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THE EFFECT OF ALLOY ADDITIONS AND HEAT TREATMENTS ON THE MECHANICAL PROPERTIES OF U-0.5% Ti ALLOY

J. I. Federer

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J. I. Federer

FEBRUARY 1969

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J. I. Federer

ABSTRACT

Specific amounts of C, Fe, Ni, Si, Y, and H were added to a base U–0.5 wt % Ti alloy. The tensile properties of the alloys in various heat-treated conditions were then compared to the properties of the base alloy in a similar condition. In gamma-quenched alloys tested at 25 and 600°C, carbon contents of 0.014 and 0.028% did not substantially affect properties, although the effects were mostly detrimental. Silicon contents of 300 and 400 ppm caused small changes in properties similar to those caused by carbon, except that the elongation at 600°C was greatly increased. Iron, nickel, and yttrium each increased the yield and tensile strengths of gamma-quenched alloys; however, iron (340 ppm) was effective at both 25 and 600°C, while nickel (700 ppm) was mostly effective at 25°C, and yttrium (0.10%) was mostly effective at 600°C. Each of these elements decreased elongation at 600°C. Aging the base alloy and alloys containing iron and nickel additions within the alpha range resulted in a further increase in strength at test temperatures of 25 and 400°C. Hydrogen in the range 8 to 14 ppm weakened and embrittled alloys tested at 25°C.

Bend testing at 25°C of the base alloy containing an electron-beam weld-fusion zone showed that cracking originated in the weld at much lower strains than had occurred prior to failure in tensile tests of unwelded specimens having the base alloy composition.

INTRODUCTION

The most important use for uranium metal has been the primary fissionable material in nuclear fuel elements. Consideration has also been given to breeding composites in which uranium metal, depleted in ^{235}U, serves as the matrix and fertile material for a dispersed, fissionable compound, such as UO_2. In addition, the high density of uranium (19.07 g/cm^3) suggests use as shielding and applications in which

inertia or kinetic energy properties are important, such as ballistics. Unalloyed uranium rapidly loses strength with increasing temperature,[1] the strength at 500°C being only one-tenth that at 25°C. Improved strength at temperatures up to 300°C can be obtained by alloying with small amounts of aluminum and zirconium.[1] Strengthening of uranium with Mo, Ni, T, and V has also been studied.[2–4]

The subject of the present study is the mechanical properties of the U–0.5 wt % Ti alloy. Previous studies have shown that substantial improvements in the strength of uranium at 25°C can be obtained with additions of 0.1 to 1.4% Ti. Some of the previous tensile data[2–4] obtained at 25°C for alloys containing small additions of titanium and after being water quenched from the gamma phase at 800°C are shown in Fig. 1. Douglass et al.[3] studied mechanical properties as a function of titanium content in the range 0.12 to 1.22% Ti and obtained strength maxima near 0.2 and 1.0% Ti. Murphy,[4] however, found the strength continually increasing with increasing titanium content over the range 0.09 to 1.4% Ti. The impurities in the alloys studied by Murphy were about 65 ppm C, 70 ppm Fe, 45 ppm Al, 50 ppm Ni, 85 ppm Si, and much smaller amounts of other elements. The greatest ductility occurred in the studies of both Murphy and Douglass et al. near 0.5% Ti. Skelly et al.[2] prepared alloys containing 0.5% Ti, 40 ppm C, and possibly 0.001 to 0.002% Zr, 0.02% W, and 3 ppm Cu (all the melts were not analyzed for impurities). Tensile and yield strengths of equal magnitude and an elongation of only 1% were obtained at 25°C.

Douglass et al.[3] also tested alloys at 25°C that had been aged at 300, 400, and 500°C after being quenched from 800°C. Both tensile and yield strengths declined slightly with increasing aging temperature, but

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[1]H. A. Saller, F. A. Rough, and W. Chubb, The Properties of Uranium Containing Minor Additions of Aluminum and Zirconium, BMI-1113 (July 16, 1956).

[2]H. M. Skelly, C. F. Dixon, and N. S. Spence, Uranium Alloy Development for Non-Nuclear Application, Progress Report No. 3, CONF-650841-2 (1965).

[3]D. L. Douglass, L. L. Marsh, Jr., and G. K. Manning, Uranium-Titanium Alloy Studies, BMI-1189 (May 1957), pp. 39–40.

[4]D. J. Murphy, "Some Properties of Uranium-Low Titanium Alloys," Trans. Am. Soc. Metals 50, 884–904 (1958).

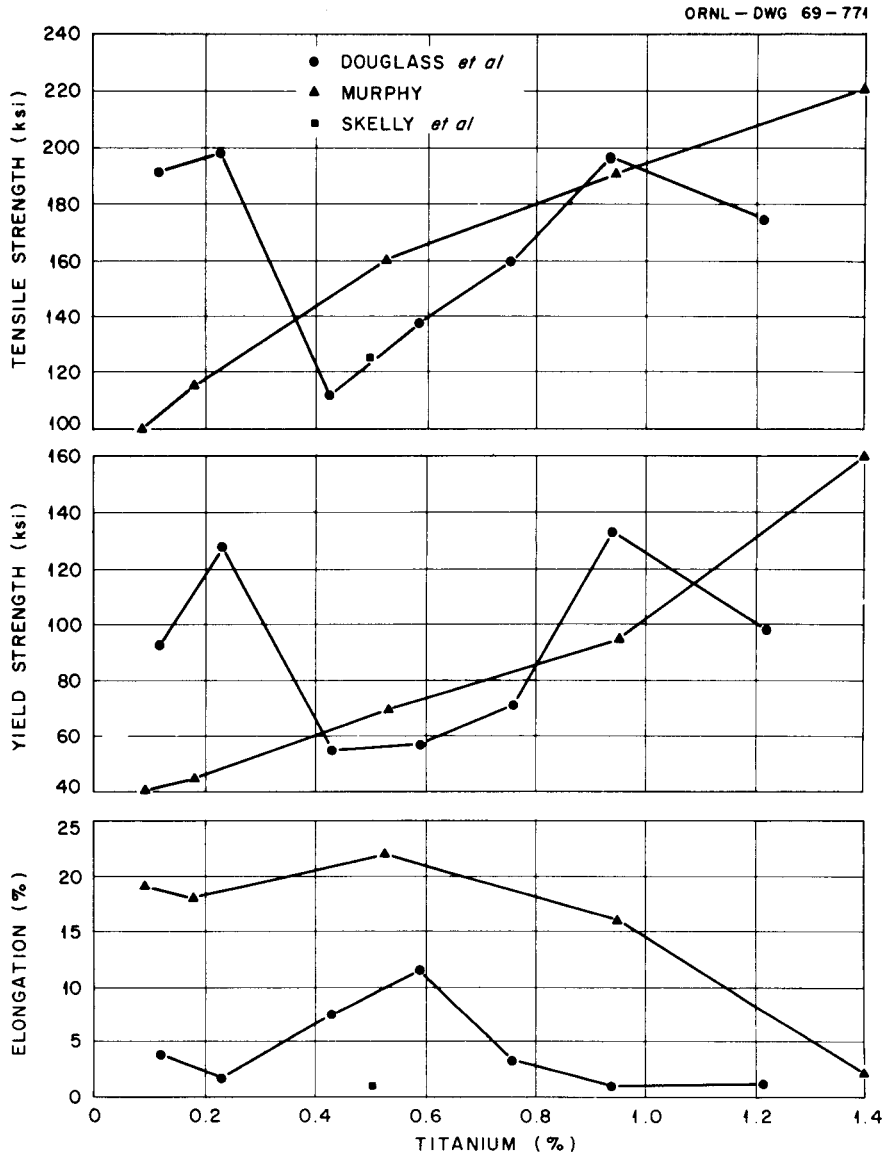

Fig. 1. Mechanical Properties of Uranium-Titanium Alloys at 25 °C After Being Water-Quenched from 800 °C.

exhibited maxima at about the same titanium contents as in the nonaged alloys. Murphy[4] concluded that optimum values of tensile and yield strength, ductility, and impact strength were obtained by water quenching from 800°C as compared with air, furnace cooling, and oil quenching. Skelly et al.[2] found that the hardness of the 0.5% Ti alloy increased from 318 VHN after water quenching from 800°C to 319, 367, and 390 VHN after aging 10 min at 400, 500, and 600°C, respectively.

Thus, previous results at 25°C show large discrepancies in tensile and yield strengths and in ductilities of uranium-titanium alloys. These discrepancies could be related to certain typical impurities or to impurities accidently introduced during preparation and fabrication of the alloys. Hydrogen, in particular, has a very detrimental effect on the ductility of unalloyed uranium and would be expected to affect the properties of dilute uranium alloys.

The objective of the present study was to evaluate the tensile properties of the U–0.5 wt % Ti (approx 2.40 at. % Ti) alloy at temperatures up to 600°C as affected by certain impurities and heat treatments. In order to limit the scope of this study, a limited number of alloys containing specific additions of either C, Fe, Ni, Si, Y, or H were tested. Each alloy was tested in the gamma-quenched condition, and some were tested after aging at several temperatures.

