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**THE PROPERTIES OF A METASTABLE
GAMMA-PHASE URANIUM-BASE
ALLOY: U-7.5 Nb-2.5 Zr**

Livermore, California

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URANIUM-BASE ALLOY: U-7.5 Nb-2.5 Zr

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ABSTRACT

The uranium-niobium-zirconium ternary system exhibits a series of solid solutions at high temperature. Addition of both niobium and zirconium stabilizes the body-centered-cubic gamma phase to a lower temperature. A study has been made of one particular ternary alloy, the U-7.5 w/o Nb-2.5 w/o Zr composition. The relationship of transformation characteristics of the alloy to its physical and mechanical properties has been determined. This weldable alloy is concluded to be an excellent material for engineering applications. Tensile yield strengths of 100,000 to 200,000 psi are obtainable, with elongations of 12% to 4%, respectively, depending on heat treatment.

INTRODUCTION

While gamma-stabilized uranium is of interest because of the more isotropic nature of the body-centered-cubic gamma phase and its amenability to heat treatment, the binary alloys of uranium-molybdenum, uranium-niobium, and uranium-zirconium that have been developed thus far each possess some shortcoming with respect to engineering applications. The gamma uranium-niobium alloys, for instance, have low yield strengths, causing them to creep or rupture under high loads; the gamma uranium-molybdenum alloys fail in a brittle manner under static or low-strain-rate loading in environments containing oxygen; the gamma uranium-zirconium alloys transform to very strong, hard alloys but they lack ductility at room temperature. In the gamma phase a balance of strength and stability at room temperature is obtained with a composition of U-7.5 w/o Nb-2.5 w/o Zr.

Data available from studies of the isothermal gamma-phase reactions of the binary alloys^{1,2} and from work on the ternary equilibrium diagram of

uranium-niobium-zirconium³ have stimulated a detailed examination of alloys in the uranium-rich corner of the ternary system. A ternary gamma-phase alloy with a nominal composition of U-7.5 w/o Nb-2.5 w/o Zr has been developed as a result of this investigation. This alloy is not susceptible to stress cracking⁴ in air or oxygen and can be heat treated to obtain a wide range of properties.

EXPERIMENTAL PROCEDURE

The selection of the U-7.5 w/o Nb-2.5 w/o Zr composition for evaluation was based, in part, on the behavior of the binary uranium-niobium and uranium-zirconium alloys. It was desired to combine a sufficient amount of niobium for stabilizing the gamma phase with a sufficient amount of zirconium for strengthening. It was also desired to retain, as much as possible, the ductility which is characteristic of the binary uranium-niobium alloys.

Small arc-melted buttons as well as plate from 10-kg and 100-kg ingots were used to study the phase stabilities, isothermal transformations, and physical and mechanical properties of the alloy. The results of these studies, along with fabrication procedures and other pertinent information, are described below.

1. Materials

The highest purity materials available were used. Emphasis was placed on low interstitial impurities. Typical analyses of materials used are given in Table I.

2. Preparation of the Alloys

The material which was used in the differential thermal analysis (DTA), isothermal experiments for the TTT curves, and hardness-versus-heat-treatment studies was prepared in the form of 200-g buttons. Ingots of approximately $2 \times 1 \times 0.375$ in. were formed in a water-cooled copper mold by arc-melting the charge in an argon atmosphere. Five remelts were made and the final ingot was homogenized in vacuum for 72 hours at 850°C. The analyses of the ingots were substantially the same as the charges, i. e., 90 w/o uranium, 7.5 w/o niobium, and 2.5 w/o zirconium.

Ingots in the 8- to 10-kg range were arc-melted for physical and mechanical property samples. Sandwich-type electrodes of individual metal

Table I. Typical analyses of materials used for uranium alloys.

