen that the yield strengths of 5154 Al and the reference 3 wt% Si alloy are nearly identical at 550°C, indicating that these metals represent the optimum selection. To substantiate this conjecture, composite fuel plates were prepared using this and several other materials combinations. During the investigation, it was found that complete metallurgical bonding between mating surfaces of the high-strength aluminum alloys was difficult to achieve with the limited reduction in thickness previously established for roll cladding 1100 Al composites. It was, therefore, necessary to resort to alcladding of the high-strength aluminum so that during composite fuel plate fabrication bonding was confined to 1100 Al surfaces.

Samples were sheared from the rolled plates and the magnitude of "dogboning" was determined by metallographic measurements. The results of these investigations are summarized in Table III. The data in this table are divided into three groups with the main variable in each group being, respectively: (1) the effect of frame material on "dogboning" in 60-mil-thick plates with 1100 Al cladding and 48 wt% U - 3 wt% Si-Al cores, (2) the effect of frame material on "dogboning" in 50-mil-thick plates with 1100 Al cladding and 50 wt% U - 3 wt% Si-Al cores, and (3) the effect of varying amounts of silicon in a nominal 48 wt% U-Al core alloy on "dogboning" in 60-mil-thick composite fuel plates with alclad 6061 Al covers and frames. Additional

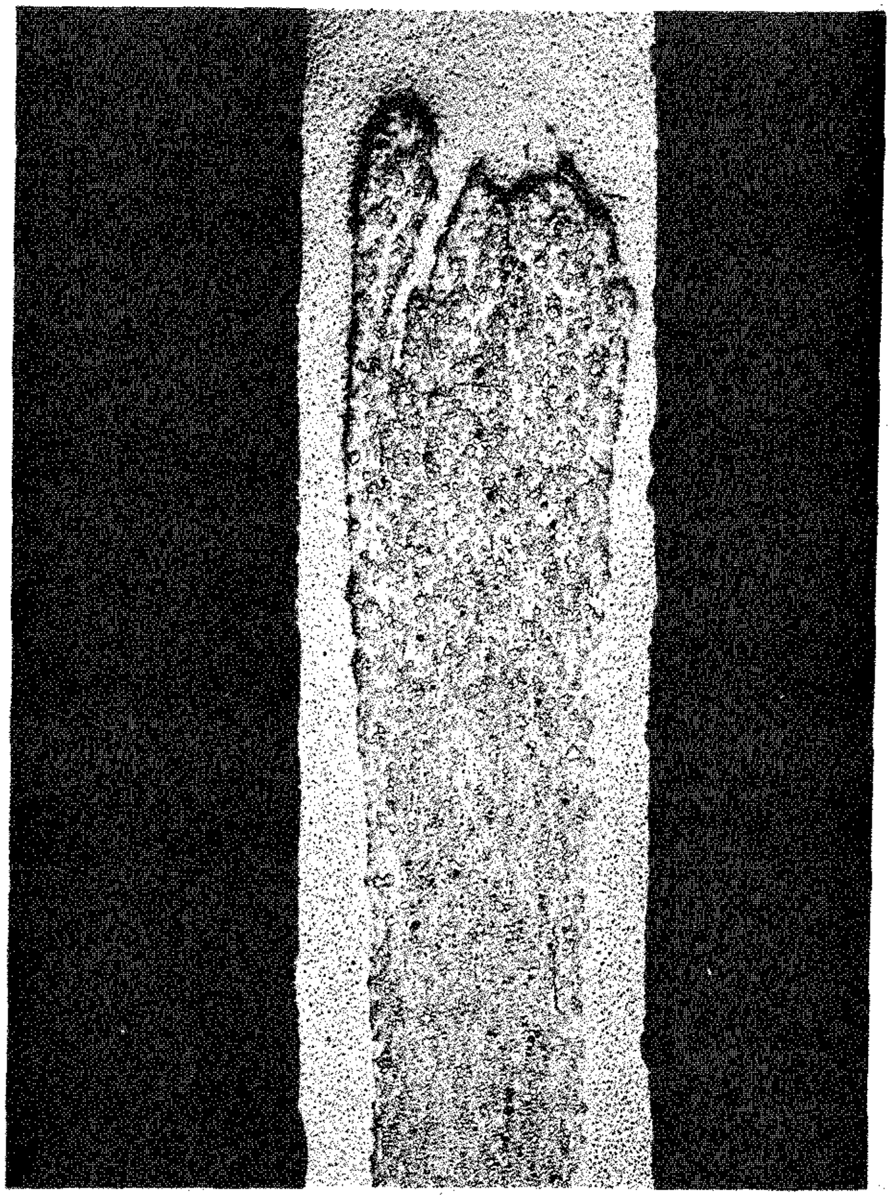


Fig. 11 (Y-16359) "Dogbone" in Composite Fuel Plate with 48 wt% U-Al Alloy Core and 1100 Al Frame and Cover Plate. Etchant: 2% HF aqueous solution. 300X.

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ORNL-LR-DWG 27090

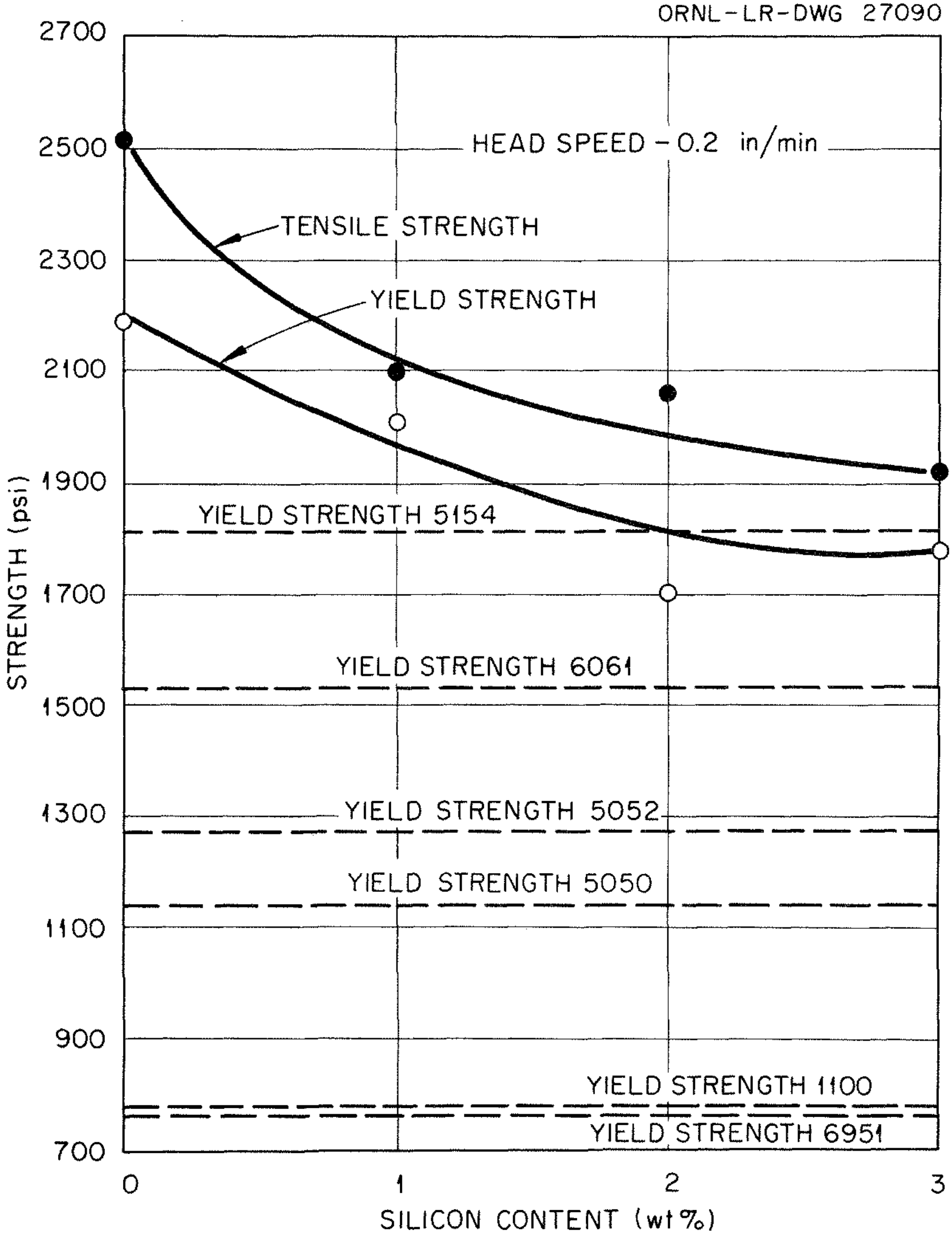


Fig. 12 Effect of Silicon Addition on Tensile Properties at 550°C of Nominal 48% U-Al Alloys.