THE URANIUM–TITANIUM BINARY SYSTEM

A knowledge of the solubility of titanium in alpha-uranium in the temperature range 200 to 600°C is important to the interpretation of the results of this study; however, none of the phase diagrams which have been determined extend below 500°C. Portions of the diagrams determined by Udy and Boulger[5] and by Knapton[6] are shown in Fig. 2. The diagrams are similar, but differ in the extent of the solubilities in the alpha-, beta-, and gamma-uranium phases and in the homogeneity range

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[5]R. W. Buzzard, R. B. Liss, and D. P. Fickle, "The Uranium-Titanium Binary System," J. Res. Natl. Bur. Std. 50, 209–214 (1953).

[6]M. C. Udy and F. W. Boulger, Trans. AIME 200, 207–210 (1954).

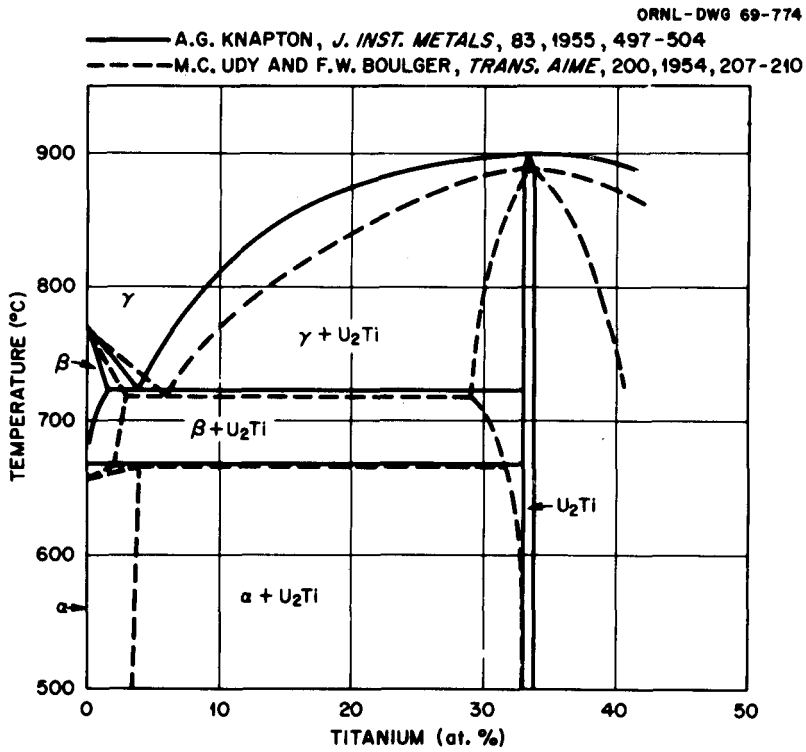

Fig. 2. Portion of the Uranium–Titanium Phase Diagram.

of the compound U_2Ti. Another version by Buzzard et al.[7] contained a uranium-rich high-temperature phase not substantiated by the other investigators. Udy and Boulger, however, used the solubilities of titanium in alpha-uranium determined by Buzzard et al. in the construction of their diagram. Knapton used iodide titanium and two grades of uranium – namely, bar material "which contained numerous coarse oxide particles" and a purer uranium powder which was "not completely free from oxide particles." Knapton obtained lower solubilities using the purer powder uranium-base alloys; the higher solubilities obtained with the less pure alloys agreed with those of Buzzard et al. Therefore, the phase boundaries obtained by Knapton with the purer alloys appear to be more reliable.

Knapton's results indicate that the solubility of titanium in alpha-uranium is 1 at. % (0.2 wt %) at 600°C, while Udy and Boulger's results indicate that the solubility is 4 at. % (0.8 wt %). If the solubility is as low as determined by Knapton, the U–0.5 wt % Ti alloy should exhibit an aging response due to the precipitation of the compound U_2Ti.

[7] A. G. Knapton, "The System Uranium-Titanium," J. Inst. Metals 83, 497–504 (1955).

EXPERIMENTAL PROCEDURE

Preparation of Alloys

Nine alloys weighing about 1800 g each were prepared by arc-melting and drop-casting. Each alloy charge was melted on a water-cooled copper hearth under a partial pressure of argon. The alloys were turned over in the hearth and remelted five or six times to ensure homogeneity before drop-casting into a water-cooled copper mold. Only titanium was added to uranium to prepare the base alloy. The other alloys were prepared by adding small quantities of C, Fe, Ni, Si, and Y to the base alloy. Each 1-in.-diam by 5-in.-long casting was cut in half and the pieces were placed inside close-fitting type 304 stainless steel tubes to which end plugs were welded. One end plug on each assembly had a 1/4-in.-diam evacuation tube connected to a vacuum pump. The stainless steel cladding was necessary to prevent spread of radioactivity and to protect the alloys during subsequent swaging operations. The welded assemblies were evacuated to 25 torr or less, then the evacuation lines were forged shut. The castings were then swaged from a 600°C furnace to a total reduction of about 65%. Swaging was accomplished with about 12 dies, the material being reheated to 600°C for passage through the die. Afterwards, three longitudinal grooves spaced about 120 deg apart were machined in the stainless steel cladding, permitting the cladding to be separated from the uranium alloy. The cladding separated without difficulty even though no parting compound was placed at the cladding–uranium alloy interface prior to swaging. Approximately 44 in. of 0.35-in.-diam rod was obtained in this manner from each casting. The rods were cut into 11-in.-long pieces for heat treatment.

The base alloy was also arc-melted and drop-cast into a 1- by 5/8- by 5-in. mold using the same procedures previously mentioned. The unprotected casting was heated to 600°C in a Na_2CO_3-K_2CO_3-$LiCO_3$ salt bath and reduced about 78% by rolling to strip measuring about 1 1/2 in. wide by 0.09 in. thick.

The composition of the alloys as vacuum annealed and gamma quenched (heat treatments are discussed later) is shown in Table 1, which also

Table 1. Analysis of Uranium-Titanium Alloys[a]

Element	Alloy Number								
	1	2	3	4	5	6	7	8	9
Ti, wt %	0.44(0.5)	0.54(0.5)	0.47(0.5)	0.45(0.5)	0.54(0.5)	0.51(0.5)	0.51(0.5)	0.48(0.5)	0.48(0.5)
C, wt %	0.003	0.028(0.06)	0.014(0.03)	0.003	0.004	0.003	0.002	0.015(0.03)	0.004
Y, ppm	< 0.1	< 0.1	< 0.1	< 0.1	< 0.1	< 0.1	< 0.1	< 0.1	0.1%
Si, ppm	30	10	10	400(300)	300(600)	30	100	200(300)	100
Fe, ppm	70	100	20	20	20	340(200)	70	220(150)	70
Ni, ppm	30	30	30	30	30	30	700(400)	300(100)	70
Al, ppm	10	100	30	3	30	100	30	100	30
Ca, ppm	50	50	20	0.3	50	20	0.5	20	2
V, ppm	0.7	2	0.7	0.7	0.02	2	0.7	2	30
W, ppm	0.8	8	8	3	2	8	30	8	30
H, ppm	b	b	0.34	b	0.11	b	0.14	b	b
N, wt %	0.0042	0.0021	0.0012	0.0021	0.0018	0.0020	0.0022	0.0016	0.0004
O, wt %	0.0070	0.0130	0.0061	0.0077	0.0079	0.0040	0.0072	0.0083	0.0039

[a]Precision values for metallic elements: Ti, ±0.02 wt %; C, ±0.0015 to 0.0002 wt % for 0.03 to 0.002 wt % C, respectively; Y, ±0.05%; others, 1/2 to 2 times listed value. Numbers in parentheses are intended contents.

[b]Analysis not obtained.

shows the intended content of certain elements in parentheses. The intended titanium content was 0.5 wt %; the actual titanium content ranged between 0.44 and 0.54 wt %. Actual carbon contents were about one-half the intended values, while the yttrium content was only one-fifth the intended value. The elements, Si, Fe, and Ni, were determined with less precision than Ti, C, and Y. When limits of precision are applied the determined contents of Si, Fe, and Ni include the intended values. The uranium melting stock used to prepare the alloys contained 10 ppm H, approximately two orders of magnitude higher than the hydrogen content of the vacuum-annealed alloys. The melting stock also contained 0.002 wt % of oxygen and nitrogen. The nitrogen content remained about the same after melting, fabrication, and heat treatment except in alloys 1 and 9. The oxygen content of each alloy was higher than that of the melting stock.

Heat Treatments

Uranium is embrittled by small amounts of hydrogen.[8] The initial heat treatment for all alloys used in this study was a vacuum anneal at 650 to 675°C for 24 hr for the purpose of minimizing the hydrogen content. Using the approximation that the concentration of a diffusing species attains one-half the original value when $x/Dt \approx 1$, where x is the diffusion distance, D is the diffusion coefficient of hydrogen in uranium, and t is the diffusion time, the appropriate vacuum annealing conditions can be estimated.[9] In this study, x = 0.175 in. (0.445 cm), the radius of the rod, and $D \approx 4 \times 10^{-5}$ cm^2/sec at 650°C (ref. 8). The pressure during annealing was 1×10^{-5} torr. For these conditions the concentration of hydrogen should attain one-half the original value in about 3.1 hr, or about four-thousandths the original value in 24 hr.

