ATL-A-104
INFORMAL AEC RESEARCH
AND DEVELOPMENT REPORT
UC-25, Metallurgy and Ceramics
SPECIAL DISTRIBUTION

Semiannual Progress Report No. 1
DEVELOPMENT OF METALLIC URANIUM FUEL ELEMENTS
OF IMPROVED IRRADIATION STABILITY

ATL Job 21002
14 December 1959

Prepared Under
Contract No. AT(04-3)-250
Project Agreement No. 2
for the U. S. Atomic Energy Commission
Division of Reactor Development

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SUMMARY

The feasibility of improving swelling resistance in metallic uranium by increasing the dislocation density is under investigation. Increasing the density of dislocations is expected to 1) increase the number of sites at which fission product gas atoms are "pinned", 2) increase the number of gas bubble nuclei, and 3) increase mechanical strength. Dislocations are introduced by a treatment which involves deformation of metastable beta or gamma phase in uranium-rich alloys, followed by transformation.

Activity to date has been concerned primarily with selection of uranium-base alloys for initial evaluation, procurement of materials, installation of specialized equipment, and development of techniques. A number of small ingots have been produced, and screening tests have been carried out on several analyses. These preliminary experiments were designed to show the deformation characteristics of the alloys in the metastable state as a function of rolling temperature and time.

Initial results on gamma-stabilized binary alloys containing 2, 3.5, 5, and 7 w/o molybdenum have shown that large reductions are possible in the 5 and 7 w/o Mo alloys at 450°C. The more dilute alloys are relatively difficult to roll, although small reductions have been achieved. Aging experiments on deformed and undeformed 5 w/o Mo alloys indicate that deformation accelerates the aging process and retards growth of the "pearlitic" eutectoid transformation product.

Results with dilute chromium and chromium-molybdenum alloys have shown that rolling is possible in the metastable beta phase for a short time at temperatures as low as 450°C. The beta phase is quite hard at lower temperatures, however, and cracking occurs in some alloys at relatively small reductions. Beta phase deformation of these alloys has not resulted in significant hardness increase, although final grain size has been reduced appreciably.

Additional experimental work is required to complete the screening of uranium alloys for application to the current program. Following this activity, several of the most promising alloys will be selected for an intensive investigation of the effect of processing variables on mechanical properties, microstructure, and structural stability. Irradiation testing and post-irradiation evaluation of treated alloys will be undertaken later as Phase II of the over-all program.

INTRODUCTION

The useful life of metallic uranium fuel elements in power reactors is limited by the decrease in density or swelling which accompanies prolonged irradiation at elevated temperatures. It is generally agreed that most of this swelling is caused by the fission product gases krypton and xenon, which precipitate from supersaturated solid solution to form gas bubbles.

This is the first in a series of status reports covering research on the feasibility of improving swelling resistance in dilute uranium-base alloys by increasing the dislocation density. Increasing the number of dislocations is expected to improve irradiation stability by 1) increasing the number of sites at which gas atoms are "pinned" and, therefore, restrained from diffusing to bubble nuclei, 2) increasing the number of bubble nuclei so that a large number of sub-microscopic bubbles with insignificant total volume will be formed, and 3) increasing the mechanical strength of the alloy so that gross swelling will be retarded.

The pre-irradiation treatment under investigation is based on a recently developed process for producing ultra high strengths in low-alloy steels.^{1,2,3,4} In this process, metastable austenite is plastically deformed below the eutectoid temperature in the "bay" of the time-temperature-transformation diagram. Subsequent transformation of the "cold-worked" austenite results in a very fine grained martensitic or bainitic structure which has high strength and presumably a high dislocation density. Since several dilute uranium alloys possess transformation characteristics similar to those of the low-alloy steels (Appendix A), it is expected that "deformed-metastable-phase-transformation" treatments can be developed to increase the strength and dislocation density of these alloys.

In the initial phase of the program, objectives are limited to demonstrating the feasibility of producing improved irradiation-resistant uranium-base alloys by the deformed-metastable-phase-transformation treatment. This will include 1) development of laboratory techniques for elevated-temperature deformation and heat treatment of metastable gamma and beta uranium-base alloys; 2) investigation of out-of-pile effects of important variables (composition, deformation, heat treatment, etc.) on stability, microstructure, and physical properties; and 3) evaluation of swelling characteristics of the most promising alloys and treatments as a function of temperature and burnup.