Uranium (ppm)		Molybdenum (ppm)		Niobium (ppm)		Zirconium (ppm)	
C	35	C	190-200	C	140	C	<600
Fe	150	Si	100	Ta	1000	Fe	600
Mg	21	Fe	50	Ti	300	Cr	87
Ni	30	Cr	20	Fe	150	Mn	15
Si	20	Sn	20	O ₂	1000	Al	28
Mn	35	Mo	Bal.	N ₂	750	O ₂	600
Pb	<4			Si	<50	N ₂	20
Cu	20			B	< 10	Ti	24
Al	<11			Nb	Bal.	Hf	100
B	<0.4					Pb	<50
Ca	11					Si	50
U	Bal.					Ni	10
						Zr	99.75%

strips, edge-welded into a 30 × 1 × 1-in. compact, were thoroughly outgassed in vacuum at 850°C and arc-melted in argon. The 3-in. -diam ingot was quartered, rewelded into a new electrode, vacuum-outgassed, and arc-melted to another 3-in. -diam cylinder about 6 in. long. After surface-machining and cropping, the ingot was heated to 950°C in argon and hot-forged to 2-in. -thick plate. This plate was then hot-rolled from 840°C to finished size (0.325-in.-thick plate), annealed at 840°C in a salt bath for 1/2 hour, and cooled to room temperature in argon.

Commercial-size plates of approximately 100 kg were produced at the Bureau of Mines in Albany, Oregon. The electrodes for these melts were formed from laminations of bars of uranium and strips of niobium and zirconium. The welded assembly was arc-melted at a pressure of about 60 microns into a 4-in. -diam, water-cooled copper mold. The ingot was faced off and butt-welded to a similar ingot, forming a new electrode which was arc-melted into a 6-in.-diam ingot. This ingot was quartered, rewelded, and melted into a final 6-in.-diam ingot in a vacuum.

The machined 90-kg ingot was heated for 1-1/2 hours in argon at 950°C and press-forged to a rectangular slab. The slab was hot-rolled at 840°C to a plate about 0.875 in. thick, then heated to 800°C in a salt bath and hot-rolled to 0.625-in. plate. The final heat treatment consisted of a 1-hour anneal at 950°C in argon followed by cooling in an argon-filled chamber to room temperature. Analyses of the 9-kg and 90-kg plates are given in Table II.

Table II. Analyses of 9-kg and 90-kg plate of U-7.5 w/o Nb-2.5 w/o Zr alloy.

	9-kg plate	90-kg plate
Nb	7.4%	7.3%
Zr	2.5%	2.5%
C	55 ppm	100 ppm
O ₂	40 ppm	35 ppm
N ₂	--	14 ppm
H ₂	--	18 ppm

TRANSFORMATIONS

1. Differential Thermal Analysis

Standard 0.375-in. × 0.250-in.-diam cylindrical samples were machined from homogenized (72 hours at 850°C) 200-g button ingots. They were cycled from room temperature to 900°C and back to room temperature at a uniform rate of 85°C per hour. Figure 1 is a photograph of a chart recording a cycled run for pure uranium showing the sharp thermal peaks of the gamma-to-beta and beta-to-alpha transformations. Figure 2 is a photograph of a cycle made on a sample of U-7.5 w/o Nb-2.5 w/o Zr, as wrought and gamma-annealed.

In the ternary alloy the temperature of the gamma-phase transformation has been lowered substantially and the beta-phase reaction has been eliminated. The alloy transforms on cooling directly from gamma to alpha phase. The original gamma-phase transformation of the pure uranium now occurs at about 500°C in the ternary alloy. It is also very sluggish and drawn out and is not completed until the temperature has fallen to 363°C. At room

temperature, x-ray analysis shows the alloy to be a mixture of alpha uranium and γ' , a bcc niobium-rich solid solution. No delta phase corresponding to UZr_3 was detected at this niobium concentration.

Results from thermal cycling of various samples made from different melts and processed under different conditions agree closely in the manner of the gamma phase transformation.

2. Dilatometry

The linear expansion and contraction of samples in the range from room temperature to 1000°C were determined. With the same heating and cooling rates used in the differential thermal analysis, a typical curve (Fig. 3) for a sample of the U-7.5 w/o Nb-2.5 w/o Zr alloy (from a 90-kg hot-worked plate) was determined.

On heating, the alloy expands normally to 600°C , at which point a very rapid expansion occurs to 640°C . These points agree quite well with those from the DTA (602 and 654°C). During cooling the cycle is normal with a somewhat exaggerated contraction at about 375°C . This point is considerably lower than that for the DTA (500°C) and is probably due to the conditions of cooling and the temperature measuring sensitivity of the equipment used.

The average linear coefficients of expansion calculated from the chart are given in Table III.

