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RFP-1652

December 21, 1971

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TRANSFORMATION KINETICS AND MECHANICAL
PROPERTIES OF THE URANIUM-7.5
NIOBIUM-2.5 ZIRCONIUM TERNARY ALLOY

Ross J. Jackson
Joseph F. Boland

DOW CHEMICAL U.S.A.
ROCKY FLATS DIVISION
P. O. BOX 888
GOLDEN, COLORADO 80401

U. S. ATOMIC ENERGY COMMISSION
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Printed in the United States of America
Available from the
National Technical Information Service
U. S. Department of Commerce
Springfield, Virginia 22151
Price: Printed Copy \$3.00; Microfiche \$0.65

December 21, 1971
Publication Date

RFP-1652
UC-25 METALS, CERAMICS
AND MATERIALS
TID-4500 - 56th Ed.

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P. O. BOX 888
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Prepared under Contract AT(29-1)-1106
for the
Albuquerque Operations Office
U. S. Atomic Energy Commission

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CONTENTS

Abstract	1
Introduction	1
Experimental	1
Alloy Preparation	1
Testing and Heat Treating Procedure	1
Results	2
Presentation of Tensile Data	2
Presentation of Impact Test Data	4
X-ray Diffraction Results	4
Metallographic Results from Non-Deformed Alloys	5
Metallographic Results from Deformed Alloys	6
Discussion	7
References	9

ACKNOWLEDGMENTS

The authors wish to acknowledge Mr. R. Lucas for technical assistance in the metallographic and x-ray diffraction aspects of the study.

TRANSFORMATION KINETICS AND MECHANICAL PROPERTIES OF THE URANIUM-7.5 NIOBIUM-2.5 ZIRCONIUM TERNARY ALLOY

Ross J. Jackson and Joseph F. Boland

Abstract. Tensile and impact test data are presented for the U-7.5 wt% Nb-2.5 wt% Zr ternary alloy in the gamma quenched and aged conditions. The mechanical test data are correlated with metallographic and x-ray studies. These alloys, although transition-phase, supersaturated solid solutions, are weakest in the as-quenched condition. Aging the alloys for up to 80-hour periods at temperatures below 400°C results in a noticeable increase in strength with no loss in supersaturation. The data are interpreted in terms of plastic deformation occurring by the motion of twin interfaces and in the pinning of twin interfaces by ordering on aging. A wide variety of mechanical properties are available from this heat treatable and corrosion-resistant alloy, and it shows promise as an economic, high density, engineering material in both nuclear and non-nuclear applications.

INTRODUCTION

This investigation of the uranium-niobium-zirconium system was undertaken as part of a general program to find a uranium-rich alloy possessing: (a) excellent corrosion resistance, (b) heat treating characteristics leading to favorable mechanical properties, (c) formability, (d) dimensional stability, (e) low neutron capture cross-section, and (f) high density.

A search of the literature revealed a number of significant research studies dealing with uranium-niobium-zirconium alloys. A uranium rich U-Nb-Zr ternary diagram has been reported.¹ Other aspects of the system such as phase stability and decomposition,² physical properties,^{3,4} corrosion,⁵ stress corrosion,⁶ and mechanical properties^{3,4,7} have been investigated.

Hoge⁷ investigated the effects of dynamic tension loading on the mechanical properties of the U-7.5Nb-2.5Zr alloy. Hoge found that tensile strength increased moderately with increasing strain rates (10^{-3} to 160 in./in.-sec.), but more important was the fact that dynamic loading caused only a slight decrease in ductility. Peterson and others^{3,4} observed the effects of short time aging (generally 8 hours or less) on tensile properties of the U-7.5Nb-2.5Zr alloy. It was the intent of the present study to: (a) extend the work of Peterson and others,^{3,4} and (b) optimize mechanical properties as a function of aging time and temperature,

(c) relate strengths to microstructure and crystal structure, and (d) postulate a strengthening mechanism.