1. See list of references appended.

Experimental procedures and current results are discussed in the following sections. Two constitution diagrams and a number of time-temperature-transformation curves taken from the literature on uranium-base alloys are reproduced in Appendix A.

EXPERIMENTAL WORK

All of the experimental work was conducted with hot rolled slabs produced from small arc-melted ingots. The slabs were heated to the beta or gamma temperature range in a molten salt bath, quenched to a lower temperature in molten lead, and then hot rolled isothermally at this temperature by rapidly returning the slabs to the lead bath between passes. In many cases the slabs were water-quenched and/or given a final heat treatment after the last pass. The experiments were evaluated by hardness tests and by visual and metallographic examination.

Results of the work accomplished to date are given in the next section. Details of the alloying, rolling, heat treating, and metallographic work are given below.

A. Charge Preparation

All of the alloys were prepared from a single 50-pound block of depleted, reactor-grade uranium. The block was milled into 1/2-inch slabs and then sawed into small pieces (approximately $1/2" \times 1/2" \times 3/4"$) suitable for preparing arc furnace charges. Each piece was wire-brushed to remove oxides and cleaned in trichloroethylene before weighing.

Molybdenum additions were made from arc-melted high-purity rods (99.95+%) supplied by Cleveland Tungsten Co. The rods were sanded to remove oxides and cleaned in ethyl alcohol before weighing. Electrolytic flake chromium (99.5+%) from Union Carbide Corp. was used for the chromium additions.

B. Melting

Melting was carried out in a non-consumable electrode arc furnace (Figure 1) using a water-cooled copper hearth of 22-cc capacity ($1-3/16"$ wide, $2-3/32"$ long, $5/8"$ deep). Approximately 330-gm charges (18-cc) were melted on the hearth under a reduced pressure of flowing argon (10 to 20 in. Hg, 5 cfm). The furnace was pumped down to less than one micron and flushed with argon at least three times before melting was started. Current was held at 250 to 350 amperes for the first 1/2 to 1 minute, then raised to 550 to 750 amperes (15 to 25 volts) for 1 to $1\frac{1}{2}$ minutes. Each button was flipped and remelted 4 to 7 times.

C. Rolling and Heat Treating

The cast ingots were hot-rolled into $\frac{1}{4}$ -inch thick slabs before the deformed-metastable-phase-transformation experiments were carried out. During this initial hot breakdown, the

slabs were reduced 10 to 20% between 625°C anneals in a molten salt bath (Figure 2). A proprietary chloride salt "Liquid Heat No. 980" was used for the annealing and also for the higher temperature heating required in subsequent treatments.

Deformation was carried out using the laboratory rolling mill shown in Figure 2; all experiments employed 6-inch-diameter hot rolls, heated with internal resistance heaters to about 200°C. The small lead bath adjacent to the rolling mill was used for interrupted quenching and for rapid isothermal rolling of alloys in the metastable state.

The screening experiments completed to date have involved an initial solution treatment at either 800°C (gamma phase) or 725°C (beta phase), followed by 1) rapid quenching to an intermediate temperature in the lead bath, 2) rapid rolling at the intermediate temperature, and 3) water-quenching or post-rolling heat treatment. For comparison, a number of additional specimens were subjected to the same or similar heat-treating cycles without deformation.

Quenching to the rolling temperature and rolling at this temperature were done as rapidly as possible so the deformation treatment could be completed before transformation was initiated. In most cases, the slabs were quenched from the salt bath into the lead bath for 20 seconds before rolling was started, and returned to the lead bath for 15 seconds between passes. Rolling time per pass (time out of the lead bath) varied from about 5 seconds in recent experiments to as long as 30 seconds in some of the earlier tests.

Twenty seconds was chosen as a conservative minimum time for the initial quenching treatment on the basis of heat transfer calculations which indicated that the center of a $\frac{1}{4}$ -inch slab would approach the surface temperature within one second under ideal conditions (Figure 3). Additional calculations indicated that a 5-mil oxide layer would increase the time by less than 50%.