3. Isothermal Transformation

The time-temperature-transformation or TTT curve is a familiar method of determining the gamma phase stability of the alloy under non-equilibrium conditions. In the present study the time of origin of transformation was estimated from the beginning of changes in electrical resistance, hardness, and microstructure. Figure 4 shows TTT curves assembled from the reaction data at the various isotherms using both the resistivity and hardness measurements. The agreement is quite good. In the upper regions from 500 to 600°C the microstructures agree quite well, but below 500°C metallographic results are difficult to interpret and hence the TTT curves are made without these data. The microstructures were identified in some cases by use of x-ray diffraction and in others by effects of chemical etchants.

Fig. 3. Linear expansion of U-7.5 w/o Nb-2.5 w/o Zr.

Fig. 4. Time-temperature-transformation (TTT) curves for U-7.5 w/o Nb-2.5 w/o Zr.

Table III. Coefficient of linear thermal expansion for U-7.5 w/o Nb-2.5 w/o Zr alloy.

Temperature range (°C)	Coefficient of expansion (in./in.-°C)
20 to 300	14.8×10^{-6}
600 to 635	124.7×10^{-6}
640 to 1000	14.4×10^{-6}
20 to 1000	16.9×10^{-6}

The resistivity measurements were made at the temperature of transformation after a direct transfer of the sample from the gamma annealing bath to the isothermal transformation bath. The equipment allows readings to be taken automatically in decades from 1 minute to 10,000 minutes. Immediate readout of resistivity changes as small as 0.0004 ohm is noted on semi-log paper. The thin, $2 \times 0.062 \times 0.062$ -in. samples are enclosed in stainless steel containers filled with helium for good heat transfer.

Small cubes 0.375 in. on an edge were used as samples for determination of hardness and microstructural changes accompanying transformation. The specimens were transformed in helium-filled containers by quenching from the gamma-annealing temperature directly into a salt bath held at a preselected temperature. After isothermal transformation the container was water-quenched. The hardness change with time at various temperatures is shown in Fig. 5. The data used for hardness portion of the TTT curves (Fig. 4) are based on the time to attain one-half of the fully aged hardness.

Microstructures of the alloy transformed at 550 and 600°C show a pearlitic type of structure. The alpha platelets nucleate at grain boundaries, and they advance into the grain interior as the gamma phase transforms. The plates appear coarser at the higher temperature. The morphology (i. e., the planes on which the growth occurs) has not been determined. Microstructures showing decomposition of the gamma phase to the pearlite structure at 600°C after 1, 2.5, and 8 hours are reproduced in Figs. 6, 7, and 8, respectively.

Figures 9, 10, and 11 show the aging effect of 0.25, 1, and 8 hours, respectively, at 450°C. Between 450 and 500°C the transformed material is much finer than at 600°C and is distributed throughout the grains as a veined

Fig. 5. Age hardening of U-7.5 w/o Nb-2.5 w/o Zr. GLL-647-1895

Isothermal Transformation of Gamma
U-7.5 w/o Nb-2.5 w/o Zr at 600°C

GLB 647 4240

Fig. 6. 1 hour (400X)

GLB 647 4241

Fig. 7 2.5 hours (400X)

GLB 647 4242

Fig. 8a. 8 hours (400X).

GLP

Fig. 8b. 8 hours /

Isothermal Transformation of Gamma U-7.5 w/o Nb-2.5 w/o Zr at 450°C
(400X, Polarized Light)

GLB-647-4244

Fig. 9. 0.25 hour.

GLB-647-4245

Fig. 10. 1 hour.

GLB-647-4246

Fig. 11. 8 hours.

or salt-and-pepper type of nucleation product. Polarized light is used to bring this out more vividly.

At the temperatures below 450°C where the hardening effect is both rapid and intensive, the microstructures show a martensitic type of strained structure but no discrete phase of precipitate. Figure 12, taken of the structure as transformed for 1 hour at 350°C, and Fig. 13, of the structure transformed for 2-1/2 hours at 400°C, show this type of pattern. Extensive line broadening in the x-ray diffraction patterns lends evidence to support the existence of a highly stressed internal structure.

PHYSICAL PROPERTIES

Some of the physical properties as determined for the alloy are as follows:

1. Density at room temperature: cast, 16.4 g/cc; wrought, 16.6 g/cc.
2. Coefficient of linear expansion: see Table III.
3. Crystallography (as gamma-quenched): gamma phase, body-centered cubic.
4. Electrical resistivity: gamma-quenched from 850°C, 70 $\mu\text{ohm-cm}$; transformed 8 hours 600°C, 53 $\mu\text{ohm-cm}$.
5. Specific heat at 23°C: 0.04 cal/g-°C.