EXPERIMENTAL

Alloy Preparation

The uranium was magnesium-reduced material and averaged 10 ppm Al, 60 ppm C, 30 ppm Fe and 30 ppm Si. The niobium was electron-beam melted material rolled to 120-mil strip. The niobium averaged 40 ppm C, 5 ppm H, 50 ppm N, 80 ppm O, 10 ppm Fe, 15 ppm Mo, 10 ppm Ni, 250 ppm Ta and 50 ppm W. The zirconium was of commercial purity.

The component metals were fabricated into box electrodes. These electrodes were arc-melted in a consumable electrode arc furnace using a water cooled copper mold. The resulting ingots were inverted and again arc-melted. The alloy was then vacuum induction melted in a zirconia coated graphite crucible and bottom poured into a preheated zirconia coated graphite mold. The vacuum induction casting step was an effort to minimize macrobanding in the cast alloy. The double arc-melted and induction cast alloy was homogenized 4 hours at 1150°C. The purity of the resulting alloy is listed in Table 1.

Table 1. Typical Chemical Analysis of Alloys Tested in this Study.

Nb	7.34 wt%	Cd	<10	Mo	<10
Zr	2.50 wt%	Co	<10	Ni	<25
C	40 ppm	Cr	<50	Pb	3
Al	60	Cu	5	Sb	<10
B	<1	Fe	110	Si	180
Ba	<100	K	<10	Sn	<1
Bi	<10	Li	<1	Ti	<50
Be	3	Mg	<1	V	<10
Ca	<25	Mn	<10	Zr	<50

Testing and Heat Treating Procedure

A prior gamma grain size averaging about 120-160 microns and somewhat random orientation was obtained by upset forging at 800°C and water quenching. Tensile and impact test blanks were machined from the hot-forged plate. The test blanks were held 30 minutes at 900°C, and then water

quenched. The blanks were aged at various times and temperatures, generally in air, and air cooled to ambient temperatures, and then machined to final dimensions. It was believed that this procedure would be similar to that of a fabrication process.

For the impact tests, a standard vee-notch Charpy bar was used. Testing was conducted on a Man-Labs 0-25 ft-lbs impact testing machine. A round tensile test bar was used. The bar was 3-inches long with $\frac{3}{8}$ by 16-inch threads. The reduced section was 0.252 inch in diameter and $1\frac{1}{4}$ inches long with a gage length of 1.00 inch. A 0.18-inch fillet radius connected the grip and reduced sections. Tensile tests were performed at room temperature in air on an Instron 10,000-pound tensile tester at a strain rate of 0.005 in./in. per minute. Except where noted, only one specimen was tested for each datum point. A strain gage extensometer was used. The reduction in area was obtained by averaging readings taken with a point micrometer at the smallest cross section. The values were reproducible to $\pm 2\%$. The modulus values were taken from the stress-

strain curve, and, although not highly accurate, are representative.

After fracture, sections of selected specimens were metallographically examined at both the grip and fracture areas. The grip areas were subjected to x-ray examination using filtered CuK_α radiation and a diffractometer. Hardness measurements were taken of the grip area using a Leitz Miniload Tester and a 200-gram load. The hardness values quoted are an average of eight readings.

RESULTS

Presentation of Tensile Data

The tensile test data for 900°C gamma-quenched and 80 hour aged at temperatures of 50 to 540°C are shown in Figure 1. Similar tensile test data for 30-hour and 6-hour ages at temperatures of 50° and 320°C are shown in Figures 2 and 3, respectively. The shape of the curves for the three aging

Figure 1. Tensile Test Data for Gamma-Quenched and 80-Hour-Aged U-7.5 wt% Nb-2.5 wt% Zr Alloy.

Figure 2. Tensile Yield and Ultimate Strength Data for Gamma-Quenched and 6- and 30-Hour-Aged U-7.5 wt% Nb-2.5 wt% Zr Alloy.