D. Metallography

Specimens were mounted in bakelite and sanded by hand through 600-grit silicon carbide paper, using Carbitol (diethyl glycol monoethyl ether) as a lubricant. Microcloth-covered laps charged with 9-micron and 6-micron diamond abrasive were used for mechanical polishing. Final polishing was done electrolytically at 20°C, in a bath of the following composition:

H_3PO_4	50 ml
Ethylene glycol	50 ml
Methyl alcohol	80 ml

Two electrolytic etching solutions were employed:

(1) For bright filled illumination:

HNO_3	2 ml
Citric acid	2 gm
Methyl alcohol	92.5 ml

(2) For examination under polarized light:

CrO_3	10 gm
H_3PO_4	11 ml
H_2O	5 ml
Methyl alcohol	80 ml

Figure 1. NON-CONSUMABLE ELECTRODE ARC FURNACE

Figure 2. ROLLING AND HEAT TREATING AREA

EFFECT OF TEMPERATURE DROP AND THICKNESS ON MINIMUM TIME REQUIRED FOR THE CENTER OF QUENCHED URANIUM SLABS TO REACH THE LEAD BATH TEMPERATURE (WITHIN 15°C)

FIGURE 3

RESULTS

Experiments completed to date can be divided into two groups: a) screening tests with gamma-stabilized alloys containing 2 to 7 w/o molybdenum, and b) screening tests with beta-stabilized alloys containing small additions of chromium or chromium plus molybdenum. Table I summarizes rolling and heat-treating experiments on the gamma-stabilized alloys; Table II summarizes work on the beta-stabilized alloys.

Results of aging experiments on U-5 w/o Mo alloys, which had been rolled in the metastable state, are plotted in Figure 4. Photomicrographs of typical microstructures are shown in Figures 5 through 12. Details of the results on individual alloy compositions are discussed below.

A. Gamma-Stabilized Uranium Alloys

1. U-7 w/o Mo Alloy (Table I, Specimen 141A)

No difficulties were encountered in rolling this alloy to a reduction of 57% at 450°C. The microstructure and hardness indicated that little or no transformation had occurred. Since the structure was similar to that of the 5 w/o Mo alloy, which was given about the same thermal and mechanical treatment (Table I, specimen 128B-1, and Figure 6A and B), additional work in this composition range was confined to the more dilute alloy.

2. U-5 w/o Mo Alloy (Table I, Specimen 128)

This alloy was rolled without difficulty in the metastable state at 450°C. Reductions of 50% increased the DPH after water quenching from about 120 to 210 and produced some deformation bands in the microstructure (Figure 6A and B). There was no metallographic evidence of the deformed alpha structure (α'), however, and aside from deformation bands, the structure appeared to be 100% metastable gamma.

A series of elevated temperature-aging treatments on specimens from the rolled alloy indicated that deformation accelerated aging and induced transformation in the slip planes. Figure 4 shows the aging characteristics of the deformed specimens as well as results on undeformed specimens that had been given the same thermal treatment (Table I).

Representative photomicrographs of specimens from these two series are reproduced in Figures 5, 6, and 7.

Aging the undeformed gamma-quenched alloy at 450°C (Figure 5) produced the pearlitic-type eutectoid structure $\alpha + \gamma'$ that is characteristic of U-Mo alloys transformed

at this temperature. The time for initiation of transformation was also in the range found by other investigators⁵ for undeformed alloys of similar composition (Figure A-21). When the same alloy was quenched to 450°C and deformed in the metastable state, however, the time for initiation of transformation (as determined by hardness measurements, Figure 4) was greatly reduced. The microstructure of deformed specimens (Figure 6), when compared with undeformed specimens (Figure 5), shows that deformation induced nucleation in slip planes and retarded growth of the pearlitic transformation product. The decrease in growth rate is also apparent in the microstructures of deformed and undeformed specimens aged for one hour at 500°C (Figures 7C and D).

Considerable difference in microhardness was found between the light-etching, apparently untransformed, gamma-phase areas and the transformed pearlitic areas in some of these specimens. For example, with a 50-g load, the pearlite areas in specimen 128A-5 (Figure 5B) had a DPH of about 135, while the untransformed area gave readings of about 580. Since the hardness of retained gamma in quenched alloys (not aged) was quite low (Table I, specimens 128A-1, 128A-3, 128A-2, and 128B-1), the increase in hardness of the alloys on aging was probably caused by a submicroscopic hardening process in the light-etching areas. This hardening could be related to the diffusionless (martensitic) transformation from gamma to the "distorted alpha" (α') structure, which occurs when U-Mo alloys in this composition range are rapidly cooled or rapidly cooled and then deformed.^{6,7} The striated or banded microstructure characteristic of the α' was not found, however, in any of the U-5 w/o Mo alloys.