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[8] W. D. Davis, Solubility, Determination, Diffusion, and Mechanical Effects of Hydrogen in Uranium, KAPL-1548 (August 1, 1956).

[9] L. S. Darken and R. W. Gurry, Physical Chemistry of Metals, McGraw-Hill, New York, 1953, p. 444.

Beginning with 10 ppm H, the calculated final value is 0.04 ppm. Table 1 shows, instead, that the hydrogen content was reduced by only about one-hundredth the original value to 0.1 to 0.3 ppm.

In addition to the vacuum anneal, subsequent heat treatments were used to produce microstructural effects or to deliberately introduce hydrogen. The various heat treatments applied sequentially to the alloys in this study were as follows:

Alloys 1 through 9

1. vacuum annealed at 650 to 675°C, 1×10^{-5} torr, 24 hr, furnace cooled, and
2. gamma treated and quenched – that is, heated to 800°C in the gamma phase in high-purity argon (0.0003 vol % H_2), water quenched.

Alloys 1, 6, and 7

1. vacuum annealed and gamma quenched, and
2. aged at 200, 400, and 600°C in high-purity argon for 1 hr, water quenched.

Alloys 1, 6, and 7

1. vacuum annealed and gamma quenched, and
2. beta treated and quenched – that is, heated to 700°C in the beta phase in high-purity argon for 1 hr, water quenched.

Alloys 3 and 5

1. vacuum annealed, and
2. annealed at 800°C in a Na_2CO_3-K_2CO_3-$LiCO_3$ salt bath for 1/2 hr, water quenched.

Alloy 1 (rolled strip)

1. vacuum annealed and gamma quenched, and
2. aged at 200 and 400°C in high-purity argon for 1 hr, water quenched.

Alloy 1 (rolled strip)

1. vacuum annealed, and
2. annealed at 800°C in a 50 vol % H_2–50 vol % Ar atm for 1 hr, water quenched; the purpose of this heat treatment was to introduce hydrogen into the alloy.

Test Specimens and Procedures

Following the heat treatments the 0.35-in.-diam rods of the nine alloys were machined into tensile specimens measuring 1 7/8 in. long with a 1.125-in.-long gage length and 0.125-in.-diam reduced section. Duplicate specimens of each alloy in the gamma-quenched condition were tensile tested at 25 and 600°C and at a strain rate of 0.02 min^{-1} using an Instron machine. Testing at 25°C was in air, and testing at 600°C was in a vacuum in the range 1 to 9×10^{-5} torr. Specimens were heated to 600°C in about 30 min and held at 600°C for 15 min before testing was begun. The same pressures and heating conditions were used for aged specimens tested at 400 and 600°C.

The base alloy prepared in the form of 0.09-in.-thick strip was used for bend tests. The strip was first cut into 3-in.-long by 3/4-in.-wide specimens, which were vacuum annealed and gamma quenched as described previously. The specimens were electrolytically etched to remove quenching scale; then a full-penetration fusion zone measuring about 1/8 in. wide was run along the longitudinal center line using an electron-beam welder. Bend tests were then conducted on specimens in the as-welded, aged, and hydrogenated conditions.

The geometry of the bend test is shown in Fig. 3. Bending was produced in an Instron machine by applying a load, P, at the center of specimens of thickness, t, as shown in Fig. 3(a). The specimens rested on two fixed supports, S, separated by a distance, L. The load was applied by a 0.4-in.-diam rounded end punch moving at 0.02 in./min. The manner of determining the strain in the outer fibers is shown in Fig. 3(b). Application of the load through a small deflection, x, initially produces a curvature in the specimens of radius, r, due to elastic strain. The outer fibers, initially of length bd, stretch by an amount, dd'. Stretching of the fibers is graphically represented by rotation of point d to point d' through angle $d\theta$. Assuming that the neutral axis is at the midthickness of the specimen, the strain can be determined in terms of r and t.

$$\text{Elongation} = dd' , \qquad (1)$$

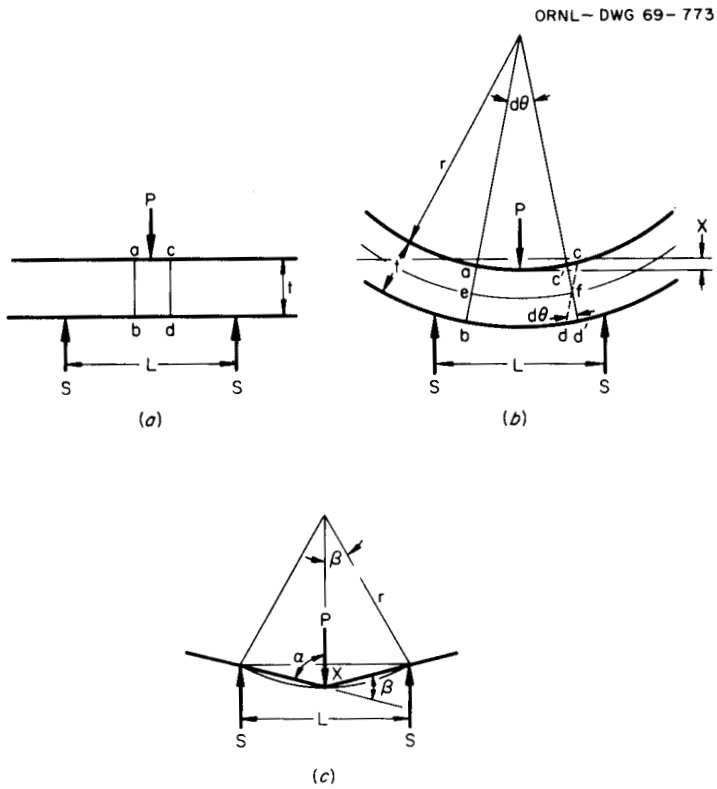

Fig. 3. Geometry of Deformation During Bending.

$$\text{Strain} = \frac{dd'}{ef} = \frac{\frac{1}{2}td\theta}{(r + \frac{1}{2}t)d\theta} = \frac{t}{2r + t} \ . \tag{2}$$

The value of r can be determined from geometrical considerations in Fig. 3(c). After a deflection of amount x the specimen is bent through the angle, β.

$$\sin\beta = \frac{L}{2r} \quad ; \quad r = \frac{L}{2 \sin\beta} \ . \tag{3}$$

Since L is a known quantity and the bend angle, β, can be measured directly from the specimens, the value of r can be determined.

The bend angle, β, can also be calculated from the deflection, x.

$$\tan \alpha = \frac{L}{2x} \tag{4}$$

$$\beta = 180 \text{ deg} - 2\alpha \ . \tag{5}$$

After the bend angle, β, is measured directly or calculated from Eqs. (4) and (5), the radius of curvature is calculated from Eq. (3). The value of r is initially large and decreases as bending occurs. If the specimen completely conformed to the contour of the punch used in this study, the maximum strain in the outer fibers would be approximately 18%.

RESULTS

Gamma-Quenched Alloys

Considerable variation in the microstructure of gamma-quenched alloys was observed. As shown in Fig. 4(a), alloy 1 contained acicular martensitic alpha-uranium in a fine-grained structure. Coarse particles of another phase(s), possibly oxide or carbide, were randomly distributed. The grain size of alloys 6 and 7 was larger by comparison and the structure was almost completely martensitic [Fig. 4(b) and (c)]. The other alloys also contained acicular martensite and had a grain structure similar to that of alloy 1. The reason for the differences in microstructures cannot be readily explained since each alloy had the same thermal and fabrication history.

Typical load versus elongation curves for gamma-quenched alloys tested at 25 and 600°C are shown in Fig. 5. Specimens tested at 25°C exhibited a tensile strength approximately two times the 0.2% yield strength while at 600°C the tensile strength was only about 10% higher than the yield strength. The large difference in tensile and yield strengths obtained at 25°C is due to strain hardening during the deformation process. The temperature of 600°C was too high for much work hardening to occur; therefore, tensile and yield strength were closer to the same value. Except in a few cases the elongation was substantially less at 600 than at 25°C.