MECHANICAL PROPERTIES

The yield strength, ultimate tensile strength, and ductility in tension, as well as the hardness and the toughness of the alloy, vary considerably depending upon the heat treatment given the alloy.

Table IV is a summary of the tensile properties for many heat treatments of the alloy. There is very little effect on these properties due to different strain rates and environment in testing. The material with the highest strength and lowest ductility is the slowly cooled, gamma-annealed alloy and the gamma-quenched alloy aged at 350 to 450°C. In the gamma-quenched condition the metal is the softest and most ductile.

Figure 5 shows the response to hardening of the gamma-quenched alloy with different heat treatments.

Table V, which gives tensile data obtained on the alloy above and below room temperature, indicates no definite transition temperature from -20°C to 150°C. Slight strengthening occurs at the lower temperature but ductility remains the same.

Isothermal Transformation of Gamma U-7.5 w/o Nb-2.5 w/o Zr
(400X, Polarized Light)

GLB-647-4247

Fig. 12. 1 hour at 350°C.

GLB-647-4248

Fig. 13. 2.5 hours at 400°C.

Table IV. Room-temperature tensile properties of U-7.5 w/o Nb-2.5 w/o Zr alloy.

Testing environment	Conditioning of sample	Strain rate (in./in.-min)	Yield strength, 0.2% offset (psi)	U. T. S. (psi)	% elong. in 2 in.	% RA
Air, R	900°C/30 min/vac./WQ	0.05	68,300	152,300	12.3	46
Air, F	840°C/salt bath/30 min/AC	0.05	87,000	182,300	10.5	24
Air, R	800°C/1 hour/vac./WQ	0.05	100,300	128,300	12.5	21
Air, R	800°C/1 hour/vac./WQ	0.05	108,300	130,300	12.0	21
Air, R	" "	0.005	91,600	120,300	14.4	21
*Air, F	" "	0.05	95,000	130,200	12.0	34
Air, R	950°C/1 hour/argon/AC	0.05	131,400	190,800	5.2	-
Vac., R	" "	0.05	134,300	191,300	5.2	-
Air, R	" "	0.0005	113,500	185,500	6.5	-
Vac., R	" "	0.0005	110,300	186,900	6.3	-
Air, R	" "	0.05	135,400	197,000	5.0	-
*Air, R	950°C/1 hour/AC + 8 hours/600°C	0.05	139,700	182,400	7.8	-
*Air, R	" "	0.05	143,800	176,200	6.9	-
Air, R	800°C/1 hour/vac./WQ + 8 hours/600°C	0.05	129,400	186,200	8.4	-
Air, R	" "	0.05	138,400	191,500	7.8	-
Air, R	900°C/1 hour/vac./WQ + 1 hour/600°C	0.05	66,500	163,600	13.0	41
Air, R	900°C/1 hour/vac./WQ + 1 hour/500°C	0.05	77,000	177,000	9.5	33
Air, R	900°C/1 hour/vac./WQ + 1 hour/400°C	0.05	210,000	219,000	1.0	1.3
Air, R	900°C/1 hour/vac./WQ + 1 hour/350°C	0.05	220,000	225,000	1.0	0.5

Table IV. Room-temperature tensile properties of U-7.5 w/o Nb-2.5 w/o Zr alloy. (Continued)

Testing environment	Conditioning of sample	Strain rate (in./in.-min)	Yield strength, 0.2% offset (psi)	U. T. S. (psi)	% elong. in 2 in.	% RA
Air, F	800°C/1 hour/vac./WQ + 1 hour/400°C	0.05	244,000	251,000	0.5	—
Air, F	" "	0.05	230,000	251,000	1.0	1.3
Air, F	" " 350°C	0.05	209,000	230,000	6	17
Air, F	" " 4 hours/350°C	0.05	260,000	270,000	2	2

* Stressed over 200 hours at 90% Y. S. before testing.

R = round-shoulder type tensile bars.

F = flat-grip type tensile bars.

WQ = water-quenched.

AC = air-cooled.

Table V. Effect of temperature of testing on tensile properties of 90 U-7.5 w/o Nb-2.5 w/o Zr alloy, vacuum-annealed at 900°C and water-quenched.