3. U-3.5 w/o Mo Alloys (Table I, Specimen 142)

This alloy was difficult to roll in the metastable state at both 450°C and 400°C. The somewhat greater reduction obtained at 400°C was probably due to the longer incubation period (time before initiation of transformation) at the lower temperature. In both cases, rolling was stopped when a marked increase in hardness or rolling resistance was encountered.

No TTT data for this composition was found in the literature, although the "C" curves for a number of binary alloys containing larger amounts of molybdenum (Figures A-18 and A-21) show that the incubation period would be a matter of minutes in undeformed samples of this analysis. As in the U-5 w/o Mo alloys discussed above, deformation probably decreased the incubation time even further.

The DPH after water quenching was significantly greater than that of the 5 w/o molybdenum alloy discussed above. The microstructure was somewhat striated when the specimen was heavily etched (Figure 8A). However, acicular or banded α' structures similar to those given in the literature^{6,7} and those obtained with the 2 w/o Mo alloy (Figure 9) were not obtained in any of the specimens from this alloy (Figure 8). Thus, it is not known whether the increase in hardness obtained by deformation was due to a martensitic transformation to α' or to a precipitation hardening process as suggested by Figure 8C.

4. U-2 w/o Mo Alloy (Table I, Specimens 143 and 83)

It was possible to roll this alloy to a reduction of only 5% at 400 and at 450°C. Severe cracking and shattering were encountered if further deformation was attempted.

The trend toward increasing hardness with decreasing molybdenum content in specimens that were water quenched from the gamma range (800°C) continued with this alloy (specimen 143A-2). The increased hardness may have been due to the martensitic transformation from γ to α' , although no evidence of the striated or banded α' structure was found in the microstructure of the water-quenched specimen (Figure 9A). The light-etching phase was clearly banded, however, and the hardness was greater when the specimens were given an interrupted quench either with or without deformation (Figure 9B, C, D). The dark patches in these photomicrographs were the normal pearlitic transformation product, which had been over-etched in bringing out the banded structure.

Specimen 143A-1, which had been held for less than 2 minutes at 450°C, transformed almost entirely to the pearlitic structure, although a few areas contained the banded structure as shown in Figure 9B. The sample that was held at 400°C for about one minute and deformed 5% (143B-1, Figure 9C) and the sample that was held at 400°C for 30 minutes without deformation (83-I, Figure 9D) both partially transformed to the pearlitic structure, probably before the martensitic phase change occurred.

No TTT diagrams were found for this alloy, although diagrams for 0.5 and 1.0 w/o molybdenum alloys have been published (Figures A-19 and A-20).⁸ These diagrams cannot be applied to the 2 w/o Mo alloy, since beta phase does not occur in alloys containing 2 or more percent molybdenum. They are of interest, however, since they are the only TTT diagrams found that show an M_s temperature (temperature below which the martensitic transformation $\gamma \rightarrow \alpha'$ will occur). This temperature apparently increases with molybdenum content in the low-alloy range.

B. Beta-Stabilized Dilute-Uranium Alloys

Uranium is not normally rolled in the relatively narrow and brittle beta temperature range, although reductions up to 90% can be obtained at 700 to 760°C if partial transformation to the softer alpha or gamma phases is prevented.⁹ The preliminary experiments discussed below indicate that rolling of metastable beta-phase alloys at temperatures as low as 450°C is also possible. Rolling becomes progressively more difficult as the temperature is lowered, however, and cracking occurs in some alloys at small reductions.

1. U-0.2 w/o Cr Alloy (Table II, Specimens 110 and 138)

The microstructures of specimens 110A-1 and 110A-2 showed considerable cracking, probably as a result of the large, single-pass reductions. However, since expansion due to the β - α transformation at room temperature may have contributed to the cracking, in subsequent tests the samples were held at an elevated temperature until transformation was complete. Specimens 138A-1 and 138A-2, for example, were heated to 500°C after the rolling treatment so that transformation would occur by the nucleation and growth mechanism which has been postulated to occur at the upper "C" portion of the TTT diagram (Figure A-5).¹⁰ It is apparent in the microstructures of these two samples that deformation reduced the transformed grain size (Figure 10A and B). The decrease in hardness accompanying deformation (Table II) is not understood. Figure 10C shows the much larger size obtained when isothermal transformation occurs at 600°C.

