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TANTALUM-BASE ALLOYS

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ABSTRACT OF THE DISCLOSURE

Alloys of tantalum with certain ranges of rhenium and tungsten content can be provided with useful reversible precipitating phases by the addition of certain amounts of hafnium, zirconium, or both together, along with carbon and optionally also oxygen and nitrogen. This stabilizes the face-centered-cubic monometal carbide precipitates of tantalum as complex (Ta, Hf)C, (Ta, Zr)C, (Ta, Hf) (C, O, N) or their equivalents, which form very fine and efficiently strengthening precipitates, Ta_2C may also be present. These alloys have high strengths at elevated temperatures, substantial ductility at low temperatures, and can be fusion-welded without losing these characteristics.

CROSS-REFERENCES TO RELATED APPLICATION

This application is a continuation-in-part of application Ser. No. 359,514, filed Apr. 13, 1964, now abandoned.

BACKGROUND OF THE INVENTION

This invention relates to tantalum-base alloys, and more particularly to tantalum-tungsten-rhenium alloys.

Several alloys in the tantalum-tungsten-rhenium system are known. Many such alloys have desirable properties and combinations of properties in their strength at high temperatures, ductility at low temperatures, ductility at room temperature in a welded condition, and otherwise. However, such alloys generally are not readily susceptible to strengthening by precipitation mechanisms or to aging through the solutioning and precipitation of dispersoids and the interplay of various precipitating phases.

SUMMARY OF THE INVENTION

It is an object of the present invention to overcome the above-discussed limitations by providing alloys in the tantalum-tungsten-rhenium system which can be hardened by aging or softened by high temperature annealing and generally affected by precipitation mechanisms to materially change their mechanical properties in a controlled manner.

Another object of the invention is to provide such alloys in which one or more distinct precipitated phases can be provided by various heat treatments, so as to allow the design of optimum thermal and mechanical treatments to meet various requirements. In the cases in which more than one precipitated phase is formed, the phases may preferably be different from each other in thermodynamic stability, morphology, distribution, size, strengthening characteristics, and effects on ductility.

Still another object of the invention is to provide such

alloys in which the precipitating phase is largely controlled by a metal capable of forming carbides and compounds of other interstitial elements and which does not deleteriously affect the strength and ductility properties of the alloy in the condition in which the alloy is to be utilized.

We have discovered that useful alloys can be produced, based on tantalum and containing substantial amounts of tungtsen and rhenium, which can be heat treated to control a precipitate comprising carbides of high melting point metals. Tests on such alloys show quite high strength levels at elevated temperatures along with the retention of substantial ductility at room temperature. Such room temperature ductility has been observed in the stress-relieved and the recrystallized conditions and in welded sheet. Furthermore, we have discovered such alloys which are capable of being heat treated to form two distinct types of carbide precipitates, interactions between said precipitates being valuable and useful in the development of desired properties in the final alloy.

In accordance with the invention, therefore, we provide a tantalum-base alloy containing substantial amounts of tungsten and rhenium giving it valuable ductility and strength, and also containing metal capable of forming carbides and other compounds of interstitial elements, carbon, and, in one form, oxygen and nitrogen, with the amounts and proportions of each so selected as to give useful precipitation-hardenable alloys.

More particularly, we provide a tantalum-base alloy containing about, by weight, 1–5% rhenium, tungsten in an amount such that the total of tungsten and rhenium is from at least 8% to about 16% of the total alloy composition, 0.5–4% halfnium or 0.25–2% zirconium, or 0.25–4% of mixtures of hafnium and zirconium, 0.005–0.1% carbon, balance tantalum. The preferred atomic ratios of hafnium or zirconium, or both together, to carbon are from about 1:1 to about 2.5:1. In another form, the alloys also contain oxygen and nitrogen in combined amount of 0.005–0.1%. Other preferred ranges of tungsten and rhenium content in alloys of the invention are as follows: 1–4% rhenium with 5–10% tungsten, and 2–3% rhenium with 7–9% tungsten. Percentages given in the specification are by weight except where indicated otherwise. Alloys of the invention can be heat treated to precipitate face-centered-cubic monometal carbides of tantalum modified by hafnium, zirconium or both, and optionally also containing oxygen, nitrogen or both. These equivalent precipitates can be expressed as (Ta, Hf)C, (Ta, Zr)C, (Ta, Hf, Zr)C, or the same with oxygen, nitrogen, or both, partially substituted for carbon, e.g., (Ta, Hf) (C, O), (Ta, Hf) (C, N), and (Ta, Hf) (C, O, N).