Temperature of test (°C)	Condition of sample	Yield strength, 0.1% offset (psi)	U. T. S. (psi)	% elong. in 2 in.	% RA
150	900°C/vac./½ hour/WQ	74,000	132,000	12.5	45
100	900°C/vac./½ hour/WQ	68,200	137,000	12.0	45
25	900°C/vac./½ hour/WQ	68,300	150,000	12.3	46
0	900°C/vac./½ hour/WQ	64,000	159,000	13.0	46
-30	900°C/vac./½ hour/WQ	65,000	166,000	15.3	47.5

WQ = water-quenched.

The standard, room-temperature Charpy-impact-test data given in Table VI correlate well with the tensile results. The toughness of the gamma-quenched alloy is good. Aging of the alloy to give high tensile strength adversely affects the impact strength and ductility. Photomicrographs (Figs.14 and 15) show the great difference in fracture pattern between gamma-quenched and aged materials. Slowly cooled material has an impact strength between these two; the structure at the fracture of gamma-annealed and air-cooled plate is shown in Fig. 16. Quenching gamma-annealed alloy directly to the transformation temperature instead of first quenching to room temperature produces better toughness for the same strength of alloy. Also, lower temperature gamma-anneals give better Charpy impact strength than high temperature anneals.

Table VII indicates that, as in the case of tensile data, there is no change in toughness from 100°C down to -40°C for the U-7.5 w/o Nb-2.5 w/o Zr gamma-annealed alloy.

The tensile properties of the alloy at temperatures up to 1000°C are quite good. Some tensile data obtained on thin strip specimens heated by electric current in a Marquardt tensile tester are summarized in Table VIII and given in graph form in Fig. 17. The tests conducted at the slower strain rate of 0.03 in./in.-min gave questionable yield-load results, particularly above 650°C. The tests at the higher strain rate of 6 in./in.-min gave much more consistent data for both yield loads and elongation.

The counteraction of the natural annealing effect by the transformation aging effect as the temperature of the test rises is apparent at temperatures

Table VI. Effect of heat treatment on room-temperature V-notch Charpy values of U-7.5 w/o Nb-2.5 w/o Zr alloy.

Heat treatment	Charpy (ft-lb)	Rockwell C hardness	X-ray structure
950°C/1 hour/argon + argon Q	5.4	41	γ
900°C/1 hour/vac. + water Q	16.3	23	γ
840°C/1 hour/salt + water Q	20.8	17	γ
800°C/1 hour/salt + water Q	15.6	22	γ
800°C/1 hour/vac. + water Q	20.7	23	γ
800°C/1 hour/vac. + tube Q	9.5	43	γ
800°C/1 hour/vac. + oil Q	15.3	22	γ
800°C/3 hours/salt + water Q	16.8	18	γ
750°C/3 hours/salt + water Q	9.6	18.5	γ
950°C/1 hour/argon Q + 350°C/1 hour	4.2	46	$\alpha + \gamma'$
950°C/1 hour/argon Q + 350°C/8 hours	3.7	47	$\alpha + \gamma'$
950°C/1 hour/argon Q + 400°C/1 hour	3.6	52	$\alpha + \gamma'$
950°C/1 hour/argon Q + 600°C/8 hours	3.5	48.5	$\alpha + \gamma'$
800°C/1 hour/vac. Q + 350°C/1 hour	9.8	45	$\alpha + \gamma'$
800°C/1 hour/vac. Q + 400°C/ $\frac{1}{2}$ hour	9.0	44	$\alpha + \gamma'$
800°C/1 hour/vac. Q + 400°C/1 hour	6.1	47.5	$\alpha + \gamma'$
800°C/1 hour/vac. Q + 400°C/2 hours	3.7	52.0	$\alpha + \gamma'$
800°C/1 hour/vac. Q + 600°C/1 hour	9.6	33	$\alpha + \gamma'$
800°C/1 hour/vac. Q + 600°C/8 hours	4.8	43	$\alpha + \gamma'$

Q = quenched.

Fractures of V-Notch Charpy Bars of U-7.5 w/o Nb-2.5 w/o Zr (400X)

GLB-647-4249

GLB-647-4250

Fig. 14. As gamma-quenched.

Fig. 15. As aged 1 hour at 400°C.