Figure 3. Tensile Elongation, Reduction in Area, and Elastic Modulus Data for Gamma-Quenched and 6- and 30-Hour-Aged U-7.5 wt% Nb-2.5 wt% Zr Alloy.

times are similar. The alloy shows a relatively low strength and high ductility in the solution quenched condition. In general, aging below 300°C results in a notable strengthening while maintaining reasonable good ductility properties. At 300° to 450°C the alloy develops a very high strength and almost complete loss of ductility. Aging at higher temperatures shows an improvement in ductility properties with a corresponding loss in strength.

To a first approximation, in Figures 1 through 3, the ultimate tensile strength, yield strength, reduction in area, and modulus are linear with temperature. For example, the yield strength is given by:

$$\text{yield strength} = 0.567 C + \frac{2}{3} t \text{ (hours)}.$$

Similar expressions can be written for the ultimate tensile strength, reduction in area, and elastic modulus. The percent elongation is proportional to C^{-2} .

In the high-temperature aging experiments, reasonably high strengths combined with good ductility properties are obtainable. The major disadvantage of this treatment is the precipitation of a niobium-poor alpha-uranium phase and an expected loss in corrosion resistance to normal environments. The most promising heat treatment for alloys where corrosion resistance is paramount is the low-temperature heat treatment, i.e., below 300°C. At temperatures of 200°C and less, no visible decomposition of the metastable γ^0 structure occurs during the 6-, 30- or 80-hour aging. These alloys show a noticeable increase in strength while maintaining good ductility.

The selection of 6, 30 and 80 hours for aging time was arbitrary. Longer times at a given temperature gave rise to a continued change in properties. Tensile test data for the gamma-quenched alloy aged for various periods of time at 200°C are shown in Figure 4. This low-temperature heat treatment results in a notable increase in strength properties and a decrease in ductility properties. To a first approximation, the tensile properties are linear with respect to the log of the aging temperature.

Figure 4. Tensile Test Data for Gamma-Quenched and 200°C Aged U-7.5 wt% Nb-2.5 wt% Zr Alloy.

For the U-7.5Nb-2.5Zr alloy, one of the best combinations of properties was obtained after aging at 180°C for 30 hours. This treatment yielded values in excess of 170 ksi tensile strength, 150 ksi yield strength, 10 ksi elastic modulus, 9% elongation (4 diameter), and 35% reduction in area, with no visible change in microstructure.

The tensile properties of butt welded U-7.5Nb-2.5Zr alloy are shown in Table 2. The alloy was in the 900°C water-quenched condition prior to welding and did not receive a post-weld heat treatment. It is seen that the inert-gas, tungsten-arc welds show slightly better tensile properties than do the electron-beam welded samples. In both types of welds, penetration was 100%; however, the fusion zone and heat-affected zone was noticeably larger for the inert-gas, tungsten-arc welds. The lower tensile properties for electron beam welded samples may be due to the observed presence of very small spherical voids in the fusion zone of the electron beam welds. Nonetheless, the tensile properties for both types of welds, although changed somewhat, do compare favorably with the tensile properties of the parent material, and one can conclude that the alloys show good weldability.

Table 2. Tensile Properties of Butt Welded U-7.5 wt% Nb-2.5 wt% Zr Alloy. Prior to welding the alloy was in the 900°C water-quenched condition. The welds did not receive a post-weld heat treatment.

Welding Procedure	No. of Samples	Ultimate Tensile Strength (ksi)	0.2% Offset Yield Strength (ksi)	Reduction in Area (%)	Elongation (%)
Non-Welded Parent Metal	5	120	66	52	26
Inert-Gas Tungsten-Arc	3	125	70	12	36
Vacuum-Chamber Electron-Beam	4	127	62	8	25

Presentation of Impact Test Data

The Charpy vee-notch impact test data for 900°C gamma-quenched and 30-hour aged at 50° to 320°C are shown in Figure 5. The bars air quenched from 900°C have an impact strength approximately one-half that of water quenched bars. The water quenched and aged bars show a decreasing impact strength with increasing aging temperature. As was the case with the tensile yield strengths, the impact-strength, aging-temperature relationship shows an approximate linear dependency.