DESCRIPTION OF THE PREFERRED EMBODIMENTS

The limits and relationships specified for tungsten and rhenium are those that have been found desirable for producing the strength, ductility, and weldability advantages of the invention.

The limits and proportions specified for hafnium, zirconium, carbon, and optionally oxygen and nitrogen are those which produce alloys that are embodiments of the invention, both in terms of structure and variability of structure, and in terms of properties and utility. The limits specified herein for hafnium and zirconium are approximately equal to each other on an atomic basis. When the two metals are used in combination, the com-

bined extreme limits of composition can be approached provided the alloy contains at least 0.5 atomic percent and no more than 4 atomic percent halfnium plus zirconium.

The characteristics and structures of the alloys of the invention which contain more than one precipitate are primarily controlled by the interplay in the tantalum-tungsten-rhenium matrix between precipitates of Ta_2C, Hfo_2, and complex compounds which will generally be preferred to hereinafter as (Hf, Ta) (C, O, N), with carbon being the predominant interstitial element. The tantalum carbide precipitate forms at elevated temperatures in a relatively large particle, well distributed condition. The hafnium-tantalum-carbo-oxy-nitride precipitate and its equivalents, on the other hand, form an extremely fine, well distributed phase of such a particle size as to effectively inhibit dislocation movement at elevated temperatures for strengthening purposes, while not seriously impairing ductility at room temperature. The particle size of the latter precipitate is so small that it often cannot be clearly discerned by optical microscopy at magnifications up to 1000×. It is generally recognized that the optimum particle size for a strengthening dispersoid in an alloy is the minimum particle size obtainable. In the first place, a given amount of dispersoid in a finer particle size will provide numerically more discrete particles, smaller interparticle spacing, and more effective inhibition of dislocation movement than if the particle size is larger. Second, the smaller the particle size of the dispersoids, the less the statistical probability of slip or other deformation of the dispersoid particle itself by the movement of dislocations within the particles, due to the smaller population of dislocations that can exist in the ultra-fine dispersoid particles. Ideally, a dispersoid should be so small that it has a perfect crystalline structure without dislocations.

Apparently, the alloy system which we have discovered, in one embodiment, takes advantage of the principle of stabilizing dispersoids having very small diameters by virtue of complexing. In other words, the (Hf, Ta) (C, O, N,) precipitates in certain alloys of the invention seem to possess an extraordinary stability, both in terms of their formation and lack of dissolution at elevated temperatures, and in terms of their seeming reluctance to grow or agglomerate into larger particles at elevated temperatures. This is apparently caused at least in part by the complex nature of the precipitate, especially as compared to other precipitates in the same system such as Ta_2C and Hfo_2. While the interstitial element purposefully added for the formation of precipitates is carbon, the residual impurities oxygen and nitrogen present in the alloys seem to play a useful role in producing certain alloys of the invention. Although a preferred composition range for oxygen and nitrogen is stated above, small but effective amounts of the two elements present as residual impurities can satisfy the same ends, and alloys without any effective oxygen or nitrogen are within the invention, provided the stabilized monometal carbide can be produced by heat treatment. Since TaC is not stable in the presence of Ta and Ta_2C, the Hf or Zr additions are necessary to stabilize the modified forms of TaC as (Ta, Hf)C or its equivalents. The total effective oxygen and nitrogen content range in alloys of the invention is from zero to 0.1%.

EXAMPLES

During the research program which culminated in the present invention, two alloys were melted, fabricated and tested. These were Alloy 10 intended to have a composition of 7% tungsten, 3% rhenium, 1.5% hafnium, and Alloy 11 having an intended composition of 7% tungsten, 3% rhenium, 1.5% hafnium and 0.1% carbon. Electrodes were fabricated by powder metallurgical techniques for both of these alloys. The electrode for Alloy 10 contained 1.8% hafnium and the electrode for Alloy 11 contained 1.8% hafnium and 0.11% carbon. The tungsten and rhenium contents in the electrodes were those of the intended compositions.