GLB-647-4251

Fig. 16. As gamma-annealed, air-cooled.

Table VII. V-notch Charpy impact values of gamma-annealed U-7.5 w/o Nb-2.5 w/o Zr alloy at various temperatures.

Temperature (°C)	Charpy (ft-lb)
100	11.0
25	12.5
0	10.8
-40	10.3

Table VIII. Elevated-temperature tensile tests of U-7.5 w/o Nb-2.5 w/o Zr alloy.

Temperature (°C)	Strain rate (in./in.-min)	Yield strength (psi)	U. T. S. (psi)	% elong. in 1 in.	RA (%)
R. T.	0.03	95,700	109,400	32	51
150	"	71,400	91,400	24	46
250	"	60,200	92,700	15	41
350	"	56,800	96,300	*	28
450	"	46,700	85,500	7	14
550	"	25,700	33,300	19	60
650	"	17,500	17,500	25	47
750	"	**	12,300	59	38
850	"	**	6,700	60	36
950	"	**	4,300	49	63
R. T.	6.0	101,000	111,000	*	42
150	"	71,200	90,300	20	39
250	"	69,400	84,800	14	38
350	"	71,400	78,300	10	36
450	"	53,800	59,600	8	44
550	"	37,600	41,000	12	52
650	"	24,100	26,000	19	59
750	"	18,800	18,850	29	60
850	"	14,100	15,000	40	74
950	"	**	10,300	60	85

* Broke outside gauge length.

** Yield load is questionable.

Fig. 17. Tensile properties of U-7.5 w/o Nb-2.5 w/o Zr at elevated temperatures.

in the range of 500-600°C. As the temperature becomes elevated to the point where nucleation processes predominate, the alloy anneals and softens rapidly.

FABRICATION PROPERTIES

For the best primary or secondary working properties the alloy must be in the gamma-phase condition above 700°C. For forging and breakdown rolling a starting temperature of 950°C is recommended. For hot-forming thin sections (finish-rolling), and to keep a fine grain size, the fabrication range is restricted between 800 and 700°C. Warm-rolling below 650°C is unsuccessful due to duplex structure formation caused by transformation reactions. Such material is "hot short" and fractures in the grain boundaries with little or no reduction.

Homogeneous material, gamma-annealed above 750°C and rapidly cooled, may be cold-worked at room temperature with reductions of 80% in thickness between anneals. Figure 18 shows the low work-hardening characteristics of the gamma-annealed-and-quenched alloy.

The recrystallization temperature for the 50% cold-rolled alloy is between 600 and 650°C, as indicated by the annealing curve of Fig. 19. Figure 20 shows that the cold-worked structure has not been removed by a 1-hour anneal at 600°C; Fig. 21 shows that after 1 hour at 650°C the structure has been completely recrystallized.

Heavy sections such as thick plates, when slowly cooled from the gamma-phase temperature, age-harden before reaching room temperature. Cold-rolling before fracture is limited to approximately 30% reduction in thickness with very little work-hardening. Transformation of the structure to pearlite at 600°C allows about the same degree of cold-rolling. Alloy transformed to maximum hardness at lower temperatures has no cold-reduction capability.

WELDABILITY

Parts made from the U-7.5 w/o Nb-2.5 w/o Zr alloy are as readily joined by TIG or electron beam fusion welding as parts made of unalloyed uranium. In the TIG process either argon or helium forms the protective atmosphere within the welding chamber. The electron beam process operates in a vacuum of at least 10^{-4} torr. The alloy becomes suddenly quite fluid and some skill is required when deep penetration is required.

Fig. 18. Hardness of U-7.5 w/o Nb-2.5 w/o Zr as a function of percent reduction by cold rolling.

Fig. 19. Hardness of U-7.5 w/o Nb-2.5 w/o Zr as a function of annealing temperature.

Recrystallization of 50% Cold-Rolled Gamma
U-7.5 w/o Nb-2.5 w/o Zr (400X)

GLB-647-4253

Fig. 20. Annealed 1 hour at 600°C.

GLB-647-4254

Fig. 21. Annealed 1 hour at 650°C.

The cracking under residual stress which is common in the U-10 w/o Mo binary alloy does not occur in the U-7.5 w/o Nb-2.5 w/o Zr alloy welds.

Table IX shows that the gamma phase alloy in the as-welded condition possesses excellent mechanical properties. The strengthening of the weld which occurs as a result of some transformation on cooling to room temperature may be enhanced by post-heating at a low temperature.