TABLE III

ANALYSIS OF DOGBONING IN COMPOSITE FUEL PLATES

Lot Code	Frame Material[a]	Cover Material	Core Material	Plate Thickness (mils)	Number of Plate Ends Examined[b]	Statistical Measurement of Minimum Cladding Thickness ($\bar{x}\pm s$) (mils)[c]	Lowest Observed Value of Cladding Thickness (mils)	Reduction of Clad Thickness in Dogbone Area (%)	Clad-Core-Clad Ratio (mils)
GROUP I									
A	1100 (10 in. wide)	1100	48 U-3 Si-Al	60	17	7.5±2.8	1	59	17-26-17
B	1100(4.5 in. wide)[e]	1100	48 U-3 Si-Al	60	10	8.2±1.2	6	53	
C	5050	1100	48 U-3 Si-Al	60	12	8.2±2.4	3	53	
D	Alclad 5050	Alclad 5050	48 U-3 Si-Al	60	20	8.4±1.6	5	53	
E	Alclad 5050[e]	1100	48 U-3 Si-Al	60	14	9.5±2.5	5	41	
F	6061	1100	48 U-3 Si-Al	60	12	9.7±1.4	7	41	
G	Alclad 5154	1100	48 U-3 Si-Al	60	14	15.5±1.6	9	12	
GROUP II									
H	1100	1100	50 U-3 Si-Al	50	10	4.4±1.5	1	69	13-24-13
I	Alclad 5050	1100	50 U-3 Si-Al	50	8	5.2±2.2	3	62	
J	Alclad 5052	1100	50 U-3 Si-Al	50	20	7.9±1.7	4	38	
K	Alclad 5154[f]	1100	50 U-3 Si-Al	50	20	10.7±1.6	7	15	
L	Alclad 5154[f]	1100	50 U-3 Si-Al	50	10	10.1±2.1	5	23	
GROUP III									
M	Alclad 6061[g]	Alclad 6061	48 U-Al	60	8	5.6±1.2	3	65	17-26-17
N	Alclad 6061	Alclad 6061	48 U-1 Si-Al	60	9	6.2±1.7	1	65	
O	Alclad 6061	Alclad 6061	48 U-2 Si-Al	60	10	6.1±1.8	3	65	
P	Alclad 6061[g]	Alclad 6061	48 U-3 Si-Al	60	10	7.6±2.1	3	53	

[a] Alcladding was nominally 7.7% of the frame thickness.

[b] Two measurements per end.

[c] $\bar{x}\pm s$ is mean ± estimated standard deviation where $s = \sqrt{\frac{\Sigma\ (x-\bar{x})^2}{n-1}}$

[d] $\frac{\text{Nominal clad thickness - Average clad thickness} \times 100}{\text{Nominal clad thickness}}$

[e] At 95% confidence level, difference between Lots B and E is significant.

[f] Plates in Lot K rolled on 20-in. x 30-in. two-high mill. Plates in Lot L rolled on 12-in. x 14-in. two-high mill. Differences between Lots K and L significant at 75 – 80% confidence level.

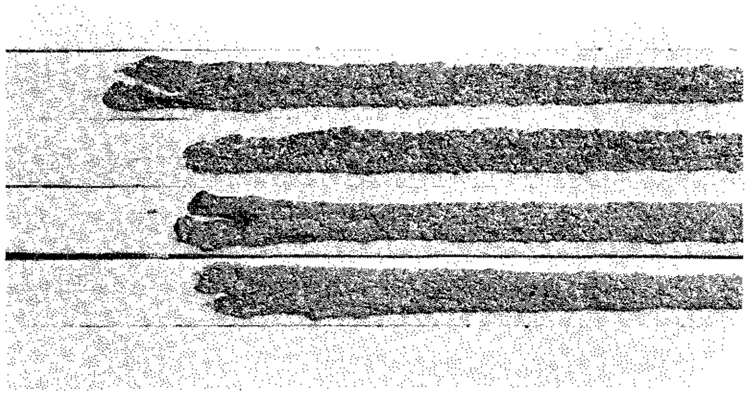
[g] At 99% confidence level, difference between Lots M and P is significant.

information was also obtained on the effect of mill size, frame width, and cover plate material on "dogboning." Results for Groups I and II are tabulated in order of increasing strength of the frame material. It can be seen that the minimum cladding thickness increases with increasing strength of the frame alloy, and the per cent reduction in cladding thickness is correspondingly diminished. Perusal of the data in Group I indicates that only slight improvements in end effects are realized by the replacement of 1100 Al frames with either 5050 or 6061 Al frames; however, selection of 5154 Al as the frame material results in a substantial improvement. This is consistent with the previously described yield-strength data. It can be further discerned from the data presented for Group I, by comparison of Lot A with Lot B, that increasing the width of 1100 Al picture frame from the reference width of 4.5 in. to a width of 10 in. results in further deterioration of the integrity of the fuel plate. Comparison of data for Lots D and E indicates that use of a higher strength alloy cover to replace the 1100 Al does not diminish the degree of "dogboning," suggesting that the strength of the frame is more important than the strength of the cover in properly constraining the core during plastic deformation.

From examination of Group II data, it can be seen that the increased frame strengths associated with increasing magnesium contents for the 5000-series alloys produce major improvements in the "dogbone" defect. The substitution of 5154 Al resulted in a reduction of merely 15% from the nominal cladding thickness compared to a reduction of 69% from the nominal with 1100 Al frames. Figure 13 (Y-23604, -24131, -23605) illustrates the comparative end effects in plates from Lots I, J, and K in which the composites were framed in 5050, 5052, and 5154 Al, respectively.

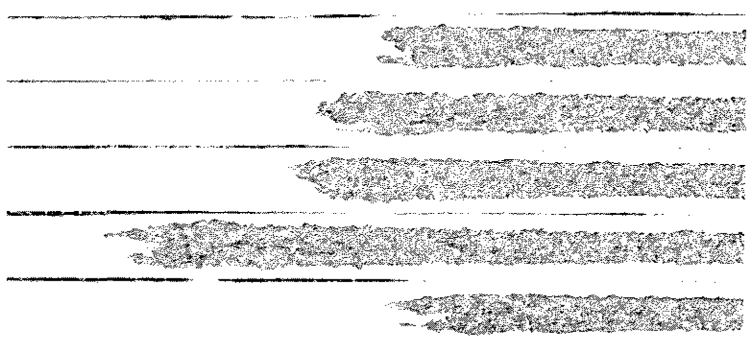
Statistical evaluation of the data for Lots K and L in Group II reveals that the difference in magnitude of "dogboning" is significant only at about the 75% confidence level. This suggests that only very slight alterations in the end effects result from changing the angle of nip of the two-high mill.

Examination of the data presented for Group III plates reveals that replacement of a 48 wt% U-Al alloy core with a ternary alloy core containing 3 wt% Si produces a slight but nevertheless statistically significant reduction in "dogboning" when the frame and cladding materials are held constant. The



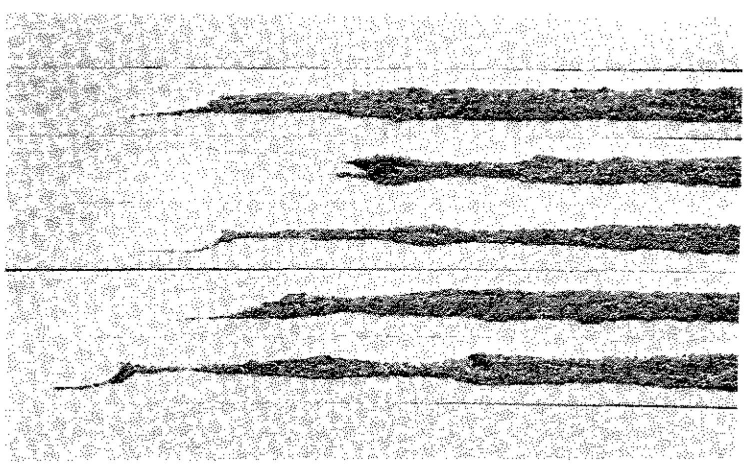
Alclad 5050 Frames
Min. Clad Thick -
5 mils (avg.)

(Y-23604) 7X



Alclad 5052 Frames
Min. Clad Thick -
8 mils (avg.)

(Y-24131) 7X



Alclad 5154 Frames
Min. Clad Thick -
11 mils (avg.)

(Y-23605) 7X

Fig. 13 Longitudinal Sections of Composite 50-Mil-Thick Fuel Plates Containing a 50 Wt% U-3 WT% Si-Al Fuel Alloy Clad with 1100 Al Showing Effect of Increasing Strength of Frame on "Dogboning". As Polished.

data are also consistent with the previously described differential yield-strength theory illustrated in Fig. 12. Visual evidence of this improvement can be seen in Fig. 14 (Y-21951), which includes representative samples of plates taken from Lots M, N, O, and P.