Both electrodes were rectangular, having a width of 1.0 inch, a thickness of 0.75 inch and were produced by welding together three equal length parts to a total length of 41 inches and a weight of about 15 pounds. The alloying constituents were provided as follows: the tantalum was −35 mesh powder containing 0.04% oxygen, 0.023% nitrogen and 0.01% carbon; the tungsten was −325 mesh powder containing 0.0080% oxygen and 0.002% carbon; the rhenium was −200 mesh powder containing less than 0.1500% oxygen and about 0.003% carbon; the hafnium was a mixture of −200 mesh and −100 mesh powder containing respectively, 0.300% oxygen, 0.0028% nitrogen, 0.0049% hydrogen, and 0.300% oxygen and 1.04% hydrogen; the carbon was grade 38 powder procured from the Acheson Colloids Company of Port Huron, Mich. The powder was consolidated into the electrode parts by mechanical pressing followed by vacuum sintering at 3000° F. for one hour. Both alloys were melted with a maximum current of 3200 amps and a current density of 4265 amps per square inch. The average voltage for Alloy 10 was 29 volts, and for Alloy 11 was 30 volts. The pressure over the melt for Alloy 10 rose to 0.8 micron, while that for Alloy 11 rose to 0.5 micron. In both cases the ratio of mold-to-electrode diameters was 3:1. Alloy 10 produced an electrode 3.0 inches in diameter by 3.4 inches long with a few shallow center cracks on the top. Alloy 11 produced an ingot 3.0 inches in diameter by 3.3 inches long in quite satisfactory condition. Straight polarity direct current was used in both cases. The actual analysis of the melted ingot of Alloy 10 was 7.69% tungsten, 1.53% rhenium, 1.39% hafnium, 0.0099% carbon, 0.0039% oxygen, 0.0059% nitrogen and 0.0004% hydrogen. The analysis of the Alloy 11 ingot was 6.19% tungsten, 2.54% rhenium, 1.49% hafnium, 0.0496% carbon, 0.0011% oxygen, 0.073% nitrogen and 0.0027% hydrogen.

Table I shows the effect on hardness of heat treatment of the as-cast alloys. Although the two alloys differed substantially in both nominal and actual carbon contents, they were similar in hardness in the as-cast condition and upon annealing at the lower temperatures. The elevated temperature anneals were performed in a vacuum of 10^{-5} torr, with the heat treatment being terminated by the injection of helium gas into the furnace to slightly below atmospheric pressure, thereby causing a moderate rate of cooling. Annealing above 3500° F. caused little hardness change in the higher carbon content Alloy 11, but increased hardness in Alloy 10, which suggests the possibility of partial precipitation during cooling, indicating that Alloy 10 contained significant amounts of interstitial elements. Some precipitates identified as Ta_2C, were visible in Alloy 11 on metallographic study, particularly after aging at 2500° F. or above 3500° F. Similar precipitates were much less abundant in the microstructures of Alloy 10 than in those of Alloy 11. The overall evidence indicates that, even in the as-cast and annealed condition, significant amounts of strengthening occurred in Alloy 10 caused by a dispersoid of a particles size so small that it was not resolved at the optical magnification used (1000×). Relative to Alloy 10, Alloy 11 contained a much larger amount of a dispersoid identified as Ta_2C. The latter apparently predominated in all the annealed conditions except at 2500° F. where a finer precipitation of (Hf, Ta) (C, O, N) made its appearance. The amount of dispersion was reduced considerably at 3000° F., but became increasingly more abundant again upon annealing at the higher temperatures. The hardness data of Table I suggests that considerable dissolution of the precipitates occurred at and above 3000° F. so that enhanced intragranular dispersion in the high temperature annealed conditions resulted from partial precipitation upon cooling. The coarseness and the

Widmänstatten-type morphology of the dispersion indicated that the precipitation occurred during the initial stages of cooling. Such an abundance of coarse precipitation surpresses subsequent fine precipitation (upon cooling through the lower temperatures). This apparently caused the lack of hardening in the high temperature annealed conditions of Alloy 11.