Figure 5. Charpy Vee-Notch Impact Strengths for Gamma-Quenched and 30-Hour Aged U-7.5 wt% Nb-2.5 wt% Zr Alloy.

The selection of 30 hours for aging time was arbitrary. Longer times at a given temperature gave rise to a continued change in properties. Charpy vee-notch impact test data for the gamma-quenched alloy aged for various periods of time at 200°C are shown in Figure 6. This low-temperature heat treatment results in a decrease in impact strength with increasing aging time. To a first approximation, the tensile

Figure 6. Charpy Vee-Notch Impact Strengths for Gamma-Quenched and 200°C Aged U-7.5 wt% Nb-2.5 wt% Zr Alloy.

properties are linear with respect to the log of the aging temperature.

X-ray Diffraction Results

A number of stable and metastable phases are observed in the U-7.5 wt% Nb-2.5 wt% Zr alloy. The equilibrium phases and relationships are apparent from the equilibrium diagram.¹ The metastable transition phases observed in the U-Nb binary system are discussed elsewhere.⁸ The nomenclature and description of phases observed in this study are listed in Table 3.

Table 3. Nomenclature and Description of Phases Observed in the U-7.5 wt% Nb-2.5 wt% Zr Alloy.

Equilibrium Phase	Metastable Transition Phase	Description of Phase
α (Orthorhombic)		The equilibrium α uranium phase and contains less than 1 wt% of niobium and zirconium.
γ_1 (bcc)		The equilibrium γ_1 uranium phase. This is the bcc solid solution containing in this case 7.5 wt% Nb-2.5 wt% Zr which is stable above 640°C.
γ_2 (bcc)		The equilibrium γ_2 uranium phase. This is the niobium-rich bcc solid solution containing approximately 4 to 8 wt% Nb-2.5 wt% Zr at room temperature.
${}_1\gamma_2$ (bcc)		A bcc phase having a niobium composition intermediate to that of γ_1 and γ_2 .
	α'' (monoclinic)	A supersaturated α -uranium transition phase. There is a notable contraction of the b_0 parameter, expansion of the a_0 and c_0 parameters, and an increase in cell angle between the a and c axes thus resulting in a monoclinic crystal symmetry with "c" axis unique. The gamma-quenched U-6.5Nb-2.5Zr alloy is α'' .
	γ° (tetragonal)	A tetragonal transition phase that is similar to the bcc γ phase except two of the cell axes have expanded while the third has contracted. The structure is based on a block of 2 by 2 by 1 γ -like bct subcells with the supercell having $c/a < 0.5$. The gamma quenched U-7.3Nb-2.5Zr is γ° .

X-ray diffraction results from the grip area of selected tensile specimens are listed in Table 4. The crystal structure of the as-quenched alloy is tetragonal γ° . To a first approximation, one can consider this a "body-centered tetragonal" cell with the c/a ratio only slightly less than one. Thus, most of the normal bcc diffraction maxima have split into two diffraction maxima that are relatively close together. On aging below 220°C, there is a gradual broadening of the overlapping diffraction pairs. The degree of broadening is proportional to both temperature and time at temperature. The broadening of overlapping diffractions can be interpreted as the lattice undergoing a decrease in c/a ratio. This could be caused by an ordering reaction in which very small

Table 4. Metallographic and X-ray Diffraction Results from Selected Tensile Specimens. All samples were water-quenched from 900°C and aged 80 hours at the indicated temperatures. The prior gamma grains were equiaxed and had an ASTM 4 grain size.