Nondestructive techniques have been developed at ORNL[10] for determining the cladding thickness at its thinnest point by use of eddy currents. It is recommended that 100% eddy-current inspection of the ends of fuel plates be conducted during manufacturing if materials combinations conducive to "dogboning" are employed in composite fuel plate manufacture.

FUEL ELEMENT ASSEMBLY

MTR-type fuel elements containing fully enriched uranium are conventionally assembled by brazing the composite fuel plates between grooved aluminum side plates using an aluminum-silicon alloy of eutectic composition. Current specifications require that all plate spacings be held to within ± 10% of the nominal dimension.

Attempts to apply similar assembly techniques to elements containing high uranium loadings indicated that maintaining the required plate spacings during brazing was an extremely difficult task, because the fuel plates tended to warp and distort non-uniformly during the brazing cycle. As shown in Fig. 15 (ORNL-LR-DWG 26316), the coefficient of thermal expansion of the silicon-modified 48 wt% U-Al alloy is markedly different from that of 1100 Al and an 18.5 wt% U-Al alloy. It is felt that this property of the high-uranium-content fuel alloy is responsible for the distortion of the components during the brazing cycle.

A total of fourteen fuel elements were brazed during the course of this development program in which plate width, depth of side-plate groove, cooling cycle, and fuel unit design were varied. All fourteen of these units contained out-of-tolerance plate spacings, as indicated from the data summarized in Table IV.

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[10] J. W. Allen, R. A. Nance, and R. B. Oliver, Eddy-Current Measurement of Clad Thickness, ASTM Special Tech. Pub. No. 223 (March 1958).

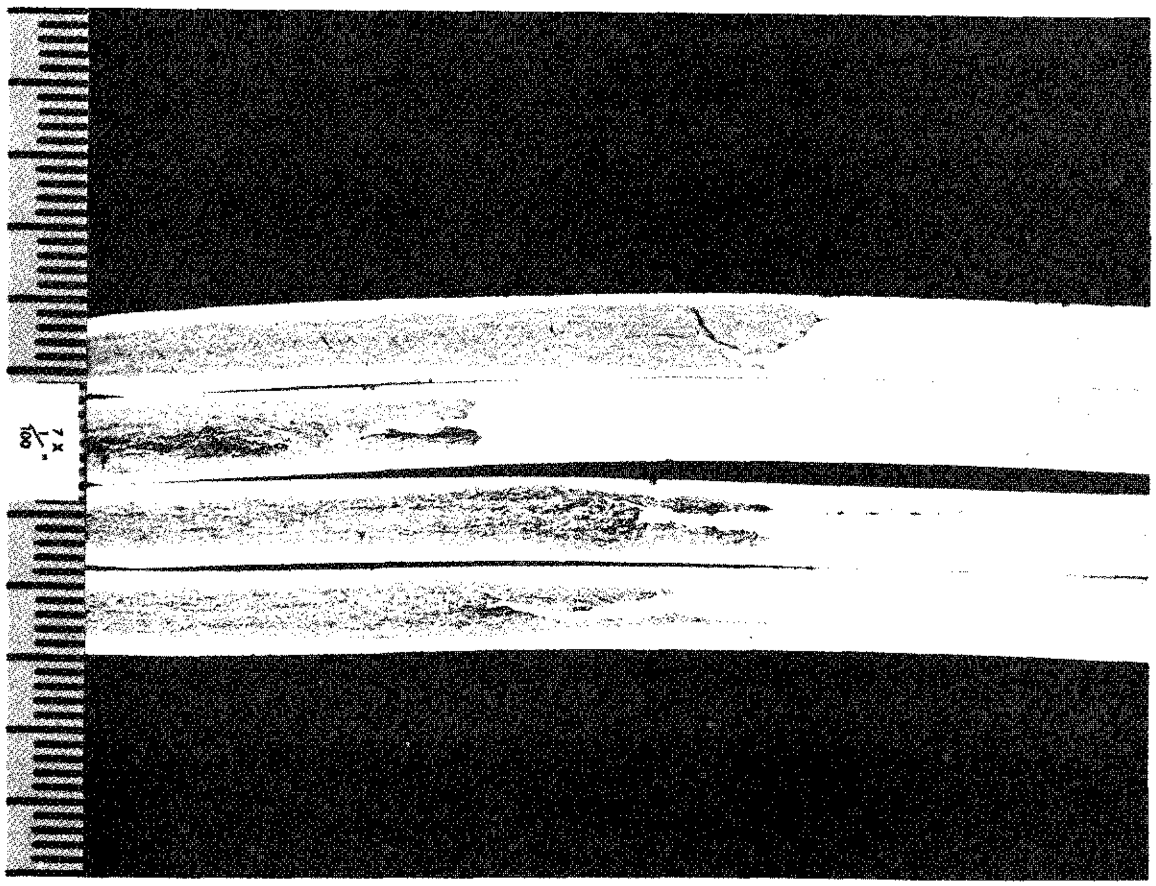

Fig. 14 (Y-21951) Longitudinal Sections of Composite Fuel Plates Showing Effect of Silicon Additions to 48 wt% U-Al Alloy on "Dogboning" in Plates Framed and Clad with Alclad 6061 Al. As-polished. 7X.

Plate	Core Composition	Min. Cladding Thickness
Top	48 wt% U-Al	5,7 mils
2nd	48 wt% U - 1 wt% Si-Al	5,5 mils
3rd	48 wt% U - 2 wt% Si-Al	6,7 mils
Bottom	48 wt% U - 3 wt% Si-Al	9,10 mils

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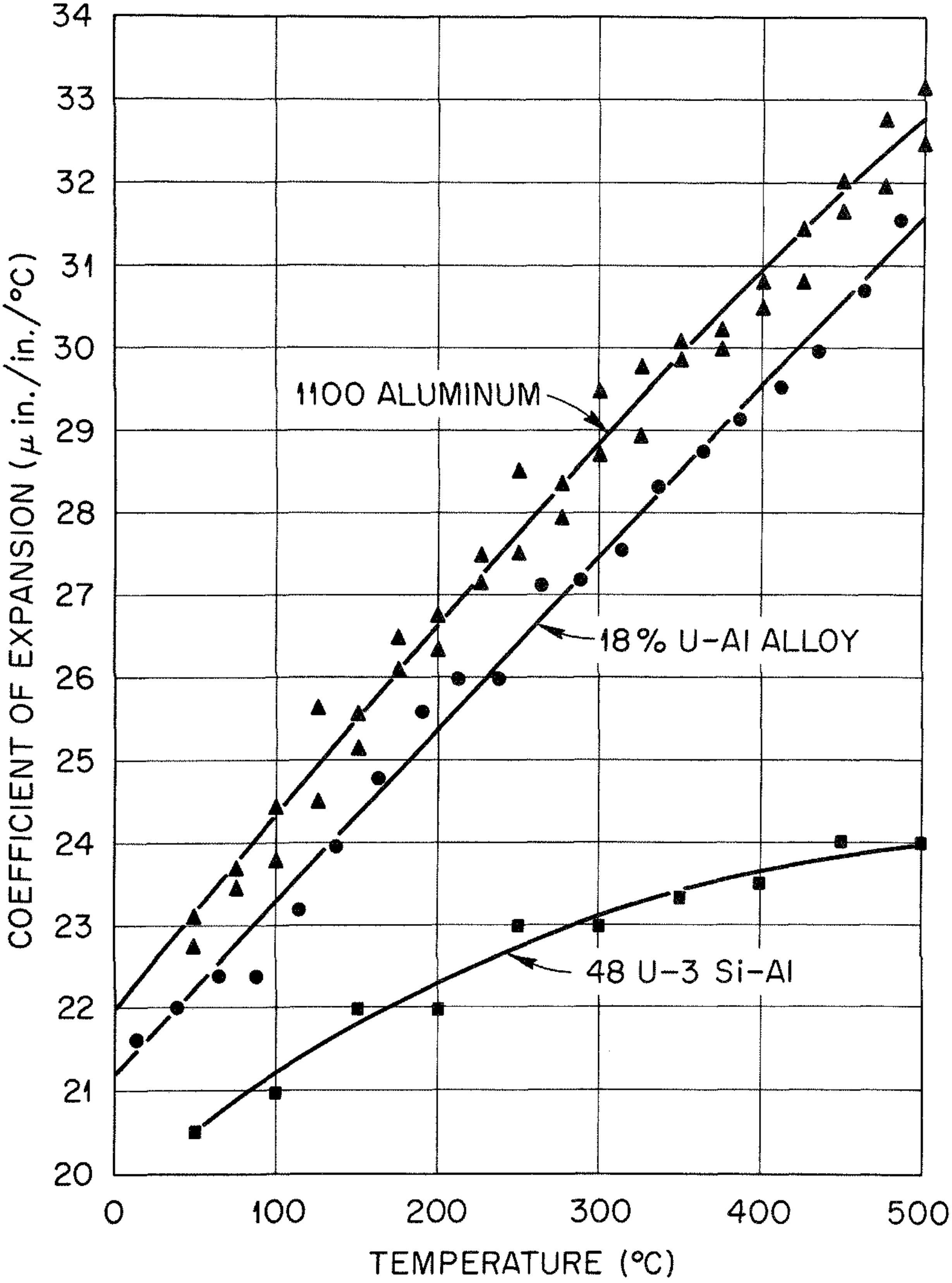

Fig. 15 Coefficient of Linear Thermal Expansion of 1100 Al and U-Al Alloys.