Specimen 110B-2 was water quenched after rolling and then immediately heated to 200°C before appreciable transformation or cracking took place. At this temperature, transformation occurs in the lower "C" portion of the TTT diagram (Figure A-5), probably by a martensitic reaction.¹⁰ This treatment resulted in an appreciable increase in hardness when compared with sample 138A-1, although the grain size was larger. As compared with the beta-quenched sample (specimen 138A-3, Figure 11C), the grain size of specimen 110B-2 (Figure 11D) appears to be somewhat smaller, but the hardnesses are equal.

Specimens 138B-1 and 138B-2 (which received the same thermal treatment without deformation) were quenched to the rolling temperature from the gamma region rather than the beta region. The gamma phase apparently transformed very rapidly to beta, since the alloy was very hard and considerable cracking occurred during the single-pass (2%) reduction. Photomicrographs of the two samples are shown in Figure 11A and B. Microcracks are apparent in the rolled specimen (Figure 11B).

2. U-0.4 w/o Cr Alloy

Two arc-melted ingots of this analysis cracked at room temperature before preliminary breakdown rolling was started. Rapid cooling on the watercooled hearth, after arc melting, apparently suppressed the β to α transformation, and cracking took place at room temperature when the expansion accompanying transformation occurred.

3. U-0.6 w/o Cr Alloy (Table II, Specimen 148A)

The first ingot of this composition cracked before rolling was started, probably by the mechanism discussed in 2. above. Cracking was prevented in the second sample by removing the ingot from the arc furnace while still warm and holding it at 600°C in the lead bath for one hour.

The single deformation test completed on this alloy (148A-1) indicated that the rolling characteristics at 450°C were similar to those of the 0.2 w/o Cr. Metallographic work is still in progress on this alloy, but preliminary results show a fine precipitate, probably chromium, which appears to be segregated on a microscopic scale. The soaking treatment at 725°C may not have allowed sufficient time for solution of the chromium, although 5 minutes is reported to be sufficient for alloys containing up to 0.87 w/o chromium.¹⁰

4. U-0.1 w/o Cr-0.2 w/o Mo Alloy (Table II, Specimens 108, 145, and 153)

The TTT diagram for this alloy (Figure A-15) indicates that the incubation period is greater than 10 minutes at temperatures between 400 and 500°C. Rolling in this temperature range may accelerate transformation, however, and it is possible that specimen 108B-1 transformed without cracking before rolling was completed.

Rolling this alloy at 450°C was very difficult (specimen 145A-1). Cracking occurred during the first two passes, and the slab would not pass through the rolls when a third 5% reduction was attempted. At 500°C the slabs did not crack, and four passes were achieved without difficulty (specimens 145B-1 and 153A-1).

Photomicrographs of three rolled slabs and three undeformed specimens that received equivalent thermal treatments are shown in Figure 12. Deformation appears to have reduced the grain size, but there was no significant change in hardness.

TABLE I
SUMMARY OF ROLLING AND HEAT TREATING EXPERIMENTS
GAMMA-STABILIZED URANIUM ALLOYS

Composition (w/o)	Specimen No.	Solution Treatment (a) (°C)	Isothermal Rolling Treatment (b)					Final Heat Treatment (c)			
			Lead Temp. (°C)	Time (min)	Number of Passes	Reduction (%)	Cooling Rate	Temp. (°C)	Time (min)	Cooling Rate	Hardness DPH
7% Mo, bal. U	141A	800	450	6.8	16	57	W.Q.	-	-	-	288
5% Mo, bal. U	128A-1	800	450	0.76	1	18	W.Q.	-	-	-	143
	-7	800	450	0.76	1	18	W.Q.	450	10	W.Q.	418
	-8	800	450	0.76	1	18	W.Q.	450	30	W.Q.	522
	-9	800	450	0.76	1	18	W.Q.	450	60	W.Q.	568
	-3	800	450	10.	0	0	W.Q.	-	-	-	117
	-2	800 W.Q.	-	-	-	-	-	-	-	-	121
	-4	800 W.Q.	-	-	-	-	-	400	60	W.Q.	134
	-5	800 W.Q.	-	-	-	-	-	450	60	W.Q.	310
	-10	800 W.Q.	-	-	-	-	-	450	120	W.Q.	535
	-11	800 W.Q.	-	-	-	-	-	450	180	W.Q.	517
	-6	800 W.Q.	-	-	-	-	-	500	60	W.Q.	415
	128B-1	800	450	1.32	3	50	W.Q.	-	-	-	210
	-2	800	450	1.32	3	50	W.Q.	400	60	W.Q.	546
	-5	800	450	1.32	3	50	W.Q.	450	10	W.Q.	486
	-6	800	450	1.32	3	50	W.Q.	450	30	W.Q.	509
	-3	800	450	1.32	3	50	W.Q.	450	60	W.Q.	566
	-7	800	450	1.32	3	50	W.Q.	450	180	W.Q.	556
	-4	800	450	1.32	3	50	W.Q.	500	60	W.Q.	500
3.5% Mo, bal. U	142A-1	800	450	2.32	4	10	W.Q.	-	-	-	497
	-2	800 W.Q.	-	-	-	-	-	-	-	-	251
	142B-1	800	400	3.78	4	15	W.Q.	-	-	-	476
2% Mo, bal. U	143A-1	800	450	1.42	2	5	W.Q.	-	-	-	577
	-2	800 W.Q.	-	-	-	-	-	-	-	-	408
	143B-1	800	400	1.06	2	5	W.Q.	-	-	-	568
	83-I	800	400	30.	0	0	W.Q.	-	-	-	530