STRESS-CRACKING

Static loading or stressing at low strain rates in environments containing oxygen⁴ have no detrimental effects upon the room-temperature tensile properties of the U-7.5 w/o Nb-2.5 w/o Zr alloy. Table X shows the properties after preloading 200 hours at 100,000 psi compared with results on samples that were not preloaded. The samples tested in air give the same results as those tested in vacuum. As would be expected, tests conducted at a strain rate of 0.0005 in./in.-min had a slightly lower yield stress and a small increase in ductility.

Table IX. Tensile properties of TIG-welded, gamma-annealed U-7.5 w/o Nb-2.5 w/o Zr alloy plates ($3 \times 4 \times 0.350$ in.).

Condition	Yield strength, 0.1% offset (psi)	U. T. S. (psi)	% elong. in 1 in.
As welded	130,000	136,500	11
Post-heated 0.5 hour / 275°C	139,500	152,300	6

Table X. Tensile properties of U-7.5 w/o Nb-2.5 w/o Zr alloy.

Condition	Environment	Hardness, R _C	Strain rate (in./in.-min)	Yield strength (psi)	U. T. S. (psi)	Elong. %
Ann. 950°C/AC	Air	42	0.05	131,400	190,800	5.2
Ann. 950°C/AC	Vac.	42	0.05	134,300	191,300	5.2
Ann. 950°C/AC	Air	42	0.0005	113,500	185,500	6.5
Ann. 950°C/AC	Vac.	42	0.0005	110,300	186,900	6.3
Ann. 950°C/AC + 200 hours/100,000 psi	Air	42	0.05	135,400	197,000	5.0
800°C/vac./WQ	Air	21	0.05	100,300	128,300	12.5
800°C/vac./WQ	Air	21	0.0005	91,600	120,300	14.4
800°C/vac./WQ + 200 hours/100,000 psi	Air	21	0.05	108,300	130,300	12.0
950°C/AC + 8 hours/250°C	Air	45	0.05	153,600	210,600	4.0
950°C/AC + 8 hours/250°C + 200 hours/100,000 psi	Air	45	0.05	157,600	209,000	3.9
950°C/AC + 8 hours/300°C	Air	48	0.05	163,700	216,700	4.0
950°C/AC + 8 hours/300°C + 200 hours/100,000 psi	Air	48	0.05	159,700	217,900	3.8
950°C/AC + 8 hours/350°C	Air	54	0.05	190,100	240,000	3.6
950°C/AC + 8 hours/350°C + 200 hours/100,000 psi	Air	54	0.05	192,800	248,600	3.5

AC = air-cooled.

WQ = water-quenched.

STRUCTURE AND PHASE IDENTIFICATION

Metallography

Individual specimens were mounted in Bakelite, ground through 180, 320, 600, and 4-0 grit papers, and then wet-polished on a microcloth wheel with 1-micron diamond paste. The final mirror finish was given to the specimens by several hours of polishing with 0.05-micron Linde B alumina on a Syntron vibratory table covered with microcloth.

Cathodic etching was used to bring out cold-worked structures and grain boundaries of annealed and transformed structures as well as precipitates and inclusions. An electrolytic etch in phosphoric-ethylene glycol-ethyl alcohol solution was useful in giving contrast to some of the transformed structures, particularly pearlites. Polarized light is effective in identifying and photographing alpha phase, as shown in Figs. 9, 10, and 11.

X-Ray Diffraction

This method was used mainly for phase identification. When large enough in area, the metallographic samples were electropolished and this surface scanned on a General Electric XRD-5 diffractometer, using copper K_{α} radiation. The intensities at the various 2θ angles were recorded on a moving chart. Diffraction patterns from a quenched specimen and from one transformed at 600°C are shown in Figs. 22 and 23, respectively. The diffraction lines from specimens transformed at other temperatures were very broad, and for this reason no attempt was made to determine lattice parameters of the transformation products.