TABLE IV

EFFECT OF VARIABLES ON BRAZING OF FUEL ELEMENTS WITH NOMINAL PLATE SPACING OF 117 MILS

Element Designation	Type of Fuel Unit	Core Alloy (Wt %)	Frame Material	Cover Material	Nominal Clad-Core-Clad Ratio (mils)	As-Machined Plate Width (in.)	No. of Out-of-Tolerance Plate Spacing Measurements*	Maximum Plate Spacing (in.)	Minimum Plate Spacing (in.)	Comments on Brazing
FRD-10	18 Plate-Mark X (modified)	48 U-3 Si-Al	5050	1100	17-26-17	2.845	20	0.148	0.088	Standard brazing conditions and plate sizes
FRD-11	18 Plate-Mark X (modified)	48 U-3 Si-Al	5050	1100	17-26-17	2.835	11	0.135	0.086	Same as FRD-10 but with 10-mil narrower plates
FRD-12	18 Plate-Mark X (modified)	48 U-3 Si-Al	Alclad 5050	1100	17-26-17	2.830	4		0.098	Same as FRD-10 but with modified side plates
FRD-13	18 Plate-Mark X (modified)	48 U-3 Si-Al	Alclad 5050	1100	17-26-17	2.830	4	0.134	0.098	Same as OR-40-1 but allowed to furnace-cool from brazing temperature
FRD-16	18 Plate-Mark X (modified)	48 U-3 Si-Al	Alclad 5050	1100	17-26-17	2.820	4	0.134	0.103	Same as FRD-10 but with 25-mil narrower plates
FRD-17	18 Plate-Mark X (modified)	48 U-3 SiAl	Alclad 5050	1100	17-26-17	2.820	2	0.133	0.101	Same as FRD-16
FRD-18	18 Plate-Mark X (modified)	48 U-3 Si-Al	Alclad 5154	1100	17-26-17	2.820	10	0.135	0.097	Same as FRD-16 but with 5154 frame
FRD-19	18 Plate-Mark X (modified)	48 U-3 Si-Al	Alclad 5154	1100	17-26-17	2.820	1		0.100	Same as FRD-18
OR-40-1	18 Plate-Mark X (modified)	48 U-3 Si-Al	Alclad 5050	1100	17-26-17	2.830	12	0.144	0.088	Same as FRD-10 but with 15-mil narrower plates
OR-40-2	18 Plate-Mark X (modified)	48 U-3 Si-Al	Alclad 5154	1100	17-26-17	2.820	12	0.140	0.082	Same as FRD-18
FRD-14	19 Plate-Mark XI	50 U-3 Si-Al	Alclad 5154	1100	13-24-13	2.802	7	0.132	0.098	Standard brazing conditions and plate sizes
FRD-15	19 Plate-Mark XI	50 U-3 Si-Al	Alclad 5154	1100	13-24-13	2.802	23	0.150	0.065	Same as FRD-14
FRD-20	19 Plate-Mark XI	50 U-3 Si-Al	Alclad 5154	1100	13-24-13	2.780	17	0.155	0.086	Same as FRD-14 but with 22-mil narrower plates
FRD-21	19 Plate-Mark XI	50 U-3 Si-Al	Alclad 5154	1100	13-24-13	2.825	31	0.157	0.062	Same as FRD-20 but with modified side plates and fuel plate widths

*68 or 72 measurements made per element.

Two alternative approaches exist in circumventing the brazing problems. One course is to analyze the heat transfer requirements of the fuel element for a given reactor application and, if possible, specify less stringent tolerances. Examination of the data in Table IV for the ten elements containing eighteen 60-mil-thick fuel plates reveals that all plate spacings were within ± 30% of the nominal. It is very probable that such elements could be used for pool reactor applications. The data obtained from the brazing of 19-plate elements reveal a significantly larger number of out-of-tolerance plate spacing measurements, and in general, a greater deviation from nominal when compared to data accumulated on the 18-plate fuel elements. It is felt that the thinner plates in the 19-plate element are large responsible for this difference. It appears that in order to braze fuel elements within reasonably acceptable tolerances, it is more desirable to select a design which specifies less than 19 plates with individual plate thicknesses being at least 60 mils.

The second possibility is to employ mechanical joining instead of brazing. Several commercial fabricators of aluminum plate-type fuel elements, as well as ORNL, are currently developing this type of element and appraising its performance under reactor conditions. In addition to evaluation of the mechanical strength of the element, the effect of crevice corrosion, particularly for extensive periods in quiescent water, must be carefully determined.

PROPERTY EVALUATION

Mechanical Testing

Standard tensile specimens were selected to measure the tensile strength, yield strength, and elongation of nominal 48 wt% U-Al alloys with 0, 1, 2, and 3 wt% Si and 3 wt% Sn. The following comparisons were made from the test data which are compiled in Table V:

1. Effect of silicon additions on the tensile properties of a nominal 48 wt% U-Al alloy at room temperature.

2. Effect of silicon additions on the tensile properties of a nominal 48 wt% U-Al alloy at 550°C.

3. Comparison of room-temperature tensile properties of 48 wt% U - 3 wt% Si-Al alloys fabricated by hot rolling and by hot extrusion.

TABLE V

TENSILE PROPERTIES OF MODIFIED AND UNMODIFIED 48 wt % U-Al ALLOYS AT ROOM TEMPERATURE AND 550°C

Billet	Specimen	Test Temperature (°C)	Analyzed U Content (wt %)	Analyzed Si Content (wt %)	Analyzed Sn Content (wt %)	Tensile Strength (psi)	Yield Strength (0.2% Offset) (psi)	Elongation (2-in. Gage Length)	Specimen Condition
Si-1-S	C	R.T.	46.10	0.12		25,900	17,800	1.5	Hot rolled at 600°C with a 75% reduction in thickness
	D	R.T.	45.72	0.20		25,000	17,700	1.0**	
Si-2-S	A	R.T.	45.75	1.10		18,200	14,600	1.5	
	B	R.T.	45.56	1.14		18,800	14,500	1.0**	
Si-3-S	1	R.T.	46.85	2.05		17,200	12,400	3.5	
	2	R.T.	47.10	1.92		16,700	12,200	1.5**	
Si-4-S	1	R.T.	48.81	3.04		17,900	13,500	3.0	
	2	R.T.	48.64	2.99		15,800	11,900	2.0	
	3	R.T.	49.16	3.00		13,400	10,400	2.0	
	4	R.T.	49.32	2.88		13,600	11,400	1.0**	
Si-1-S	1-3	550	44.27	N.A.****		2,320	2,050	15***	
Si-16-S	16-1	550	45.88	N.A.		2,620	2,260	14	
	16-3	550	45.67	N.A.		2,460	2,100	22	
	16-4	550	46.93	N.A.		2,650	2,340	10	
Si-2-S	2-1	550	46.74	1.08		2,520	2,520	4	
	2-2	550	47.04	1.15		1,980	1,810	10	
	2-3	550	48.18	1.10		2,040	1,920	5	
	2-4	550	47.38	0.90		1,850	1,790	3	
Si-3-S	3-1	550	47.40	2.13		1,990	1,610	6	
	3-2	550	46.45	2.16		2,140	1,800	15	
Si-4-S	4-1	550	48.46	3.13		2,020	1,690	5	
	4-2	550	48.17	3.16		2,230	2,030	4	
Si-10-S	10-2	550	51.74	3.28		2,010	1,890	5	
	10-3	550	47.66	3.02		1,790	1,780	5	
	10-4	550	47.00	2.89		1,560	1,500	5	
FR-12	E-1	R.T.	48.39	2.97		12,770	8,700	3.0	Extruded at 600°C with a 94% reduction in area
	E-2	R.T.	48.61	3.02		12,880	9,540	3.0	
	E-3	R.T.	48.84	2.97		13,900	10,000	2.5	
Si-26-S	1	550	49.22		2.98	1,940	1,820	6	Hot rolled at 600°C with a 75% reduction in thickness
	2	550	47.52		3.06	1,840	1,740	6	
	3	550	48.18		3.66	1,920	1,700	5	
	4	550	46.35		2.06	1,690	1,540	6.5	
	5	550	47.96		2.79	1,930	1,780	7	
Si-26-S	6	R.T.	46.81		2.75	16,220	12,220	3	
	7	R.T.	48.57		2.94	16,100	12,540	2.5	
	8	R.T.	49.36		3.44	13,970	10,730	1.0	
	9	R.T.	48.47		2.79	15,520	11,620	1.5	
	10	R.T.	47.39		3.17	16,580	11,210	3.5	

*R.T. – room temperature.
**Specimen broke outside gage length.
***Specimen cracked in two places.
****Not Analyzed – no silicon addition made.