(a) Heated for 10 minutes in salt bath at temperature indicated.

(b) Quenched from solution treatment temperature into lead bath and rolled at lead bath temperature.

(c) Lead bath or oven.

TABLE II
SUMMARY OF ROLLING AND HEAT TREATING EXPERIMENTS
BETA-STABILIZED DILUTE URANIUM ALLOYS

Composition (w/o)	Specimen No.	Solution Treatment (a) (°C)	Isothermal Rolling Treatment (b)					Final Heat Treatment (c)			Hardness DPH
			Lead Temp. (°C)	Time (min)	Number of Passes	Reduction (%)	Cooling Rate	Temp. (°C)	Time (min)	Cooling Rate	
0.2% Cr, bal. U	110A-1	725	525	0.18	1	35	Air	-	-	-	421
	-2	725	450	0.10	1	29	Air	-	-	-	431
	110B-1	725	450	0.97	1	2	W.Q. (1 min)	200	15	Air	392
	-2	725	450	3.59	7	15	W.Q. (1 min)	200	15	Air	397
	138A-1	725	450	7.60	18	35	Heated	500	30	Air	266
	-2	725	450	8	0	0	Heated	500	30	Air	349
	-3	725	-	-	-	-	-	200	15	Air	403
		(W.Q. 1 min)									
	-4	725	600	30	0	0	Air	-	-	-	301
	138B-1	800	450	0.7	1	2	W.Q. (1 min)	200	15	Air	383
0.6% Cr, bal. U	148A-1	725	450	4.3	5	10	Not cooled	450	180	Air	372
	-2	725	450	180	0	0	Air	-	-	-	393
0.1% Cr, 0.2% Mo, bal. U	180A-1	725	500	0.7	1	8	Air	-	-	-	345
	108B-1	725	500	11.5	16	60	Air	-	-	-	366
	145A-1	725	450	3	2	10	Heated	500	120	Air	359
	-2	725	450	8	0	0	Heated	500	120	Air	355
	145B-1	725	500	2	4	10	W.Q. (1 min)	200	180	Air	342
	-2	725	500	2	0	0	W.Q. (1 min)	200	180	Air	405
	-3	725	500	180	0	0	Air	-	-	-	355
	153A-1	725	500	2	4	10	Not cooled	500	180	Air	378

(a) Heated for 3 to 5 minutes in salt bath at temperature indicated.

(b) Quenched from solution treatment temperature into lead bath and rolled at lead bath temperature.

(c) Lead bath or oven.

EFFECT OF DEFORMATION AT 450°C ON ELEVATED-TEMPERATURE AGING CHARACTERISTICS OF METASTABLE URANIUM-5 w/o MOLYBDENUM ALLOYS

FIGURE 4

A. (128 A-2)

500X

B. (128 A-5)

500X

C. (128 A-10)

500X

Figure 5. U-5 w/o Mo ALLOY

A. Water quenched from 800°C.

B. Water quenched from 800°C, aged 60 minutes at 450°C.

C. Water quenched from 800°C, aged 120 minutes at 450°C.

A. (128 B-1)

150X

B. (128 B-1)

500X

C. (128 B-5)

500X

D. (128 B-7)

500X

Figure 6. U-5 w/o Mo ALLOY

A. and B. Quenched to 450°C , rolled 50%, water quenched

C. Same as A. and B. but aged 10 minutes at 450°C

D. Same as A. and B. but aged 180 minutes at 450°C

A. (128 B-2)

500X

B. (128 B-3)

500X

C. (128 B-4)

500X

D. (128 A-6)

500X

Figure 7. U-5 w/o Mo ALLOY

- A. Quenched to 450°C , rolled 50%, water quenched, aged 60 minutes at 400°C
- B. Same as A. but aged 60 minutes at 450°C
- C. Same as A. but aged 60 minutes at 500°C
- D. Water quenched from 800°C , aged 60 minutes at 500°C