Typical fractures obtained at 25 and 600°C are shown in Fig. 6. All specimens tested at 25°C broke without much reduction in area. The deformation prior to failure occurred over much of the gage length rather than the point of failure. On the other hand, specimens tested at 600°C

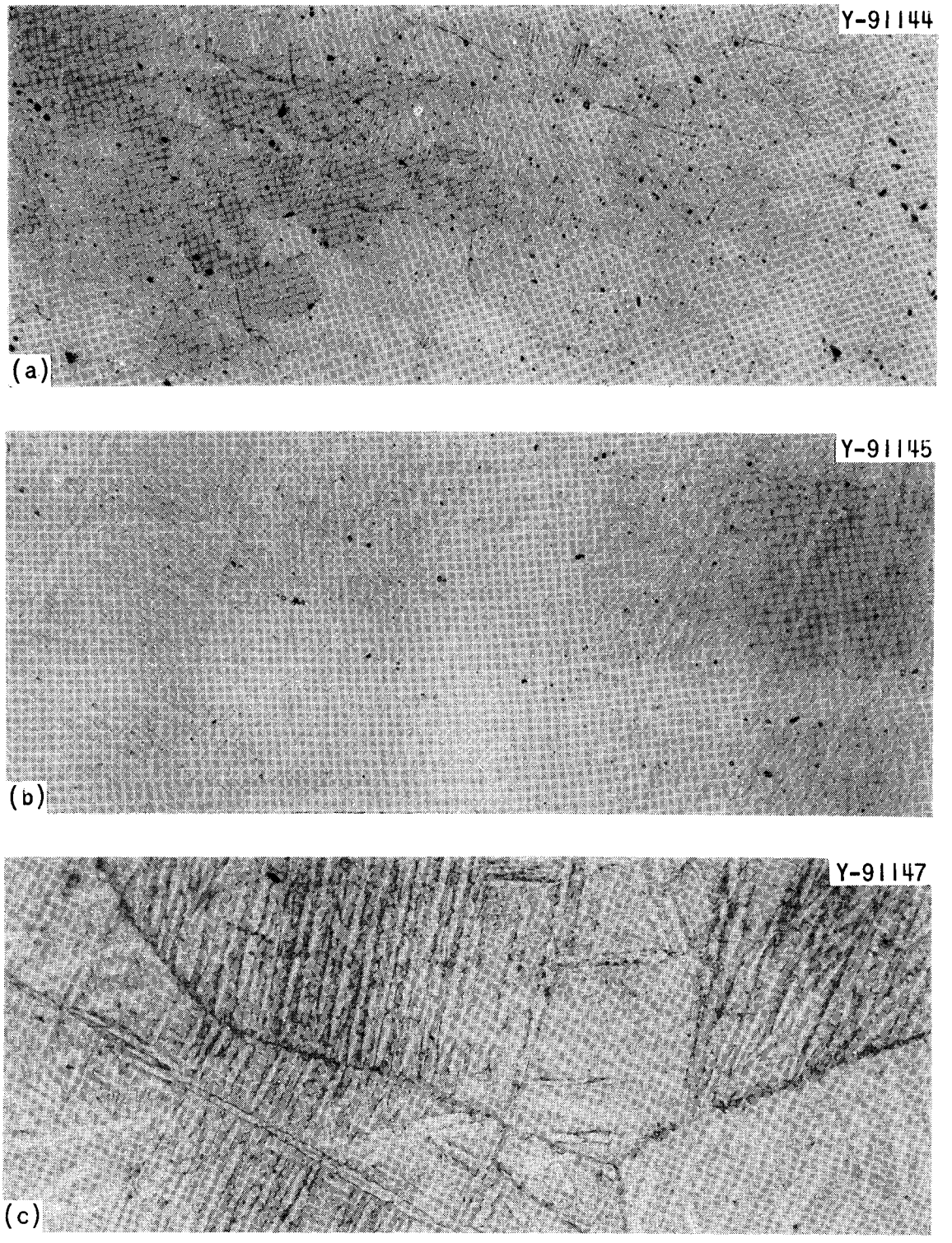

Fig. 4. Microstructures of Gamma-Quenched Alloys. Electrolytically etched in a solution containing 1 part chromic acid and 1 part acetic acid using 10 v for 6 sec. 200×. (a) Alloy 1, the base alloy. (b) Alloy 6, 340 ppm Fe addition. (c) Alloy 7, 700 ppm Ni addition.

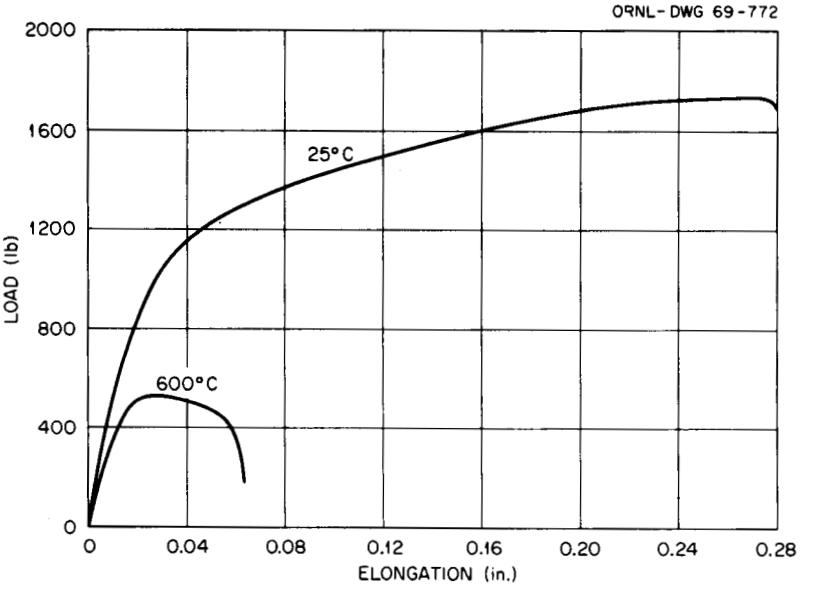

Fig. 5. Typical Load Versus Elongation Curves for Gamma-Quenched U–0.5% Ti Specimens Pulled in Tension at 25 and 600°C.

Fig. 6. Typical Broken Surfaces in Gamma-Quenched U–0.5% Ti Specimens Pulled in Tension at 25 and 600°C.

generally necked down near the point of failure, as shown in Fig. 6. These specimens exhibited cup and cone sections typical of failure by a shear mechanism.

The results of tensile tests at 25 and 600°C are presented in Tables 2 and 3 and shown graphically in Figs. 7 and 8. Comparison of data in these figures shows that the reproducibility between duplicate specimens was considerably less at 600 than at 25°C. Nevertheless, the values obtained for each alloy at 600°C were averaged for comparison with one another and with average values obtained at 25°C.

Table 2. Tensile Properties and Hardness at 25°C of Gamma-Quenched U–0.5% Ti Alloys

Alloy Number	Elongation[a] (%)	Strength, psi		Hardness (DPH)
		0.2% Yield	Tensile	
		$\times 10^3$	$\times 10^3$	
1	24.9	68.5	143.0	342
1	20.7	68.1	146.2	
2	18.6	69.8	142.2	321
2	20.0	69.2	145.1	
3	22.2	68.9	143.5	327
3	20.4	69.8	143.5	
4	25.4	67.2	146.8	354
4	19.6	65.9	150.8	
5	28.1	73.4	159.0	354
5	22.6	72.3	155.2	
6	18.8	83.3	168.5	368
6	22.8	84.6	169.5	
7	18.4	82.1	167.7	347
7	20.2	81.0	167.5	
8	16.3	89.8	160.8	334
8	16.8	81.1	157.2	
9	23.7	72.2	158.5	347
9	26.4	69.8	155.7	

[a]Calculated for 1-in. gage length.

Table 3. Tensile Properties at 600°C of Gamma-Quenched U–0.5% Ti Alloy

Alloy Number	Elongation[a] (%)	Strength, psi	
		0.2% Yield	Tensile
		$\times 10^3$	$\times 10^3$
1	6.3	31.4	35.1
1	5.3	40.5	44.4
2	7.5	34.9	39.9
2	7.4	31.7	35.0
3	8.4	38.1	42.3
3	6.1	36.0	37.3
4	13.3	34.6	38.9
4	5.4	43.5	47.0
5	11.4	34.3	37.4
5	22.0	33.6	36.3
6	2.8	46.3	49.8
6	3.7	42.0	46.5
7	5.0	38.8	43.0
7	b	b	b
8	16.4	29.6	31.8
8	15.7	27.8	30.6
9	4.5	42.1	46.7
9	4.5	42.6	47.8

[a]Calculated for 1-in. gage length.
[b]Duplicate specimen not tested.

The values obtained at 25°C (Fig. 7) show that alloys 4 through 9 had higher tensile strengths than alloy 1, the base alloy. However, three of these alloys – namely 6, 7, and 8 – also had substantially higher yield strengths. Thus, iron (alloy 6) and nickel (alloy 7) appear to be strengthening agents for the base alloy. The yield strengths of the nine alloys ranged between 44 and 54% of the tensile strengths. The average elongations of the nine alloys ranged between a minimum of 16.6% for alloy 8 to a maximum of 25.4% for alloy 5. The hardness of the nine alloys at 25°C shown in Table 2 ranged from a minimum of 321 DPH for alloy 2 to a maximum of 368 DPH for alloy 6. Thus, alloy 6 was both the hardest and strongest alloy tested at 25°C.