DISCUSSION

According to work by Dwight and Mueller² on the ternary uranium-niobium-zirconium system, a monoeutectoid valley should occur for an alloy of the composition U-7.5 w/o Nb-2.5 w/o Zr at about 638°C. This would represent the lower boundary of the gamma phase for this alloy under equilibrium conditions. According to the present data, the conditions of the DTA determinations in this study were not near equilibrium for cooling at the rate used (85°C per hour); thus the transformation of gamma phase was depressed to 500°C. This, however, indicates a sluggish-to-transform

X-Ray Diffraction Pattern of U-7.5 w/o Nb-2.5 w/o Zr

Fig. 22. Annealed at 900°C
1 hour and quenched. All
gamma phase.

Fig. 23. Annealed at 900°C 1 hour and isothermally transformed 8 hours at 600°C. Alpha and gamma phases.

alloy that is quite gamma-stable. During heating the peak of the reaction's heat absorption is at 623°C and the material is all gamma at 654°C. This is in quite good agreement with the equilibrium temperatures.

Both metallographic and x-ray techniques failed to detect more than one gamma phase. This indicates that the alloy avoids any miscibility gap of the gamma phases such as are present in the binary systems of uranium-niobium and uranium-zirconium.

Isothermal transformation of the gamma phase just below the critical temperature at 600°C proceeds by nucleation of alpha plates at grain boundaries followed by their inward growth towards the interior of the grains. The series of photographs in Appendix I illustrates this process. This pearlitic eutectoid is verified by x-ray diffraction to be alpha plus niobium-rich gamma (γ'). Transformation by nucleation of fine alpha within grains and grain boundaries takes place as low as 400°C. Below 400°C the characteristic microstructure, high hardness, and distorted x-ray patterns suggest transformation is by a martensitic or shear process. The gamma phase decomposition does not occur martensitically on quenching as it does in the uranium-zirconium binary alloy because of the presence of niobium. This mechanism does operate when a low enough temperature is reached during slow cooling or isothermal aging and is due to the presence of zirconium in the alloy.

The operation of the aforementioned mechanisms of transformation offers control of the mechanical properties by various heat treatments which have been discussed. The general overall corrosion resistance of this alloy in either the gamma or transformed condition has not been a subject of this study. It is known that the alloy is not susceptible to stress-cracking. It is reasonable to believe that the gamma alloy will be more tarnish-resistant in air than the transformed alloy, based upon behavior of polished metallographic samples.

CONCLUSIONS

The incorporation of niobium and zirconium with uranium in the proportions U-7.5 w/o Nb-2.5 w/o Zr has resulted in an alloy which by judicious control of heat treatments may have a wide range of properties. A soft, ductile alloy, suitable for severe fabrication operations, is produced by gamma-annealing and quenching. It may then be strengthened to satisfactory

strength and toughness levels by aging at various temperature levels. The alloy possesses excellent atmospheric corrosion resistance and is not subject to stress-cracking in oxygen-bearing environments.

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APPENDIX I

Optical Photomicrographs and Electron Micrographs of Gamma-
Annealed and Isothermally Transformed U-7.5 w/o
Nb-2.5 w/o Zr Alloy

TRANSFORMATION STRUCTURE OF THE ALLOY URANIUM-7 $\frac{1}{2}$ w/o NIOBIUM-2 $\frac{1}{2}$ w/o ZIRCONIUM

SPECIMEN HISTORY

Annealed at 900°C for 1 hour, aged at 600°C
for the indicated time, and water quenched.

ETCHING

Electro-etched in an orthophosphoric acid
solution.

REPLICATION

Direct positive carbon with Pt-Pd shadowing.

A-MATRIX

Niobium rich b.c.c. solid solution.

B-TRANSFORMATION PRODUCT

α Uranium.

TRANSFORMATION STRUCTURE OF THE ALLOY URANIUM-7 $\frac{1}{2}$ w/o NIOBIUM-2 $\frac{1}{2}$ w/o ZIRCONIUM

One hour
Photomicrograph

10 μ

One hour
Electron micrograph

1 μ

TRANSFORMATION STRUCTURE OF THE ALLOY URANIUM- $7\frac{1}{2}$ w/o NIOBIUM- $2\frac{1}{2}$ w/o ZIRCONIUM

Two and one half hours
Photomicrograph

10 μ

Two and one half hours
Electron micrograph

1 μ

TRANSFORMATION STRUCTURE OF THE ALLOY URANIUM- $7\frac{1}{2}$ w/o NIOBIUM- $2\frac{1}{2}$ w/o ZIRCONIUM

Eight hours
Photomicrograph

10 μ

Eight hours
Electron micrograph

1 μ

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