A. (142 A-2)

500X

B. (142 A-1)

500X

C. (142 B-1)

500X

Figure 8. U-3.5 w/o Mo ALLOY

A. Water quenched from 800°C.

B. Quenched to 450°C, rolled 10%, water quenched.

C. Quenched to 400°C, rolled 15%, water quenched.

A. (143 A-2)

500X

B. (143 A-1)

500X

C. (143 B-1)

500X

D. (83-I)

500X

Figure 9. U-2 w/o Mo ALLOY

- A. Water quenched from 800°C
- B. Quenched to 450°C, rolled 5%, and water quenched
- C. Quenched to 400°C, rolled 5%, and water quenched
- D. Quenched to 400°C for 30 minutes and water quenched

A. (138 A1)

B. (138 A2)

C. (138 A4)

500X

Figure 10. U-0.2 w/o Cr ALLOY

POLARIZED LIGHT

A. Quenched from 725°C to 450°C, rolled 35%, transformed at 500°C

B. Same as A. but without deformation

C. Quenched from 725°C to 600°C, transformed at 600°C

A. (138 B-2)

500X

B. (138 B-1)

500X

C. (138 A-3)

500X

D. (110 B-2)

500X

Figure 11. U-0.2 w/o Cr ALLOY

POLARIZED LIGHT

- A. Quenched from 800°C to 450°C for 30 sec, water quenched for 60 sec, heated to 200°C for 15 min.
- B. Same as A. but rolled 2% at 450°C.
- C. Water quenched for 60 sec from 725°C, heated to 200°C for 15 min.
- D. Quenched from 725°C to 450°C, rolled 15%, water quenched for 60 sec, heated to 200°C for 15 min.

A.
(145 A-1)
500X

B.
(145A-2)
500X

C.
(145 B-1)
500X

D.
(145B-2)
500X

E.
(153 A-1)
500X

F.
(145B-3)
500X

Figure 12. U-0.1 w/o Cr-0.2 w/o Mo ALLOY

POLARIZED LIGHT

- A. Quenched from 725°C to 450°C, rolled 10%, transformed at 500°C
- B. Same as A. but without deformation
- C. Quenched from 725°C to 500°C, rolled 10%, transformed at 200°C
- D. Same as C. but without deformation
- E. Quenched from 725°C to 500°C, rolled 10%, transformed at 500°C
- F. Same as E. but without deformation

FUTURE PROGRAM

During the next few months, screening experiments will be continued on the alloy systems discussed above. Additional binary and ternary alloys, including those containing columbium and zirconium, will be screened. Rolling of flat slabs will remain the principal method of deformation under investigation. Small ingots prepared in the non-consumable electrode furnace will also be used for most of the experiments. However, melting will be carried out in the vacuum resistance melting furnace for the most promising alloys, to duplicate conventional uranium melting practices and to produce larger ingots required for detailed study.

Selection of alloys for evaluation is based primarily on available information developed in completed studies on the transformation kinetics of uranium-base alloys. However, some alloy development and time-temperature-transformation work (TTT diagrams) will be conducted as dictated by results obtained during the course of the program.

Processed alloys will continue to be evaluated by metallographic techniques and by physical property measurements at room and elevated temperatures. In the most promising cases, creep behavior and response to various special treatments will also be determined.

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APPENDIX A
TIME-TEMPERATURE-TRANSFORMATION DIAGRAMS
AND CONSTITUTIONAL DIAGRAMS

The time-temperature-transformation characteristics of several dilute uranium alloys are shown in the "TTT" diagrams given in Figures A-1 through A-23. Constitutional diagrams for the U-Mo and the U-Cr systems are given in Figures A-24 and A-25. References from which the diagrams were taken are given on the graphs.