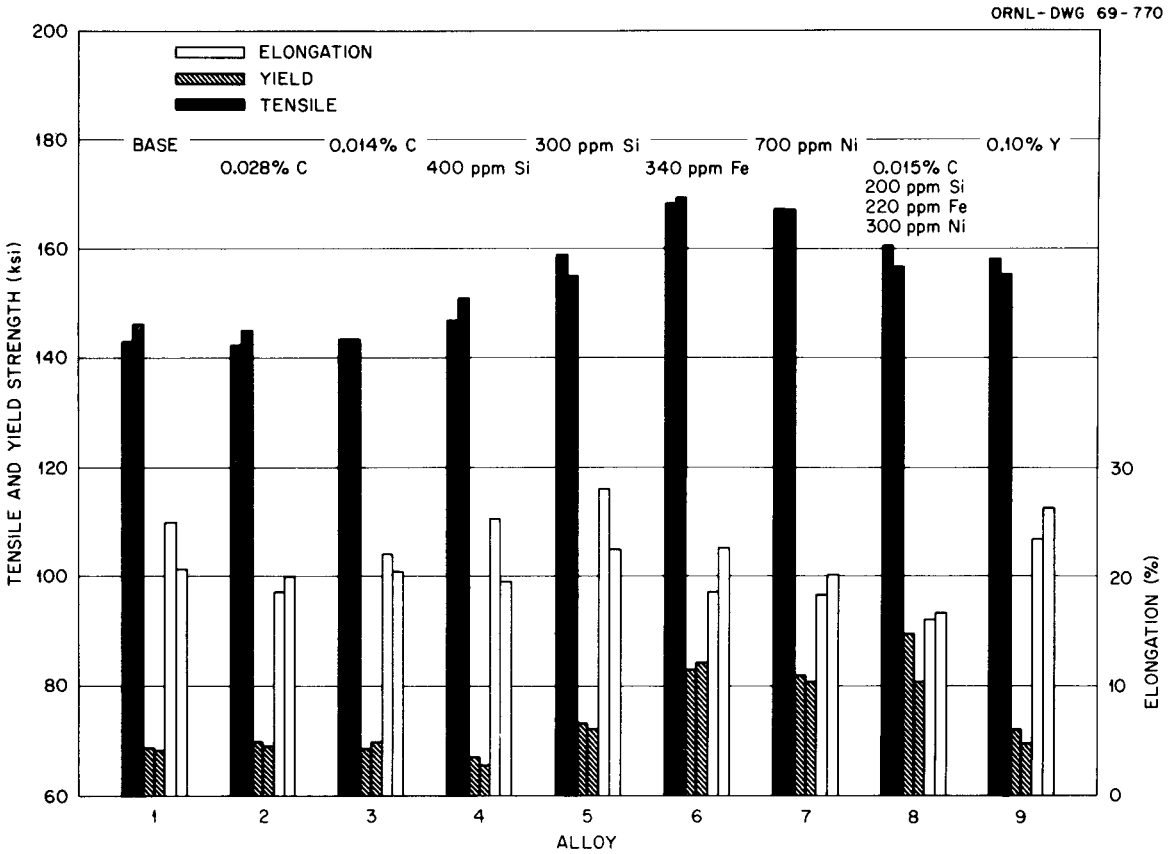

Fig. 7. Uranium–0.5% Ti Alloys Water Quenched from 800 °C then Tested at 25 °C at a Strain Rate of 0.02 min^{-1}.

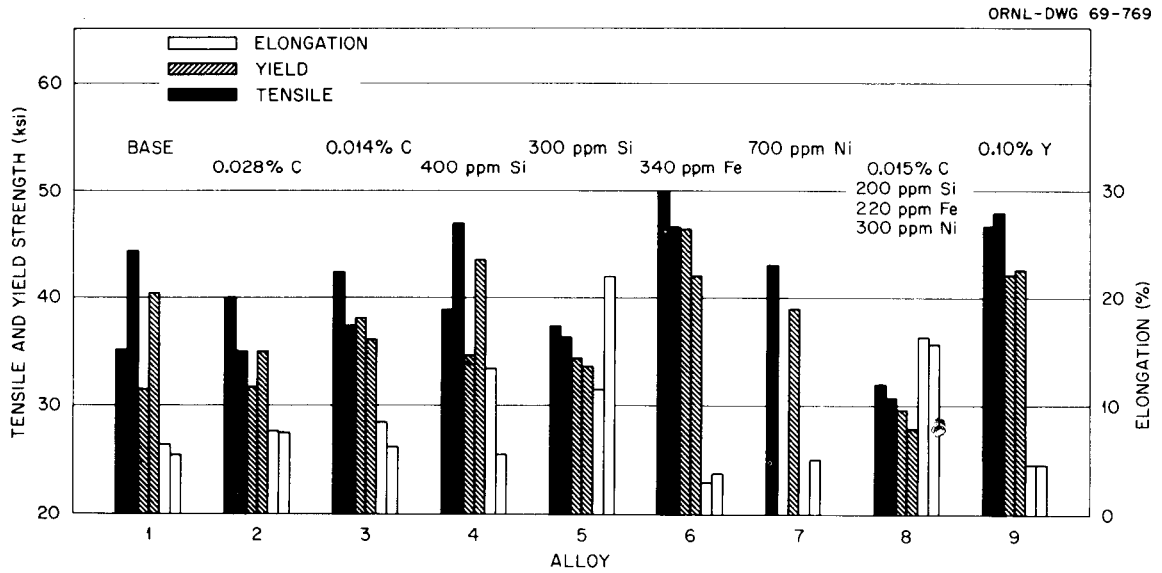

Fig. 8. Uranium–0.5% Ti Alloys Water Quenched from 800 °C then Tested at 600 °C at a Strain Rate of 0.02 min^{-1}.

Figure 8 shows that alloys 4, 6, 7, and 9, containing Si, Fe, Ni, and Y additions, respectively, had higher tensile and yield strengths than the base alloy at 600°C. The strengths in all cases were much lower than obtained at 25°C, and the yield strengths ranged between 89 and 93% of the tensile strengths. Alloys 6, 7, and 9 had elongations similar to that of the base alloy (about 6%) which was considerably lower than was obtained at 25°C. In spite of scatter in the data, the average elongations of alloys 4 and 5 (silicon additions) and alloy 8 were higher than that of the base alloy.

Aged Alloys

The effect of aging on the tensile properties of three of the alloys was also investigated, and the results were compared with the values obtained for beta- and gamma-quenched specimens. Alloy 1 was included to provide a basis for comparison. Alloys 6 and 7, containing iron and nickel, respectively, were studied because high strengths and elongations were obtained at 25°C in the gamma-quenched condition. The results of these tests are presented in Tables 4, 5, and 6. The data are shown graphically in Figs. 9, 10, and 11. Although duplicate specimens were not tested, a similar behavior was shown by the three alloys at each test temperature. The indicated trends, therefore, are probably representative for the test conditions.

Figure 9 shows the tensile and yield strengths and the elongations obtained at 25°C. Specimens aged at 200°C had properties not significantly different from the gamma-quenched specimens. Aging at 400 and 600°C, however, resulted in higher tensile and yield strengths for all three alloys compared to the gamma-quenched condition, but a large decrease in elongation occurred after aging at 600°C. Annealing at 700°C was detrimental to both tensile strength and elongation.

The hardness of the three alloys at 25°C after aging is also shown in Table 4. Although the hardness of alloy 1 was not significantly affected by aging, a correlation of hardness and strength was obtained for alloys 6 and 7. The highest hardness for these alloys occurred for the aging temperatures which produced highest strength – namely 400°C for alloy 6 and 600°C for alloy 7.

Table 4. Tensile Properties and Hardness at 25°C of Aged, Beta-, and Gamma-Quenched U-0.5% Ti Alloys

Alloy[a] Number	Aging[b] or Annealing Temperature (°C)	Elongation[c] (%)	Strength, psi		Reduction in Area (%)	Hardness (DPH)
			0.2% Yield	Tensile		
			$\times 10^3$	$\times 10^3$		
1	200	20.8	66.5	154.3	18.4	347
1	400	18.8	138.5	173.3	19.8	334
1	600	6.3	88.1	165.0	0	354
1	700	8.9	60.7	127.7	9.0	
1	800	22.8	68.3	144.6	23.1	342
6	200	18.8	87.4	183.2	14.4	376
6	400	12.4	129.5	209.5	7.4	437
6	600	5.0	117.8	178.2	3.1	409
6	700	7.4	77.9	142.5	3.3	
6	800	20.8	84.0	169.0	19.2	368
7	200	14.2	76.3	178.0	11.8	354
7	400	8.5	116.6	182.7	7.2	393
7	600	5.3	127.2	188.0	2.5	401
7	700	8.3	77.0	140.0	4.1	
7	800	19.3	81.6	167.6	16.9	347

[a]Alloy 1: base alloy; alloy 6: 340 ppm Fe addition; alloy 7: 700 ppm Ni addition.

[b]1 hr in argon, water quenched.

[c]Elongation in 1 in.

Table 5. Tensile Properties at 400°C of Aged and Beta-Quenched U-0.5% Ti Alloys

Alloy[a] Number	Aging[b] or Annealing Temperature (°C)	Elongation[c] (%)	Strength, psi	
			0.2% Yield	Tensile
			$\times 10^3$	$\times 10^3$
1	200	18.1	50.5	71.5
1	400	13.4	60.6	78.8
1	600	9.3	58.2	75.5
1	700	12.9	40.2	51.9
6	200	16.2	72.5	102.4
6	400	7.5	96.0	121.2
6	600	5.2	75.2	94.6
6	700	14.5	53.3	67.6
7	200	17.0	72.7	104.9
7	400	12.8	88.5	116.0
7	600	6.1	77.9	97.1
7	700	15.4	50.0	63.9

[a]Alloy 1: base alloy; alloy 6: 340 ppm Fe addition; alloy 7: 700 ppm Ni addition.

[b]1 hr in argon, water quenched.

[c]Elongation in 1 in.