Aging Temperature (80 hours)	X-Ray Phases*	Microstructure at Grip (all equiaxed grains)	Microstructure at Fracture Zone	Fracture Mode
As W. Q.	γ°	1 phase, no bands, no sub-g.	banded elong. g. (3.4/1)	Trans.
70C	$\gamma^\circ(b)$	1 phase, no bands, no sub-g.	banded elong. g.	Trans.
100C	$\gamma^\circ(b)$	1 phase, no bands, no sub-g.	banded elong. g.	Trans.
160C	$\gamma^\circ(b)$	1 phase, no bands, no sub-g.	banded elong. g. (2.2/1)	Trans.
220C	$\gamma^\circ(b) + \alpha''(?)$	weak banding, no sub-g.	banded elong. g. (1.4/1)	Trans.
260C	$\gamma^\circ(b) + \alpha''(b)$	banded, no sub-g.	banded elong. g. (1.3/1)	Mixed
300C	$\gamma^\circ(b) + \alpha''(b)$	banded, no sub-g.	banded elong. g. (1.1/1)	Mixed
360C	$\alpha''(b) + \gamma^\circ(b)$	strong banding, no sub-g.		Mixed
450C	$\alpha(b) + {}_1\gamma_2$	mottled matrix		Inter.
510C	$\alpha + {}_1\gamma_2$	2% Pearlite at g.b., mottled matrix		Inter.
540C	$\alpha + {}_1\gamma_2$	3% Pearlite at g.b., mottled matrix		Inter.

*Abbreviations: W.Q. = water quenched; Trans. = transcrystalline; Inter. = intercrystalline; g.b. = grain boundaries; incl. = inclusions; ppt. = precipitate; (b) = broad; (s) = sharp; (t) = trace; (w) = weak.

volumes on a microscale undergo an ordering of niobium and zirconium solute atoms. An increase in order would decrease the c/a ratio.

Aging above 220°C but below 400°C results in the appearance of the a'' monoclinic transition phase. The effect of composition on the observed transition phases a' , a'' , or γ^0 has been discussed elsewhere.⁸ It is possible that the heterogeneous ordering reaction results in a solute decrease sufficient to form monoclinic a'' . The a'' structure occurs at a niobium composition below U-6.9 wt% Nb-2.5 wt% Zr. Thus, a large change in composition is not required to form a'' . It is also possible that the coherency strains developed between the high and low solute volumes favors the formation of the a'' structure. It is more likely that a'' is the structure that results when the alloy becomes highly ordered.

Above 400°C, the quasi-equilibrium precipitates, $\alpha + \gamma_2$, form the x-ray diffraction pattern. The α structure is near equilibrium composition but could also be interpreted as a very low solute a' structure. The γ_2 structure is a gamma-like structure that is intermediate in composition to equilibrium γ_1 and γ_2 . The γ_2 structure gradually changes towards the γ_2 composition with increasing time and temperature. Eighty hours at 540°C is not sufficient time to form equilibrium γ_2 .

Metallographic Results From Non-Deformed Alloys

Metallographic results from the grip and fracture zones are listed in Table 4. The as-quenched alloy has an equiaxed, single-phase microstructure free of bands and subgrains (Figure 7a). The prior gamma grain size was ASTM 4. On aging below 220°C, there is no apparent change in microstructure. On aging above 220°C but below 400°C, the prior gamma grains become banded (Figure 7b). The intensity of the banding increases with temperature and time at temperature. The appearance of the bands coincided with the x-ray observation of the appearance of a'' . Perhaps the bands represent a'' plates forming in a γ^0 matrix by a shear transformation. It is more likely that the bands are twins and serve as a self-deformation process for the removal of ordering stresses. The increasing intensity of the bands with temperature and time at temperature could be partly due to an accelerated precipitation or pre-precipitation reaction at the band interfaces.

On aging above 400°C, the quasi-equilibrium precipitates $\alpha + \gamma_2$ are apparent in optical micrographs. The precipita-

tion is general in nature and results in a very fine and uniformly distributed product. The individual particles are not resolved by the light microscope but they cause a general darkening of the etched parent grain.

Above 500°C, a discontinuous pearlitic-like reaction occurs at grain boundaries. The individual lamellae are not resolvable with light optics but are readily seen with surface replicas and electron optics. The products of the discontinuous precipitation are believed to be $\alpha + \gamma_2$. These are the products observed in the general precipitation except now the γ structure is more solute rich. Both discontinuous (pearlitic-like) and general precipitation are observed in alloys aged above 500°C. An alloy that has transformed by both mechanisms is shown in Figure 7c. It is likely that the discontinuous precipitation is a recrystallization process which serves to remove the transformation stresses and at the same time achieve a more equilibrium phase composition.