4. Comparison of the tensile properties at room temperature and 550°C for nominal 48 wt% U-Al alloys in which UAl_4 suppression was achieved with 3 wt% Si and 3 wt% Sn.

Data illustrating the effect of silicon addition on the room temperature tensile and yield strengths of a nominal 48 wt% U-Al alloys are graphically presented in Fig. 16 (ORNL-LR-DWG 20518). The yield strength can be seen to decrease by 34% and the tensile strength by 40% from the unmodified alloy to the 3 wt% Si alloy. The decrease in strength can be attributed to the decreasing amounts of dispersed intermetallic compound associated with increasing silicon contents and more complete UAl_4 suppression.

The data in Fig. 16 were replotted as a function of the ratio $\frac{UAl_3}{UAl_3+UAl_4}$, determined from x-ray spectrometer patterns and are summarized in Fig. 17 (ORNL-LR-DWG 27088R). These data show that the decrease in yield strength is closely related to the quantity of UAl_4 suppressed. For example, the addition of 1 wt% Si to the fuel alloy suppresses 75% of the UAl_4, and correspondingly there is a 25% decrease in yield strength. The most significant decrease in yield strength and greatest amount of UAl_4 suppression per wt% Si occur in this 1 wt% Si alloy. Additional silicon produces less percentage suppression of the UAl_4 phase,and the attendant percentage decreases in yield strength, per wt% Si,are similarly smaller.

Comparison of the strength data for the 48 wt% U - 3 wt% Si-Al and 48 wt% U - 3 wt% Sn-Al yields the following average results:

	Room Temperature		550°C	
	Tensile Strength (psi)	Yield Strength (psi)	Tensile Strength (psi)	Yield Strength (psi)
3 wt% Si	15,200	11,800	1920	1780
3 wt% Sn	15,680	11,660	1860	1720

These data indicate the close correlation in tensile properties for alloys in which UAl_4 suppression is achieved with two different ternary additions. The results from the tin-bearing alloy are also plotted in Fig. 17 and, within the accuracy of the x-ray measurements, reasonable agreement with the silicon-bearing alloys is revealed substantiating the postulate that loss in strength is associated with decreasing amounts of the dispersed phase.

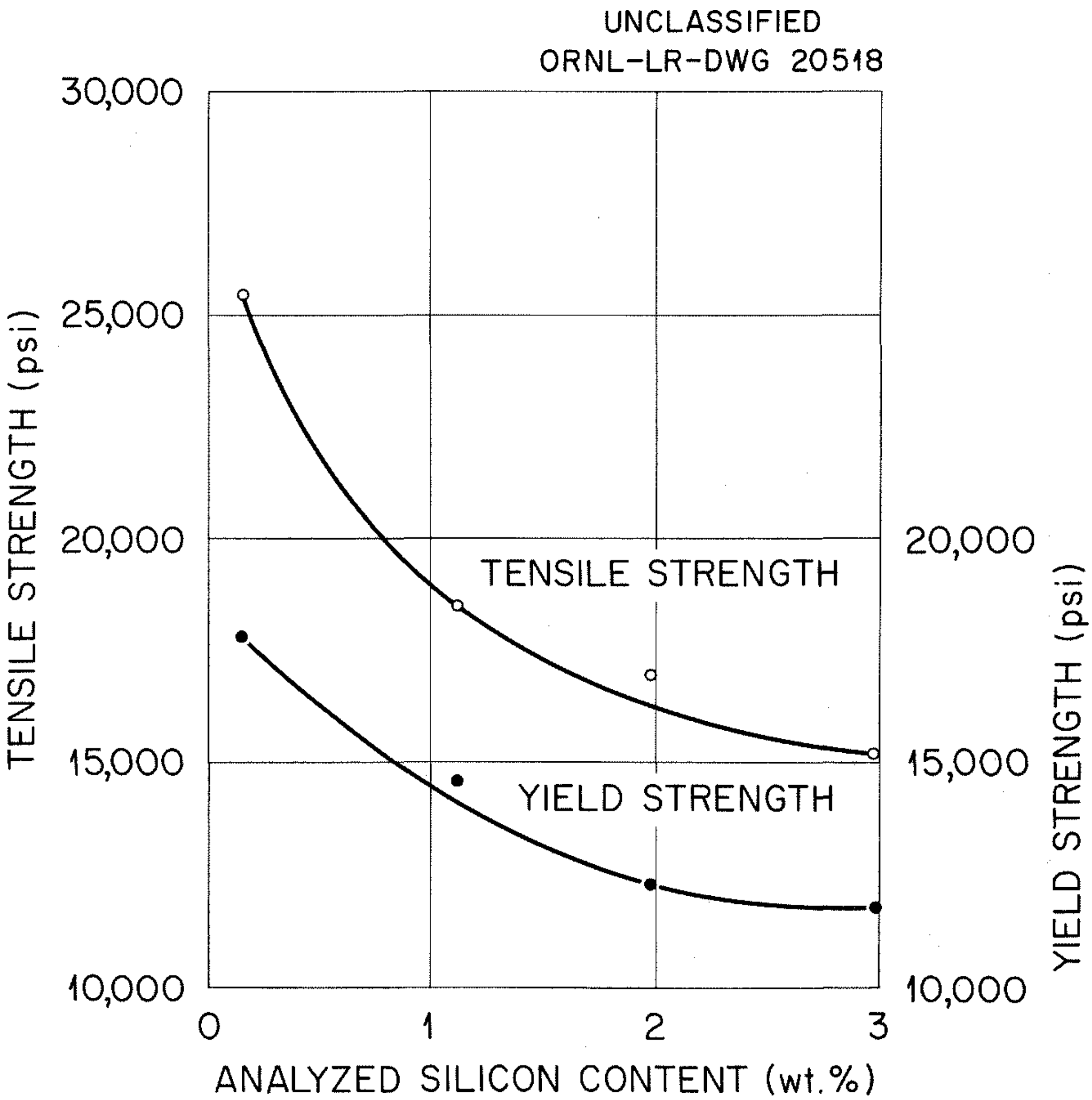

Fig. 16 Effect of Silicon on the Room Temperature Tensile Properties of 48 wt.% Uranium-Aluminum Alloys.

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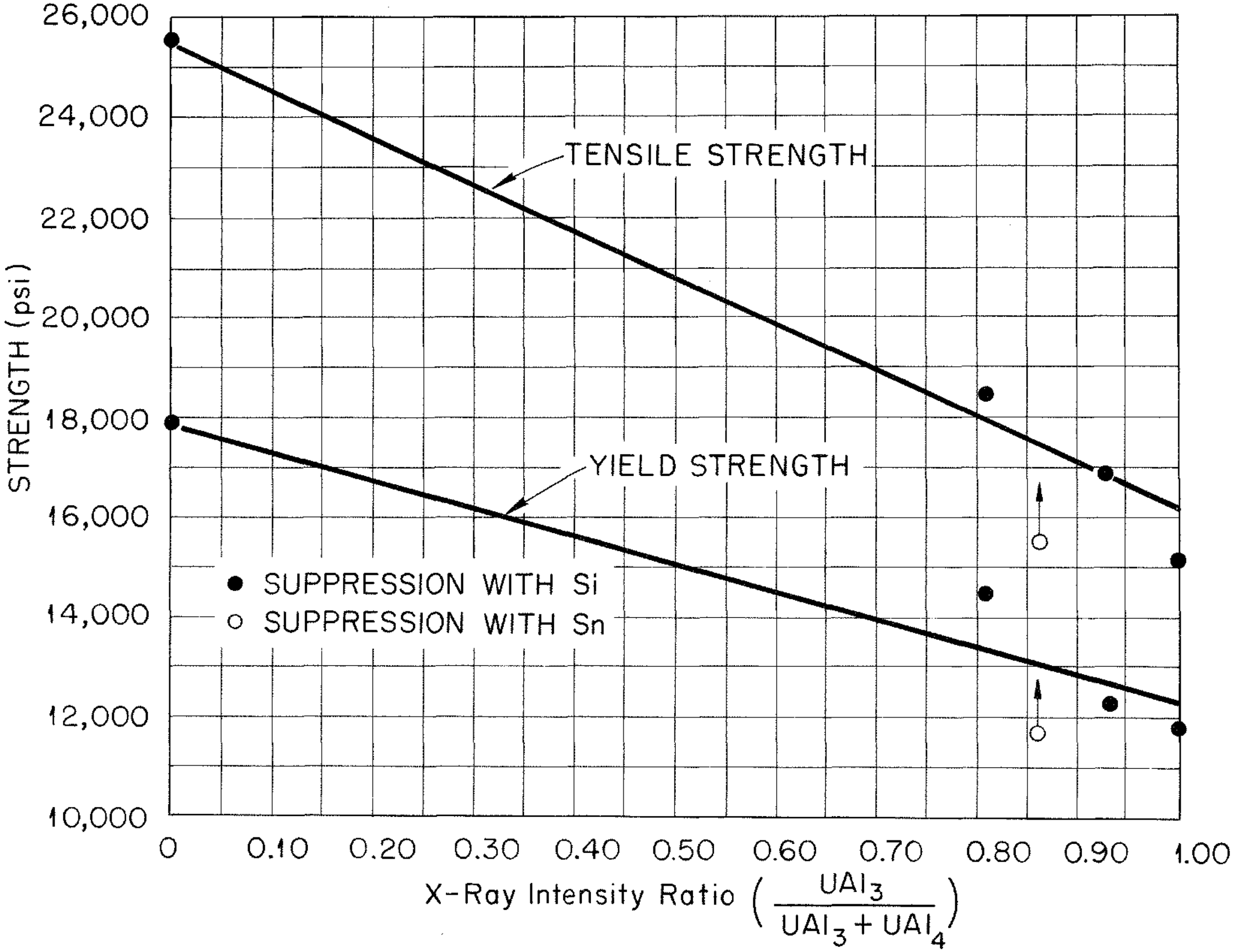