Table 6. Tensile Properties at 600°C of Aged, Beta-, and Gamma-Quenched U–0.5% Ti Alloys

Alloy[a] Number	Aging[b] or Annealing Temperature (°C)	Elongation[c] (%)	Strength, psi	
			0.2% Yield	Tensile
			$\times 10^3$	$\times 10^3$
1	200	7.4	31.2	34.7
1	400	10.7	24.6	38.1
1	600	13.0	27.1	31.0
1	700	11.6	23.2	26.1
1	800	5.8	36.0	39.8
6	200	4.7	42.6	47.8
6	400	3.1	42.6	50.2
6	600	3.6	35.0	39.6
6	700	15.2	24.6	26.8
6	800	3.3	44.2	48.2
7	200	2.8	41.5	47.5
7	400	3.2	40.0	46.4
7	600	4.0	35.2	41.3
7	700	17.9	23.3	24.8
7	800	5.0	38.8	43.0

[a]Alloy 1: base alloy; alloy 6: 340 ppm Fe addition; alloy 7: 700 ppm Ni addition.

[b]1 hr in argon, water quenched.

[c]Elongation in 1 in.

The results of testing aged specimens at 400°C are shown in Fig. 10. All alloys had the highest tensile and yield strengths after aging at 400°C and lowest elongations after aging at 600°C. The elongations in specimens annealed at 700°C are almost as high as for specimens aged at 200°C; however, the strengths are lower, so no advantage is gained by annealing at 700°C.

The results for specimens tested at 600°C are shown in Fig. 11. Aging alloys 6 and 7 at 200 and 400°C resulted in properties not significantly different from those of specimens water quenched from 800°C. The strengths of alloys 6 and 7 decreased with increasing aging temperature, then decreased to a minimum for specimens annealed at 700°C; however, a large increase in elongation occurred after annealing at 700°C. Alloy 1,

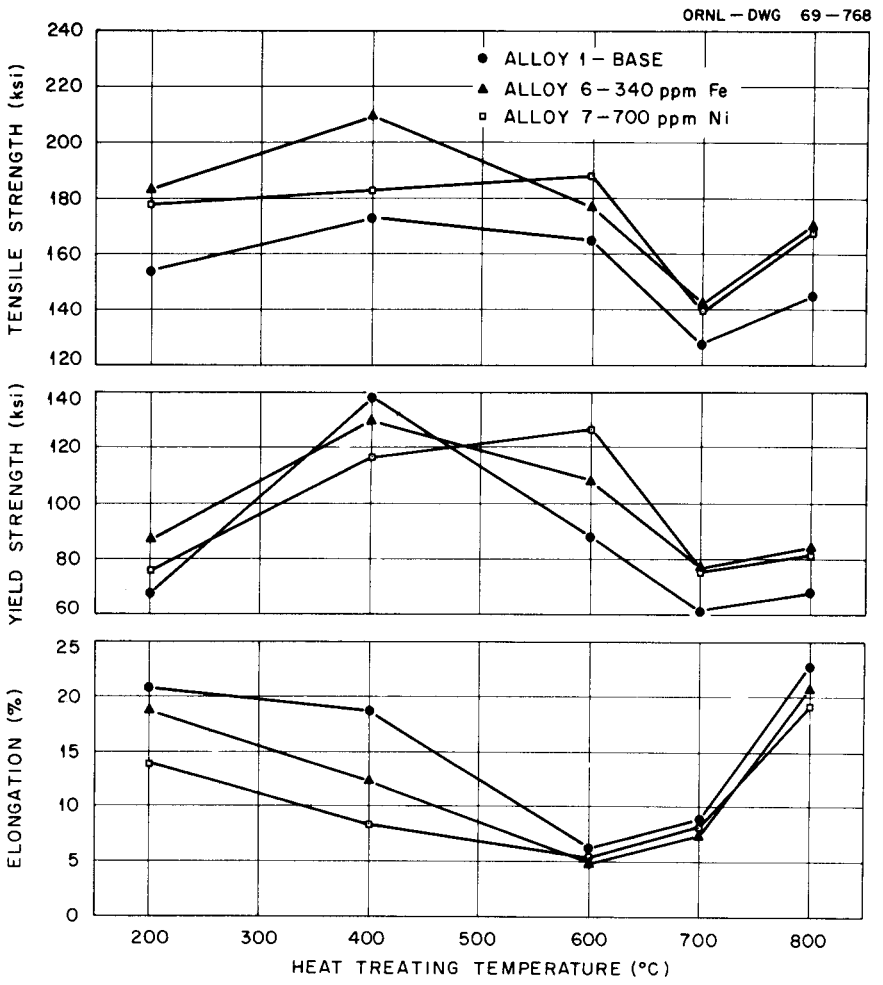

Fig. 9. Tensile Test Results at 25°C for Aged, Beta-, and Gamma-Quenched U–0.5% Ti Alloys.

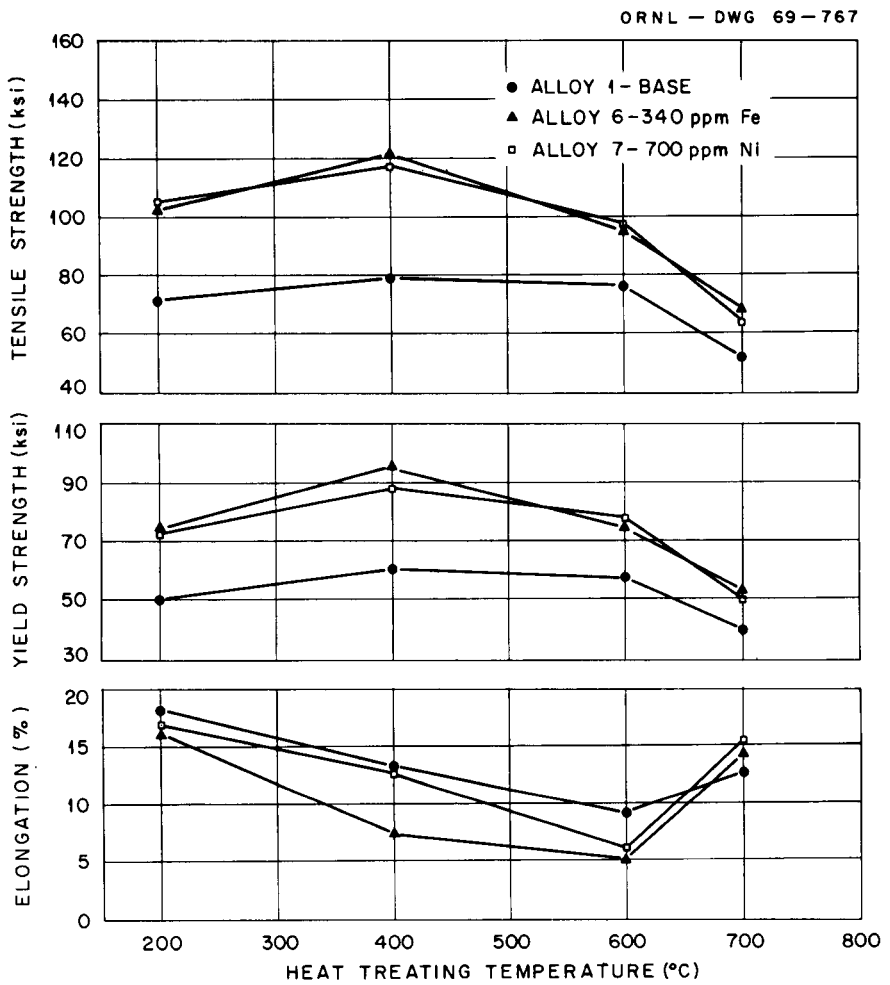

Fig. 10. Tensile Test Results at 400°C for Aged and Beta-Quenched U-0.5% Ti Alloys.

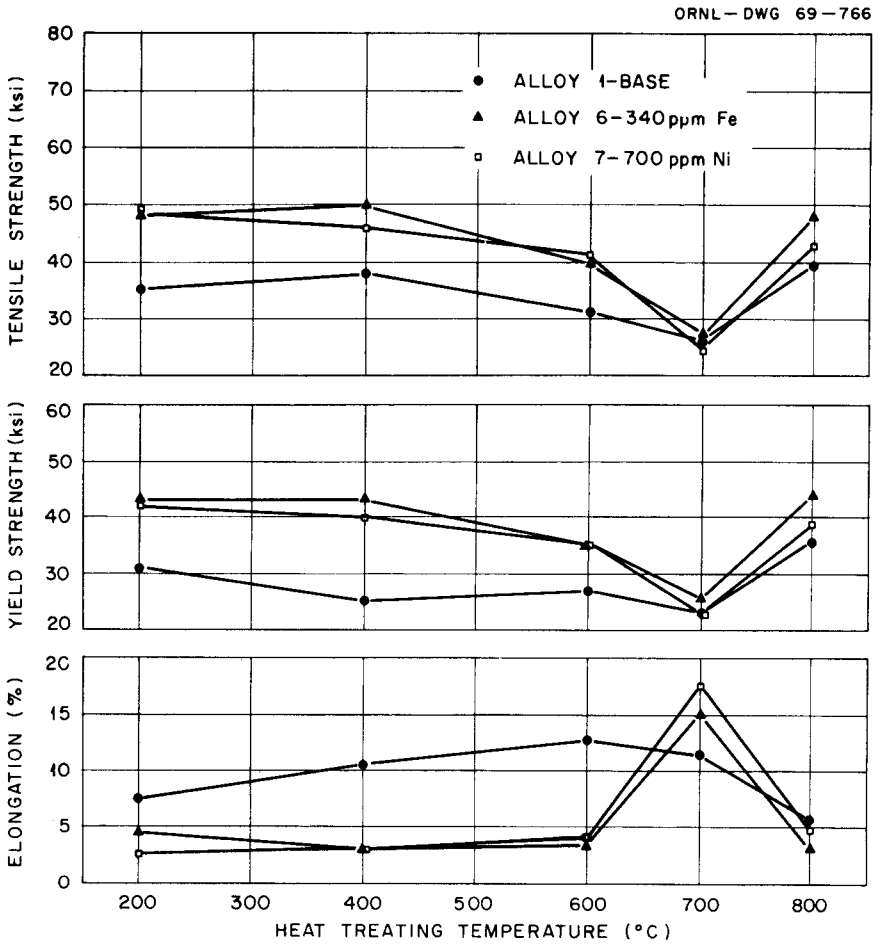

Fig. 11. Tensile Test Results at 600°C for Aged, Beta-, and Gamma-Quenched U-0.5% Ti Alloys.