Metallographic Results From Deformed Alloys

The microstructure at the fracture zone and the fracture mode is listed in Table 4. At the fracture zone, the samples that were not banded in the grip area now have banded grains (see Figure 8a). These bands are more distinct than those formed during 200-400°C aging, but otherwise are similar. The bands in the fracture zone are most likely deformation twins. This observation supports the argument that the bands appearing on aging are self-deformation twins serving to remove transformation stresses. For the samples aged below 300°C, the prior-gamma grains were elongated in the direction of pulling. The length-to-width ratio of these distorted grains is shown as a function of aging temperature in Table 4.

The fractures of solution quenched specimens and specimens aged below 260°C were predominantly transcrystalline. Specimens aged for 80 hours at 260-360°C showed a mixture of transcrystalline and intercrystalline fractures. Specimens aged for 80 hours above 400°C fractured in a predominantly intercrystalline manner. Micrographs showing the change from transcrystalline to intercrystalline fracture with increase in aging temperature are shown in Figure 8. The change from transcrystalline to intercrystalline fracture is attributed to the strengthening of the matrix on aging which causes a relative weakening of the grain boundaries.

Figure 7. Microstructures of Quenched and Aged U-7.5 wt% Nb-2.5 wt% Zr Alloy Prior to Deformation. (a) Sample in as-quenched condition showing equiaxed, single-phase microstructure free of bands and subgrains, 100X; (b) Sample aged 80 hr at 360°C showing banded microstructure developed on aging, 100X; and (c) Sample aged 80 hr at 540°C showing general precipitation within grains and localized precipitation at grain boundaries, 200X.

Figure 8. Microstructures Illustrating Change in Appearance of Fracture Surface (Right Edge of Microstructure) for the U-7.5 wt% Nb-2.5 wt% Zr with Increase in Aging Temperature. (a) Tensile bar aged 80 hr at 160°C showing elongated grains and transcrystalline fracture 100X, (b) Sample aged 80 hr at 300°C showing mixed transcrystalline and intercrystalline fracture 100X, and (c) Sample aged 80 hr at 540°C showing intercrystalline fracture, 200X.

DISCUSSION

The yield strength, tensile strength, and elongation for pure uranium average 30 ksi, 83 ksi, and 12%, respectively.⁹ The same properties for the gamma quenched U-7.5Nb-2.5Zr

alloy are 66 ksi, 120 ksi, and 26%. The improved ductility can be attributed to the increased number of slip or twin planes present in the higher symmetry tetragonal γ° phase. The increased strengths would be due to the increased difficulty of moving a dislocation on its slip plane.

The critically resolved shear stress for dislocation slip must be increased significantly on alloying since the alloy deforms by a twinning mechanism. The conclusion is based on the following observations: (a) the material exhibits a thermally-induced and reversible shape-memory behavior, i.e., a dimensional memory for the as-quenched shape; (b) a lattice reorientation with no change in unit cell parameters occurs during an externally induced plastic deformation; and (c) a surface tilt accompanies a plastic deformation. Thus, it is suggested that the increased strength on alloying is due to an increased critically resolved shear stress for dislocation slip due to solid solution and order hardening; and the increased elongation is due to a high mobility of twin interfaces and to the large number of available twinning planes.

On aging below 200°C the alloy undergoes a partial ordering of niobium and zirconium atoms. Electron diffraction evidence for ordering in this temperature range has been presented by Peterson.⁴ The ordering is considered heterogeneous in nature. This gives rise to coherency strains between ordered and disordered regions and increased difficulty in moving a twin interface or dislocation, and hence, an increase in strength. Above 200°C but below 400°C, the increased thermal energy allows greater order and the monoclinic a'' structure forms. The ordering stresses are quite large and the alloy undergoes a self-deformation process by twinning to relieve these stresses. This is what causes the highly banded microstructures (Widmanstätten-like structure) that appears on aging in this temperature interval. Here again, the coherency strains causes the large increases in strength.