Fig. 17. Effects of UAl_4 Suppression on Room Temperature Tensile Properties of 48 wt % U-Al Alloys.

Information describing the effect of silicon on the tensile properties of the 48 wt% U alloy at 550°C was previously discussed in "Composite Fuel Plate Fabrication" of this report and plotted in Fig. 12. Tests performed at this temperature on alloys containing 3 wt% Si showed decreases in tensile strength and yield strength of 24% and 19%, respectively, over the unmodified alloy.

The average room temperature values calculated from data listed in Table V show that the 3 wt% silicon-modified alloy, reduced 75% by hot rolling at 600°C, has tensile and yield strengths which are ~ 15% higher than the same alloy reduced 94% by extrusion at 600°C. This information indicates that extruded alloys may be superior to hot-rolled material as stock for fuel cores to be rolled into composite plate.

Irradiation Testing

Irradiation testing in the MTR of two full-size fuel elements containing nominal 48 wt% U-Al fuel alloys (with 20% enriched uranium) to burn-ups of 25%[11] and 62%[12] of the U-235 atoms indicated that no observable radiation-induced effects resulted from these exposures. Consequently, it was felt that silicon-modified alloys would exhibit similar stability under irradiation; a confidence which has been subsequently borne out by a demonstration test in the MTR of a full core loading of fuel elements, manufactured by Babcock and Wilcox Company, which contained a 45 wt% U-Al alloy modified with 3 wt% Si.[13]

However, it was felt that a quantitative inspection of silicon-modified alloy specimens irradiated over a wide range of burn-ups would be desirable.[14] For this evaluation, a series of miniature fuel plates were prepared. The miniature plates were 6 in. long, 1 in. wide, and 0.060 in. thick and contained a fuel core 5 in. long, 0.70 in. wide, and 0.026 in. thick. The frames on these plates were alclad 5050 Al, the covers 1100 Al and the core a nominal 48 wt% U - 3 wt% Si-Al alloy enriched 19.20% in the U-235 isotope. Pertinent data on the dimensions and calculated fuel loadings of these plates are

[11] W. C. Thurber, J. H. Erwin, and R. J. Beaver, The Application of a Nominal 48 wt% U-Al Alloy to Plate-Type Aluminum Research Reactor Fuel Elements, ORNL-2351 (February, 1958).

[12] Unpublished ORNL information.

[13] N. Jesson, Babcock and Wilcox Company, private communication.

[14] C. F. Leitten, Jr. and W. C. Thurber, Phase I, Foreign Reactor Fuel Sample Irradiation, ORNL CF 58-2-109 (Revised) (October, 1958).

included in Table VI. A radiograph of typical plates is shown in Fig. 18 (Y-26323). These miniature plates are fabricated using procedures identical to those employed in the manufacture of full-size composite fuel plates.

The miniature plates were inserted into leaky rabbit capsules and eight of the plates divided into groups of two for duplicate determinations are presently being irradiated in MTR process water to burn-ups of 20, 40, 60, and 80% of the U-235 atoms. A typical specimen in a leaky rabbit capsule is shown in Fig. 19 (Y-26348).

Chemical Reprocessing

A limited study by the Chemical Technology Division was directed toward the problems arising in chemical reprocessing of fuel alloys containing silicon, since it is known that emulsions of collodial silica adversely affect pulsed column operation. For this purpose, a series of samples of 48 wt% U - 3 wt% Si-Al alloys clad with 1100 Al and framed in 5050 Al were submitted to the Chemical Technology Division for evaluation. Their studies[15] indicated that:

1. The 3 – 4 hr feed digestion specified in the original flow sheet[16] had essentially no effect on removal of soluble silica.

2. Digestion at high (4 M) acid conditions was only slightly effective in reducing emulsification of the silica.

3. Dehydration at 150 – 160°C resulted in phase disengagement times of about one minute, which are probably satisfactory for column operation.

Studies are continuing which may result in replacement of the high temperature dehydration with a less rigorous treatment.

Pneumatic Pressure Testing

To determine the amount of fuel plate deflection and permanent set which would result from hydraulic pressure differentials during in-pile operation of these high-uranium-content fuel elements, three fuel elements were tested using air pressure to simulate hydraulic loading conditions. The elements were

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[15] Chemical Technology Division Monthly Prog. Rep., June 1957, ORNL-2362.

[16] J. R. Flanary, J. H. Goode, A. H. Kibbey, J. T. Roberts, and R. G. Wymer, ORNL-1993.

TABLE VI

PERTINENT DATA ON URANIUM-ALUMINUM ALLOY MINIATURE FUEL PLATES FOR MTR IRRADIATION

Core Number	Finished Core Length (In.)*	Finished Core Width (In.)**	Finished Core Area (cm^2)	Core Weight (g)	U Content of Core (g)***	U-235 Content of Core (g)****	U-235 Surface Density (g/cm^2)
OR-35-A-1	4.906	0.688	21.77	6.588	3.207	0.616	0.028
OR-35-A-2	4.938	0.688	21.91	6.519	3.173	0.609	0.028
OR-35-A-3	4.906	0.688	21.77	6.479	3.154	0.606	0.028
OR-35-A-4	4.938	0.688	21.91	6.619	3.222	0.619	0.028
OR-35-A-5	4.906	0.688	21.77	6.516	3.172	0.609	0.028
OR-35-A-6	4.969	0.688	22.05	6.510	3.169	0.608	0.028
OR-35-A-7	4.938	0.688	21.91	6.668	3.246	0.623	0.028
OR-35-A-8	4.969	0.688	22.05	6.514	3.171	0.609	0.028
OR-38-A-9	4.875	0.688	21.63	6.637	3.231	0.620	0.029
OR-35-A-10	4.938	0.688	21.91	6.619	3,222	0.619	0.028

* ± 0.031 in. (from radiographs).
** ± 0.016 in. (from radiographs).
*** Based on average of three samples from heat D-668.
**** Based on an enrichment of 19.20% in U-235.

pressurized internally and the deflection of the bottom fuel plate was measured using the experimental rig shown in Fig. 20 (Y-18619). The unit was then depressurized and the permanent set associated with a particular pneumatic loading was determined.

Pertinent specifications on the three units tested are included in Table VII. One unit was a standard MTR 19-plate unit (Mark XI) while the other two were 18-plate elements with 48 wt% - 3 wt% Si-Al cores and 5050 Al or 5154 Al frames. Figures 21 (ORNL-LR-DWG 30502) and 22 (ORNL-LR-DWG 30503) illustrate, respectively, the deflection and permanent set as a function of differential pressure for these elements. It can be seen that element FRD-19 is considerably more resistant to deformation than is the standard Mark XI element, indicating that this type of element should be entirely satisfactory in hydraulic environments similar to those existing in the MTR and other tank-type reactors.

Corrosion Testing

Specimens of a nominal 48 wt% U - 3 wt% Si-Al alloy, both clad and unclad, were corrosion tested in static, deionized water at 60 and 90°C for periods of 3, 6, and 12 weeks. The unclad specimens, each 1/2 in. wide and 2 in. long x 0.030 in. thick, were obtained by machining the cladding from composite fuel plates which had received an 85% reduction in thickness at 600°C after insertion into picture frames. The clad specimens were also taken from the same composite fuel plates, but in this case the fuel alloy was exposed only among the specimen perimeter and at the hanger hole. The major surfaces were protected by a roll-bonded cladding of 1100 Al. These specimens were 2 in. long, 1/2 in. wide, and 0.120 in. thick.