the base alloy, exhibited a gradual decline in strength for increasing aging temperatures, accompanied by a gradual increase in elongation. The properties of the base alloy at 600°C would be best compromised by aging at 400 or 600°C.

Hydrogen-Contaminated Alloys

Several alloys were charged with hydrogen by annealing at 800°C in a Na_2CO_3-K_2CO_3-$LiCO_3$ salt bath to allow the effect of hydrogen on tensile properties to be examined. The tensile test results are presented in Table 7. The data show that the elongations at 25°C of the base alloy are directly related to the hydrogen content. Specimens B and C which contained 14 ppm H were embrittled by exposure to the salt bath. Due to the curtailed elongation, the tensile strength was decreased by about 14%, while the yield strength apparently was not affected.

Table 7 shows that alloys 3 and 5, containing 0.014% C and 300 ppm Si, respectively, were also embrittled by exposure to the salt bath. Comparison of tensile data for these specimens with vacuum-annealed and gamma-quenched specimens (also included in Table 7) shows that the lower elongations obtained at 25°C are associated with high hydrogen contents. Lower tensile and yield strengths accompanied the decrease in ductility except the yield strength of alloy 5, which increased. Alloys 3 and 5 were also tested at 400°C at a pressure of about 1×10^{-5} torr. The tensile and yield strengths of these alloys at 400°C were comparable to those of the base alloy tested at 400°C in the gamma-quenched and aged (200°C) condition, but the elongations were lower.

Bend Testing of Welded Base Alloy Specimens

The bend test results for welded base alloy specimens are shown in Table 8. The bend angle was measured directly from the specimens and also calculated from the deflection using Eqs. (4) and (5), p. 11, this report. The measured and calculated values agreed within ±0.5 deg. The strains were then calculated from Eq. (3), p. 11, this report.

Table 7. Tensile Properties of Hydrogenated U–0.5% Ti Alloys

Alloy Number	Alloy Addition	Hydrogen Content (ppm)	Test Temperature (°C)	Elongation[a] (%)	Strength, psi	
					0.2% Yield	Tensile
					$\times 10^3$	$\times 10^3$
1A[b]	Base alloy	c	25	26	87.0	168.0
1B[b]	Base alloy	14	25	< 5	88.0	145.0
1C[b]	Base alloy	14	25	< 5	d	143.0
1D[b]	Base alloy	c	25	26	d	168.0
3[e]	0.014% C	0.34	25	21.3	69.4	143.5
3[f]	0.014% C	13	25	4.9	63.0	111.3
3[f]	0.014% C	11	400	14.0	48.3	70.0
5[e]	300 ppm Si	0.11	25	25.4	72.9	157.1
5[f]	300 ppm Si	8	25	7.8	79.6	137.0
5[f]	300 ppm Si	13	400	10.1	51.3	70.1

[a]Calculated for 1-in. gage length.

[b]Differed from alloy 1 used in all other tests in that the specimens were machined from cast slabs of U–0.5% Ti alloy. Specimens A and B were vacuum annealed at 800°C for 24 hr at 1×10^{-5} torr. Specimens A, B, C, and D were then heated to 800°C in high-purity argon for 1 hr and water quenched. Specimens B and C were then heated for 1 hr in a salt bath at 800°C and water quenched. The tensile specimens had a 1/4-in. reduced section and a 1-in. gage length.

[c]Not determined, but probably less than 0.5 ppm.

[d]Not determined.

[e]Vacuum annealed and gamma quenched (see Fig. 7).

[f]Heated for 1 hr in a salt bath at 800°C and water quenched.

Table 8. Results of Bend Testing Electron-Beam-Welded U–0.5% Ti Alloy at 25°C

Specimen Number	Condition[a]	Hydrogen Content (ppm)	Bend Angle (deg)	Strain (%)	Condition at Point of Maximum Strain
1	A	0.1	44	4.6	No cracks
2	A	0.1	69	5.7	Cracked in weld
3	A	0.1	30	4.5	Cracked in weld
5	B	0.1	34	4.3	Broke completely
6	B	0.1	22	3.1	No cracks
6	B	0.1	44	6.2	No cracks
7	C	0.1	17	3.1	No cracks
7	C	0.1	33	5.3	Broke completely
8	C	0.1	26	3.3	Cracked in weld
9	D	2	52	4.8	Cracked in weld
10	D	2	39	4.0	Cracked in weld
11	D	2	44	4.1	Cracked in weld
12	D	2	20	2.7	Cracked in weld

[a]Condition of specimens: A – vacuum annealed, gamma quenched, electron-beam welded; B – condition A plus 1 hr at 200°C, water quenched; C – condition A plus 1 hr at 400°C, water quenched; D – vacuum annealed, gamma quenched, annealed 1 hr at 800°C in 50% Ar–50% H_2, water quenched, electron-beam welded.

Three specimens (1, 2, and 3) were tested in the gamma quenched, then welded conditions. Two similar specimens (5 and 6) were aged at 200°C; two others (7 and 8) were aged at 400°C. Four specimens (9, 10, 11, and 12) were annealed at 800°C in 50% Ar–50% H_2 prior to welding. The limited amount of data do not permit any conclusions relative to the thermal history or hydrogen content of the specimens; however, two facts are clear:

1. Cracking originated in the fusion zone at the point of maximum strain.

2. Cracking occurred at much lower strains than was obtained for the nonwelded base alloy tested in tension at 25°C.

DISCUSSION

Effect of Alloy Elements

In order to evaluate the effect of alloy elements on the mechanical properties of U–0.5% Ti alloys, the properties at 25 and 600°C are compared with those of the base alloy in Table 9. The listed values were calculated from the average values for duplicate specimens. Comparison of the data must be tempered by the uncertainty in Fe, Ni, and Si analyses (1/2 to 2 times the listed value) and by different amounts of nitrogen and oxygen. Although the values in Table 9 represent a limited amount of data, the following generalizations can be made:

1. Neither 0.014 nor 0.028% C had a large effect on properties relative to the base alloy which contained 0.003% C; however, increasing carbon appears to be detrimental to properties, except for elongation at 600°C. Therefore, the carbon content of the alloy should be minimized.

2. Increasing silicon content from 300 to 400 ppm decreases strength and ductility at 25°C and ductility at 600°C, but slightly increases strength at 600°C. The relatively large increase in ductility at 600°C makes silicon a potentially useful alloying element.

3. An iron content of 340 ppm significantly increases strength at both 25 and 600°C, but significantly decreases ductility at 600°C.

4. A nickel content of 700 ppm increases strength at 25°C, less so at 600°C. However, nickel decreases ductility at both temperatures.

5. A yttrium content of 0.10% increases strength at 600°C, less so at 25°C. Yttrium, while increasing ductility at 25°C, decreases ductility at 600°C.

6. The alloy containing additions of C, Fe, Ni, and Si is stronger and less ductile than the base alloy at 25°C, but is weaker and much more ductile at 600°C. The increased ductility at 600°C is probably due to the silicon addition.

7. Iron is a more effective strengthener than Ni, Si, or Y.

Table 9. Effect of Alloy Elements on Mechanical Properties of U–0.5% Ti Alloys Relative to the Base Alloy[a] at 25 and 600°C

Alloy Number	Alloy Element	Difference Relative to the Base Alloy, %					
		Elongation		0.2% Yield Strength		Tensile Strength	
		25°C	600°C	25°C	600°C	25°C	600°C
2	0.028% C	–15	8	2	–5	–1	–6
3	0.014% C	–7	7	2	2	–1	0
4	400 ppm Si	–1	62	–2	9	3	8
5	300 ppm Si	11	188	7	–3	9	–7
6	340 ppm Fe	–9	–43	26	23	17	21
7	700 ppm Ni	–15	–14	20	8	16	8
8	0.015% C, 200 ppm Si, 220 ppm Fe, 300 ppm Ni	–27	178	23	–20	10	–22
9	0.10% Y	10	–22	4	18	9	19

[a]All alloys vacuum annealed and gamma quenched before testing.