Figure A-1. TTT DIAGRAM OF HIGH-PURITY URANIUM

Figure A-2. TTT DIAGRAM OF TECHNICAL-PURITY URANIUM

From M. D. Jepson, et al, Second U.N.Conf. on Peaceful
Uses of Atomic Energy, 15/P/27, p 15 (1958)

Figure A-3. TTT DIAGRAM OF
U-0.066 w/o Cr ALLOY

Figure A-4. TTT DIAGRAM OF
U-0.098 w/o Cr ALLOY

Figure A-5. TTT DIAGRAM OF
U-0.13 w/o Cr ALLOY

From D. W. White, Transactions of AIME, 203, p 1221 (1955)

Figure A-6. TTT DIAGRAM OF U-0.38 w/o Cr ALLOY

Figure A-7. TTT DIAGRAM OF U-0.86 w/o Cr ALLOY

From D. W. White, Transactions of AIME, 203, p 1221 (1955)

Figure A-8. TTT DIAGRAM OF U-0.11 w/o Cr ALLOY

Figure A-9. TTT DIAGRAM OF U-0.11 w/o Cr-0.048 w/o Fe ALLOY

From M. D. Jepson, et al, Second U.N. Conf. on Peaceful
Uses of Atomic Energy, 15/P/27, p 15 (1958)

Figure A-10. TTT DIAGRAM OF U-0.41 w/o V ALLOY

Figure A-11. TTT DIAGRAM OF U-0.059 w/o Fe ALLOY

From M. D. Jepson, et al, Second U.N. Conf. on Peaceful
Uses of Atomic Energy, 15/P/27, p 15 (1958)

Figure A-12. TTT DIAGRAM OF U-0.050 w/o Mo ALLOY

Figure A-13. TTT DIAGRAM OF U-0.020 w/o Mo ALLOY

From M. D. Jepson, et al, Second U.N. Conf. on Peaceful
Uses of Atomic Energy, 15/P/27, p 15 (1958)

Figure A-14. TTT DIAGRAM OF U-0.10 w/o Mo-0.082 w/o Cr ALLOY

Figure A-15. TTT DIAGRAM OF U-0.20 w/o Mo-0.11 w/o Cr ALLOY

From M. D. Jepson, et al, Second U.N. Conf. on Peaceful Uses of Atomic Energy, 15/P/27, p 15 (1958)

Figure A-16. TTT DIAGRAM OF U-0.057 w/o Al ALLOY

Figure A-17. TTT DIAGRAM OF U-0.10 w/o Ti ALLOY

From M. D. Jepson, et al, Second U.N. Conf. on Peaceful
Uses of Atomic Energy, 15/P/27, p 15 (1958)

Figure A-18. TTT CURVES FOR THE BEGINNING OF TRANSFORMATION IN URANIUM-MOLYBDENUM ALLOYS

From R. K. McGeary, WAPD-127 (1956) and van Thyne and McPherson, Trans. of ASM, 49, p 598 (1957)

Figure A-19. ISOTHERMAL TRANSFORMATION DIAGRAM OF U-0.5 w/o Mo

Figure A-20. ISOTHERMAL TRANSFORMATION DIAGRAM OF U-1.0 w/o Mo

From E. G. Zukas, Proc. of 2nd Nuclear Eng. and Sc. Conf., 2, p 173 (1957)

Figure A-21. TTT DIAGRAM FOR A U-5.4 w/o Mo ALLOY ILLUSTRATING INITIATION OF TRANSFORMATION AS DETERMINED BY VARIOUS TECHNIQUES

Figure A-22. TIME-TEMPERATURE RELATIONS FOR ISOTHERMAL TRANSFORMATION OF U-5.0 w/o Zr-1.5 w/o Nb ALLOY BROUGHT DOWN FROM THE GAMMA RANGE

Figure A-23. TTT DIAGRAM ILLUSTRATING INITIAL HARDNESS CHANGE FOR U-Nb and U-Nb-X ALLOYS

From R. J. van Thyne, WASH-701, p 11 (1957)

Figure A-24. CONSTITUTIONAL DIAGRAM OF URANIUM-MOLYBDENUM

Reproduced from F. A. Rough and A. A. Bauer, Constitutional Diagrams of Uranium and Thorium Alloys, Reading, Mass., Addison-Wesley Publ. Co., Inc. (1958), p 41.

Figure A-25. CONSTITUTIONAL DIAGRAM OF
URANIUM-CHROMIUM

Reproduced from F. A. Rough and A. A. Bauer,
Constitutional Diagrams of Uranium and Thorium
Alloys, Reading, Mass., Addison-Wesley Publ.
Co, Inc. (1958), p 16.

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