Evaluation of the extent of corrosion was accomplished primarily by visual examination and was supplemented by weight-gain measurements, solution pH, and resistivity determinations. No correlation could be made between either pH or resistivity measurements and the extent of corrosion. Weight gains on the order of 2 mg/cm^2 or less were observed for all specimens regardless of test temperature, duration of test, or specimen condition (i.e., clad or unclad).

Some blistering was observed under most of the test conditions but the extent of blistering appeared to be independent of exposure conditions. In fact, in some cases the over-all performance appeared to be superior in

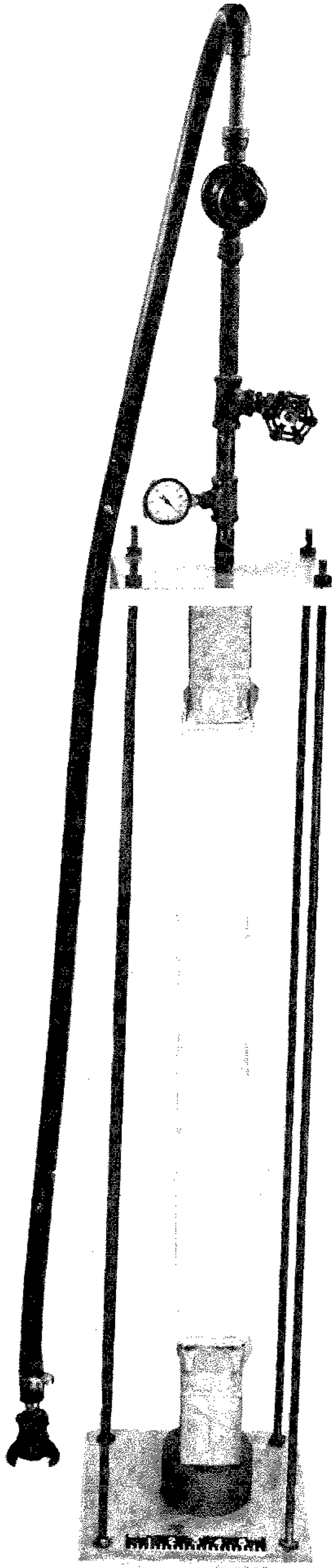

Fig. 20 (Y-18619) Pneumatic Pressure Testing Rig for Aluminum Fuel Elements.

TABLE VII

SPECIFICATIONS ON FUEL ELEMENTS FOR DIFFERENTIAL PRESSURE TESTS

	Side Plates						Clad-	
Element	Material	Thickness (in.)	Frame Material	Cladding Material	Core Alloy	Plate Thickness (mils)	Clad-Core-Clad Ratio (mils)	Number of Plates per Element
Mark XI	1100 Al	3/16	1100 Al	1100 Al	18% U-Al	50*	15-20-15	19
FRD-10	1100 Al	3/16	5050 Al	1100 Al	48% U-3% Si-Al	60	17-26-17	18
FRD-19	1100 Al	3/16	Alclad 5154 Al	1100 Al	48% U-3% Si-Al	60	17-26-17	18

*Top and bottom plates 65 mils thick with 17.5-20-17.5 clad-core-clad ratio. Deflection of bottom plate was actually measured.

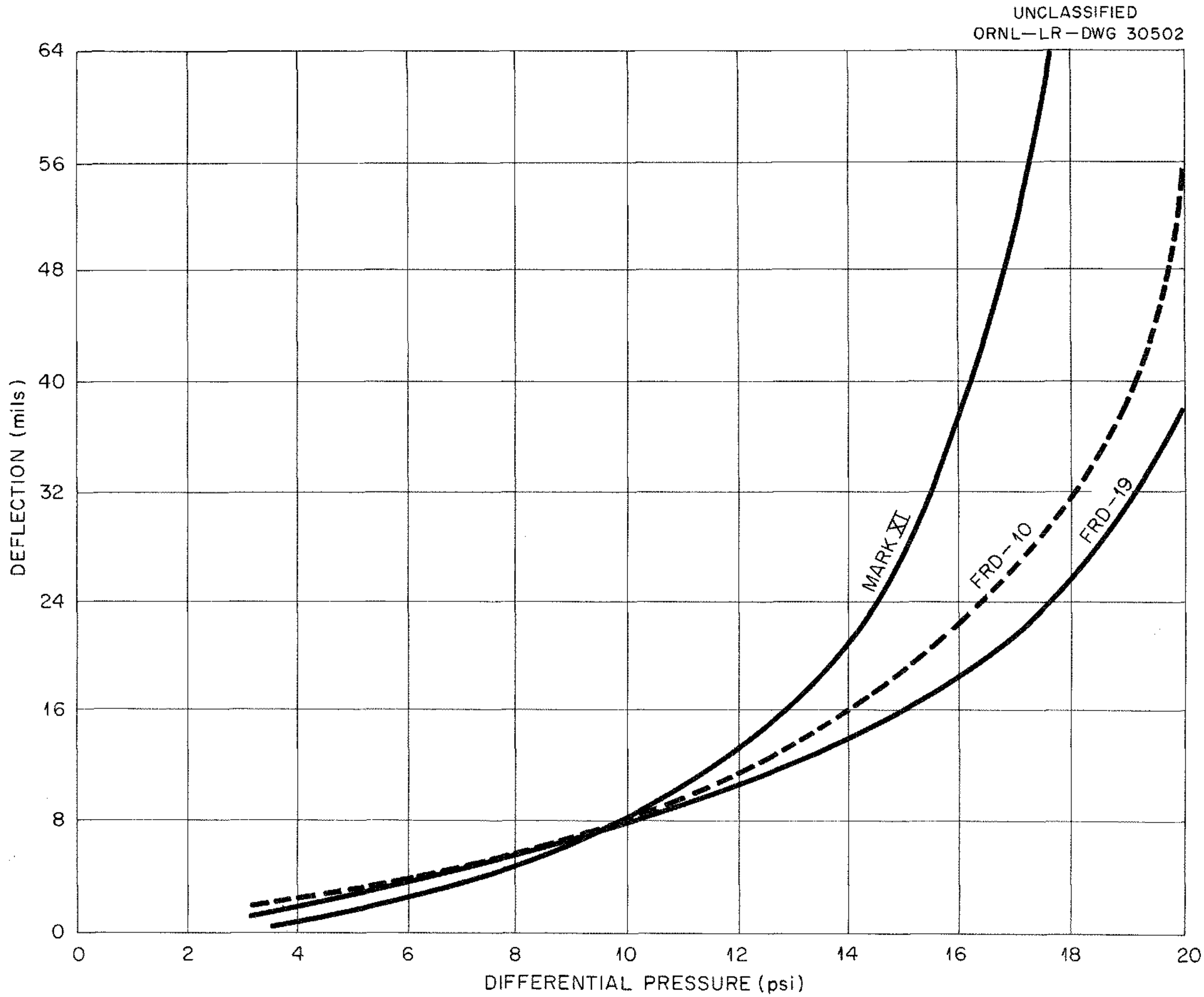

Fig. 21 . Deflection of Curved-Plate Aluminum Fuel Elements Under Differential Pressure.