The binary phase diagrams for Fe, Ni, and Si with uranium show that these elements are in solution with gamma-uranium at 800°C (ref. 10). However, alpha-uranium has a very low solubility for iron and nickel, and possibly for silicon also. Upon quenching from 800°C these elements might be retained in supersaturated solid solution, as is titanium, and thereby impart the strengthening effect on the alloys that is apparent in Fig. 7. Carbon, on the other hand, has negligible solubility in uranium at any temperature and exists as particles of UC. Due to this low solubility, redistribution of carbon by gamma quenching is not possible, and no strengthening effect similar to that exhibited by Fe, Ni, and Si occurred. Carbon additions of 0.014 and 0.028% to the U–0.5% Ti alloy did not improve the properties compared to the base alloy.

[10]M. Hansen, Constitution of Binary Alloys, 2nd Ed., McGraw-Hill, New York, 1958.

Table 7 shows that hydrogen was detrimental to both strength and ductility of alloys 1, 3, and 5. This result was expected since unalloyed uranium behaves in a similar manner. Hanks, Taub, and Doll[11] showed that uranium annealed for 30 min in a salt bath (35% Li_2CO_3–65% K_2CO_3) or in a hydrogen atmosphere had lower strength and ductility than vacuum-annealed material. The effect was related to about 1 ppm or less hydrogen pickup during annealing. Marsh, Muehlemkamp, and Manning[12] observed a tensile transition from ductile-to-brittle failure in the vicinity of 25°C, which was sensitive to, if not caused by, hydrogen contamination. Davis[13] also observed a tensile transition from ductile to relatively brittle fracture with increasing hydrogen content in the range 0.2 to 0.4 ppm. Uranium which was alpha annealed in hydrogen also had lower strength than vacuum-annealed material.[12,13] Neither Hanks et al. nor Davis observed metallographic evidence for the presence of hydrogen. Polson, Hall, and Guay[14] have also studied hydrogen contamination of uranium as the result of exposure to various carbonate and chloride salts used for annealing. In the present study the tensile tests using specimens exposed to a salt bath showed that hydrogen was definitely detrimental to mechanical properties at 25°C, especially ductility as measured by elongation.

Effect of Heat Treatment

Water quenching U–0.5% Ti alloys from the gamma phase at 800°C results in acicular martensitic alpha-uranium, which is interpreted as uranium that is supersaturated with titanium. This structure occurs when cooling rates are fast enough to suppress transformation of gamma

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[11]G. S. Hanks, J. M. Taub, and D. T. Doll, Effect of Annealing Media on the Mechanical Properties of Uranium, LA-1619 (Aug. 1, 1953).

[12]L. L. Marsh, G. T. Muehlenkamp, and G. K. Manning, Effect of Hydrogen on the Tensile Transition in Uranium, BMI-980 (1955).

[13]W. D. Davis, Solubility, Determination, Diffusion, and Mechanical Effects of Hydrogen in Uranium, KAPL-1548 (Aug. 1, 1956).

[14]C. E. Polson, C. F. Hall, and A. E. Guay, "The Effect of Various Heat Treating Salts on the Hydrogen Content of Uranium Metal," Reactive Metals, Interscience, New York, 1958, pp. 441–460.

uranium to other phases by the usual nucleation and growth type of phase transformation. As previously mentioned Murphy[15] obtained optimum mechanical properties in U–0.53% Ti alloys after water quenching from 800°C as compared with slower cooling rates. Water quenching produced a completely martensitic structure, but slower cooling produced partially martensitic structures. The tensile and yield strengths and elongation of the U–0.44% Ti base alloy reported here agree well with similar data obtained by interpolating Murphy's data as shown below.

	U–0.44% Ti, Gamma Quenched, Tested at 25°C	
	Murphy (by interpolation)	Present Study
Tensile strength, psi	148,000	144,600
Yield strength, psi	63,000	68,300
Elongation, %	20.9	22.8

Aging U–0.5% Ti alloys at temperatures of 200, 400, and 600°C could produce two possible effects. If the composition of the alloy is such that the equilibrium phase is alpha-uranium, then aging would likely produce a softening effect as titanium, held in supersaturated solution due to quenching, is accommodated in the uranium lattice by diffusion. However, if the composition of the alloy is such that two phases, alpha-uranium and U_2Ti, coexist at equilibrium, aging would promote precipitation of U_2Ti and probably strengthen the alloy.

The results obtained for alloys 1, 6, and 7 (base alloy, iron addition, and nickel addition, respectively) tested at 25, 400, and 600°C after various aging or annealing treatments was previously shown in Figs. 9, 10, and 11. For specimens tested at 25°C (Fig. 9) aging resulted in an increase in tensile strength of each alloy. Apparently a temperature of 200°C is too low for much atomic mobility; thus the aging response was small. Aging at 400°C resulted in the largest strengthening effect in alloys 1 and 6. The maximum tensile strengths

[15] D. J. Murphy, "Some Properties of Uranium-Low Titanium Alloys," Trans. Am. Soc. Metals 50, 884–904 (1958).

of these alloys after aging were 20 and 25% higher than the values obtained for water-quenched alloys. Slight to moderate decreases in elongation accompanied the increases in strength. The strength of alloy 7 was about 11% higher after aging at 600°C compared to the strength in the gamma-quenched condition, but the elongation was greatly decreased. These results indicate that the composition of alloys 1, 6, and 7 is such that the equilibrium phases are alpha-uranium and U_2Ti. During aging, titanium in supersaturated solution reacts to form finely dispersed U_2Ti particles and strengthens the alloys. Beta quenching (700°C), however, resulted in lowest strength. Equilibrium is probably attained in 1 hr at 700°C and the equilibrium phases are beta-uranium and U_2Ti. Beta-uranium transforms to alpha-uranium on cooling, which is much less strong than martensitic alpha-uranium.

Aging at 400°C resulted in highest strength for alloys tested at 400°C, and again beta quenching produced the lowest strength (Fig. 10). For specimens tested at 600°C (Fig. 11) aging at 200 and 400°C resulted in small increases in strength relative to the gamma-quenched condition, but aging at 600°C resulted in lower strength. The latter behavior may be due to overaging – that is, the combination of 1 hr at 600°C and the reheat to 600°C for testing may have caused agglomeration of the U_2Ti particles, reducing the effectiveness of the compound as a strengthening agent.

Of course, speculation on the precipitation of U_2Ti during aging would not be necessary if the compound were actually observed in the microstructure. Metallographic examination, however, has not revealed the presence of the compound. Two factors make such an observation difficult: (1) the compound U_2Ti probably exists as an extremely small particle that would be difficult to resolve with a light microscope and (2) other compound particles, such as oxides and carbides, might be mistaken for U_2Ti. A careful metallographic study of specimens aged for long times, and complemented by the electron microscope, would probably resolve the uncertainty concerning U_2Ti.

SUMMARY AND CONCLUSIONS

The effect of certain alloy additions and heat treatments on the tensile properties of the U–0.5% Ti base alloy has been studied. Although the range of alloy additions of C, Fe, Ni, Si, and Y were limited in this study, the effect of specific additions of these elements on the properties of gamma-quenched alloys at 25 and 600°C was studied. Although carbon additions of 0.014 and 0.028% did not substantially affect properties, the effects were mostly detrimental. Alloys containing 300 and 400 ppm Si had much higher elongation at 600°C than any other alloy tested. An iron addition of 340 ppm significantly increased strength at both 25 and 600°C; 700 ppm Ni increased strength at 25°C, less at 600°C; 0.10% Y increased strength at 600°C, less at 25°C. Each of these latter elements decreased elongation at 600°C. Hydrogen contents of 8 to 14 ppm were detrimental to both strength and elongation at 25°C in the base alloy and alloys containing 0.014% C or 300 ppm Si additions.

Aging the base alloy and alloys containing additions of 340 ppm Fe or 700 ppm Ni within the alpha range, then tensile testing at 25 and 400°C, resulted in increased strength accompanied by lower elongation when compared with the gamma-quenched condition. The optimum aging temperature for the alloys was about 400°C. Aging did not improve the properties of these alloys when tested at 600°C.

A few bend tests at 25°C of the base alloy having a full-penetration fusion zone produced by electron-beam welding showed that cracking occurred initially in the weld at much lower strains than occurred prior to failure in tensile tests of the unwelded base alloy. Sufficient tests were not conducted to correlate results with heat treatments and hydrogen content of the specimens.

The results that have been obtained allow the following conclusions concerning the mechanical properties of U–0.5% Ti alloys:

1. Carbon is mostly detrimental to the mechanical properties at 25 and 600°C, and should be minimized.

2. Silicon does not substantially affect the strength at 25 or 600°C, but greatly increases the elongation at 600°C.

3. Moderate strength improvements can be obtained by adding Fe, Ni, or Y to the base alloy. Iron is most effective over the range 25 to 600°C.

4. The strength of the alloy can be further increased by aging. The optimum aging temperature is about 400°C.

5. Hydrogen weakens and embrittles the alloy at 25°C.

6. Gamma-quenched material is stronger and more ductile at 25°C than beta-quenched material; at 600°C gamma-quenched material is still stronger, but less ductile.

7. The ductility of welds is lower than that of gamma-quenched material.

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