The increase in yield strength on aging below 400°C is probably partially due to the diffusion of interstitial impurities to twinning dislocations and interfaces making them less mobile, thus partially causing the increase in yield strength. The changes in elastic modulus that occur on aging are believed due to changes in bonding (i.e., small changes in unit cell parameters at 200°C and below; and changes in symmetry and composition that occur at the higher aging temperature), and to changes in the elasticity of twin interfaces (due to pinning action of interstitial impurities and ordering).

On aging above 400°C, the quasi-equilibrium $\alpha + \gamma_2$ precipitates are apparent in optical micrographs and x-ray diffraction patterns. The precipitation is general in nature and results in a very fine and uniformly distributed product. This precipitation results in a lowering of strengths. Above 500°C, a discontinuous pearlitic-like reaction occurs at grain boundaries. The products of the discontinuous precipitation are believed to be $\alpha + \gamma_2$. Both discontinuous (pearlitic-like) and general precipitation

are observed in alloys aged above 500°C. It is likely that the discontinuous precipitation is a recrystallization process which serves to remove the transformation stresses and at the same time achieve a more equilibrium phase composition. Accompanying this secondary precipitation reaction is a decrease in strength properties.

The relatively low extensions and high reduction in areas that occurs in alloys aged below 300°C are worthy of note. This suggests that once deformation has started, it is considerably easier to continue deforming in that area than to initiate deformation elsewhere. It is suggested that this effect is due to an interstitial atmosphere pinning the twin interfaces, which, once freed, give rise to severe localized deformation.

Microstructural examination revealed no evidence of circumferential cracking. At similar compositions in the uranium-molybdenum system,¹⁰ severe circumferential cracking was observed and was attributed to stress-corrosion in air. This phenomenon has limited the use of uranium-molybdenum alloys. It is noteworthy that the uranium-niobium-zirconium alloy, at least in the range of heat treatment studied, did not show circumferential cracking.

The U-7.5 wt% Nb-2.5 wt% Zr alloy will stress corrosion crack in specific environments. The alloy will stress corrosion crack at relatively low stress levels in water containing small amounts of both dissolved oxygen and chloride ion. When the alloy is properly stress relieved (850°C for one hour) and not heavily loaded for long periods of time, one need not consider the possibility of stress corrosion cracking in atmospheric environments. In welded structures, one should be very careful to fully stress relieve the part before subjecting it to long times in a moist atmosphere. Likewise, long-term, continuous high-load applications should be avoided until the harmful corrodants are well defined.

The U-7.5 wt% Nb-2.5 wt% Zr alloy in the gamma quenched and aged condition (i.e., aged below 400°C) is considerably more corrosion resistant to atmospheric and certain other environments than is unalloyed uranium. Accelerated corrosion tests consisting of immersion in 0.1N HCl at 70°C for two weeks yielded an extrapolated mils-per-year loss of 1.5 and 547 for the gamma-quenched U-7.5Nb-2.5Zr alloy and pure uranium, respectively. Alloys aged below 400°C yielded similar results, while alloys aged above 400°C yielded significantly higher weight losses, but still much less than unalloyed uranium. The corrosion resistance of the low-temperature aged alloys coupled with the wide range of desirable mechanical properties exhibited by these alloys suggests that uranium alloys possessing these characteristics be referred to as "stainless uranium alloys".

The alloy is readily formable by normal hot and cold metal forming techniques. If the alloy is cold formed (i.e., below 450°C), it must be reheated to the gamma region for one hour and cooled rapidly. If this is not done, the alloy is subject to dimensional instability and delayed cracking.

The alloy has a density of 16.6 g/cm³. In material applications where high density and high strength are the prime considerations, the uranium alloy considered in this report offers an economic, readily-formable, corrosion-resistant engineering material having a wide range of mechanical properties.

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