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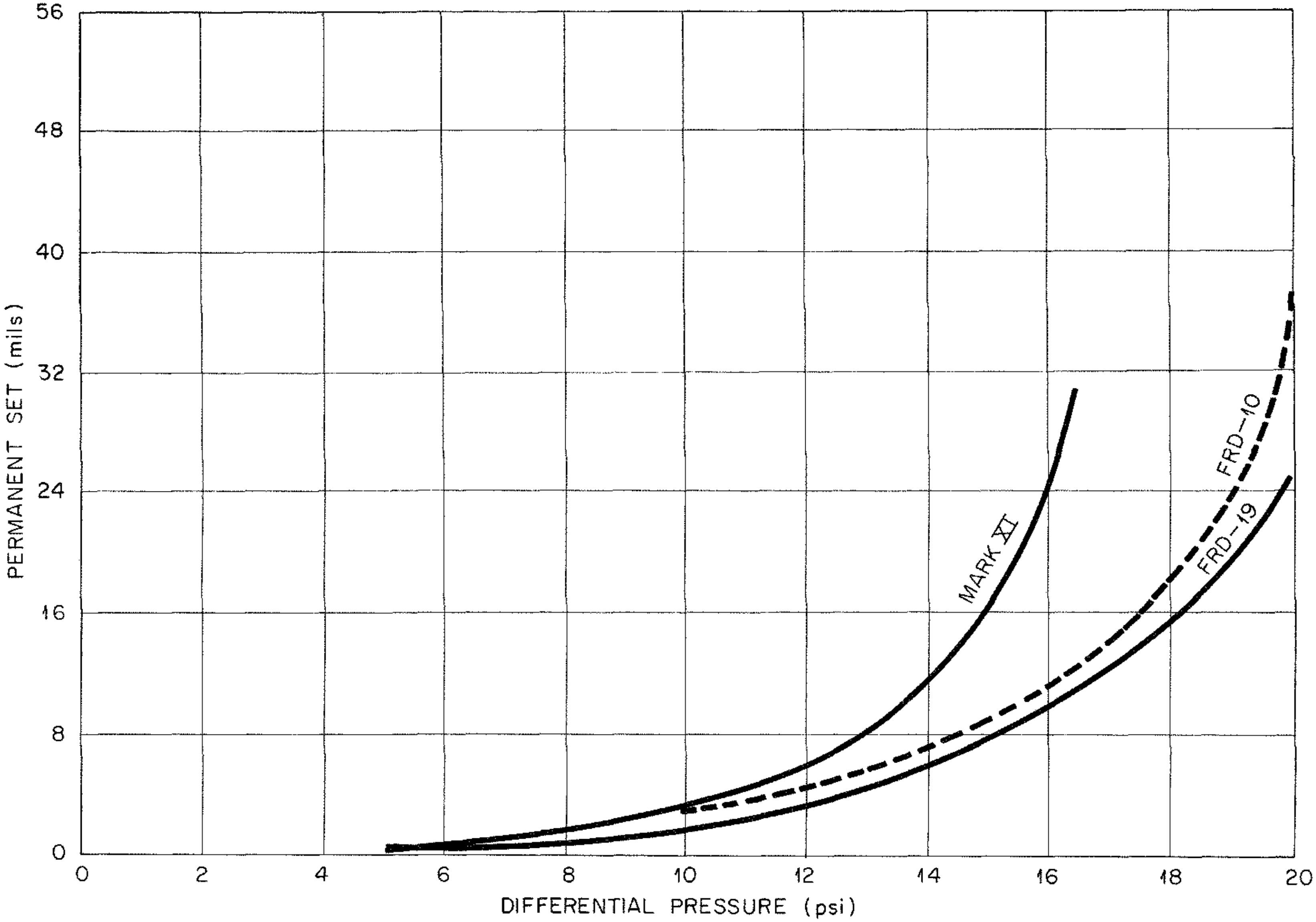

Fig. 22. Permanent Distortion of Curved-Plate Aluminum Fuel Elements Under Differential Pressure.

specimens tested for 12 weeks to those tested for 3 weeks. These apparent inconsistencies probably can be attributed to the complex nature of the corrosion system.

Figure 23 (Y-12 Corrosion Plate 3074-2) shows the surface appearance of an unclad specimen after 6 weeks exposure in 60°C water. The surface blisters are quite obvious. These blisters may possibly have nucleated at sites of carbide inclusions or other homogeneities. Figure 24 (T-15159) pictures the generation of a blister in the core of a clad specimen exposed to 90°C water for 12 weeks.

In general, it can be stated that the corrosion of 48 wt% U - 3 wt% Si-Al alloy appears to be characterized by very little general surface attack. However, some blisters do form in the exposed fuel alloy. Tests of these alloys are continuing to exposures of 26 weeks under the same conditions.

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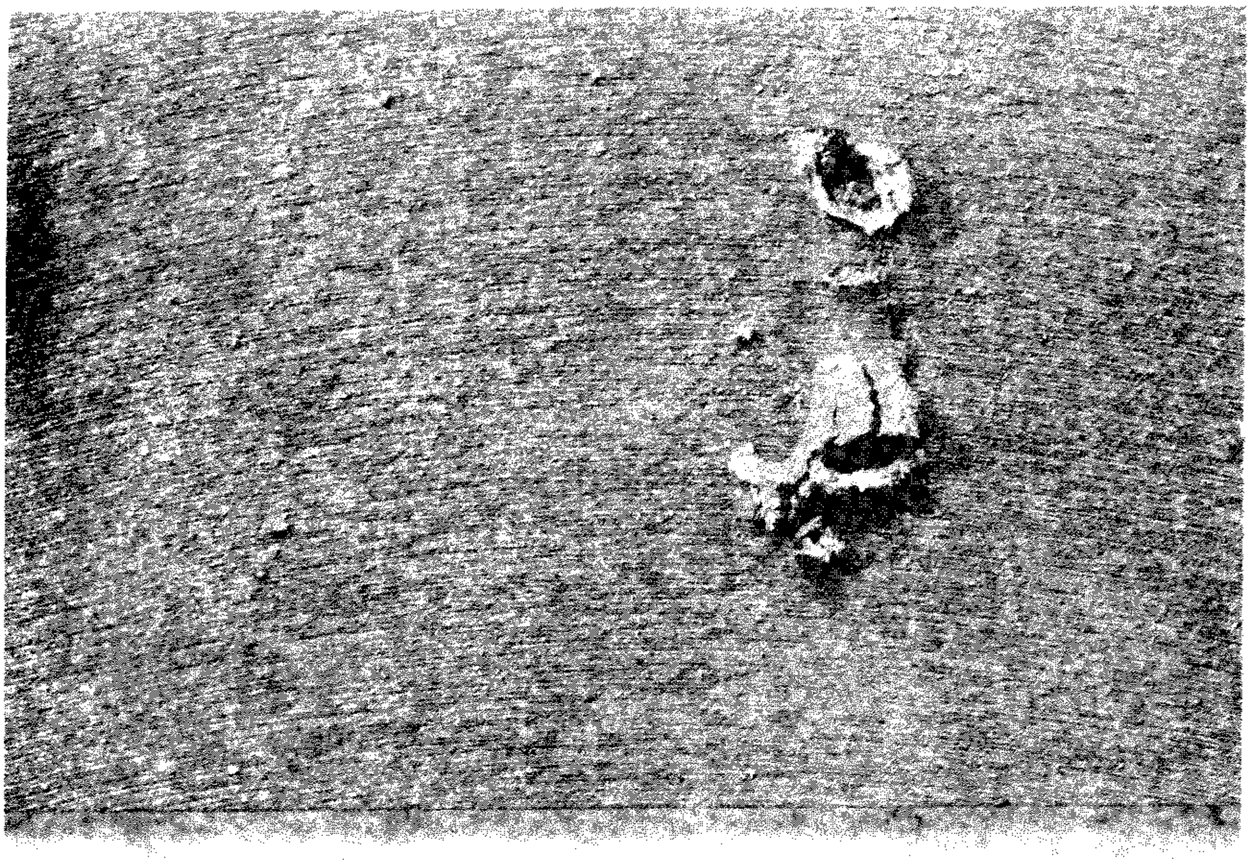

Fig. 23 (Y-12 Corrosion Plate 3074-2) Blisters on Surface of 48 wt% U - 3 wt% Si-Al Corrosion Specimen Exposed in Static 60°C Water for 6 Weeks. As-machined surface. 10X.

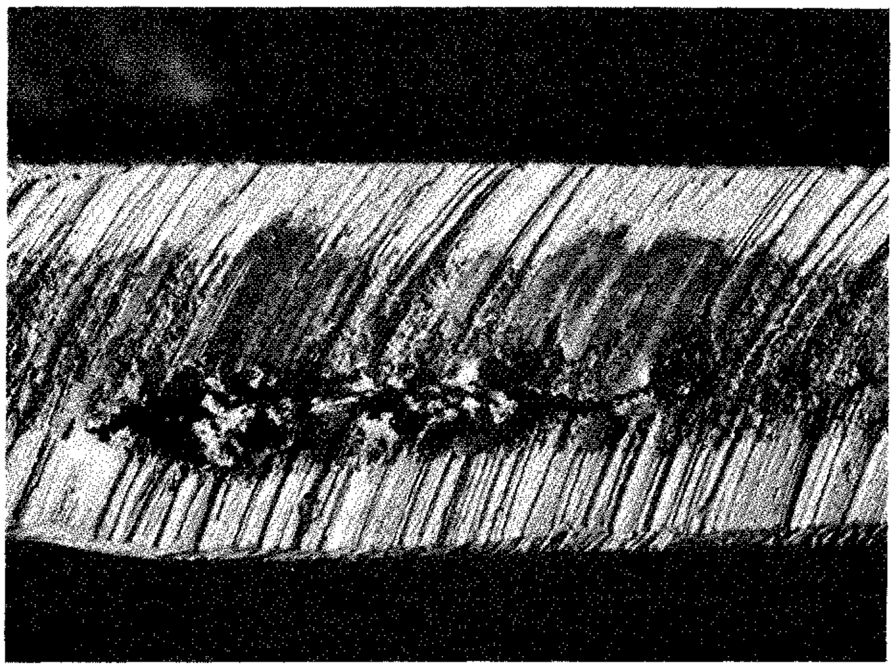

Fig. 24 (T-15159) Cross Section of 1100 Alclad 48 wt% U - 3 wt% Si-Al Alloy Showing Blister in Core at Exposed Area. Specimen treated in 90°C water for 12 weeks. As-machined. 16X.

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