The two ingots were extruded to sheet bar in cylindrical unalloyed molybdenum jackets. The ingots were ground to approximately 2.6 inches diameter and 2.5 inches long. The molybdenum jackets served both for oxidation protection and to make up the 3 inch diameter required for the extrusion facility utilized. The jacketed billets were soaked at 3750° F. for ten minutes prior to extrusion through a rectangular die of nominally 2.05 inches by 0.67 inch, corresponding to a nominal extrusion ratio of 5 to 1. Both alloys extruded readily to sheet bars approximately 1.75 inches by 0.7 inch by 10 inches.

Specimens from the extruded alloys were annealed at various times at temperatures between 2000° F. and 3000°F. to guide subsequent rolling procedure and to study the aging reactions. The hardness changes of both alloys are included in Table I below. In both the extruded and annealed conditions the hardness of Alloy 10 remained higher than that of Alloy 11. The effect of the one-hour annealing on the as-extruded condition lay essentially in the gradual hardness decrease. VHN refers to Vickers Hardness Number.

TABLE I.—EFFECT OF HEAT TREATMENT ON HARDNESS OF ALLOYS

One-Hour Annealing Temp., ° F.	Hardness, VHN	
	Alloy 10	Alloy 11
As Cast	353	345
2,500	341	329
3,00	345	345
2,250	353	332
3,500	358	341
3,750	391	341
4,000	381	345
As Extruded	353	321
2,000	341	317
2,250	336	317
2,500	332	313
2,750	341	313
3,000	336	306

Both alloys were substantially but not entirely recrystallized in the as-extruded condition. Few if any precipitates were visible in the Alloy 10 microstructure in the as-extruded condition, but the substructure was readily decorated by precipitation upon heating, particularly in the temperature range of 2250° F. to 2500° F. Above the latter temperature, the structural changes were characterized by increasing dissolution and what appears to be complete recrystallization at 3000° F. Alloy 11 contained an abundant dispersion in the as-extruded condition which did not undergo significant changes upon annealing except at 3000° F. when appreciable dissolution appeared to have occurred.

Both alloys were satisfactorily rolled to 0.05 inch thick sheet by the same procedure. The molybdenum jacket was removed from the extruded bars and the bars were rejacketed in titanium to produce a sandwich of tantalum alloy between two sheets of titanium welded around the edges for a total thickness of approximately 0.6 inch. The sandwich was rolled to a reduction in thickness of 60% at 2300° F. at a 20% reduction per pass. The titanium jacket was then removed by pickling and, where necessary, by hand grinding. The material was then stress relieved at 2300° F. for 1.5 hours followed by direct final cross-rolling at 1000° F. to the desired thickness without a jacket. The total reduction in thickness effected at 1000° F. was about 66%, using 10% per pass. The slight scaling experienced at this temperature was readily pickled off in a solution of 60 milliliters H_2O, 20 milliliters Hf, and 20 milliliters of HNO_3. Both the sheet quality and the material yield were considered to be quite satisfactory.

Some material from each of the ingots had previously been rolled to 0.25 inch thick sheet by a different procedure. The procedure consisted of cross-rolling the molybdenum-jacketed bar for 50% at 2200° F., annealing at 2400° F. for one hour, cross-rolling at 2000° F. for 50%, annealing at 2200° F. for one hour, cross-rolling at 1800° F. for 40–50%, removing the molybdenum jacket, and finish-rolling at 1000° F. The term cross-rolling indicates that each rolling pass is made at about 90° to the previous pass. The extruded-on-molybdenum sheathing did not provide fully adequate protection against contamination, and the above-described procedure using a titanium jacket is a more satisfactory way of reproducing the sheet. However, the molybdenum-jacketed material did produce good sheet which was utilized for a study of the effects of annealing after the removal of the contaminated areas of the sheet. Subsequent studies on the effects of annealing on the material rolled in titanium sheathing showed little difference in the reactions of the two types of materials. The effect of one-hour vacuum-annealing on hardness and grain growth is shown in Table II below. In the table, mm. stands for millimeter and RX stands for recrystallized.

TABLE II.—EFFECT OF ONE-HOUR ANNEALING ON HARDNESS AND GRAIN SIZE OF ROLLED ALLOYS

Annealing Temp., ° F.	Alloy 10		Alloy 11	
	Hardness, VHN	Grain Size, mm.	Hardness, VHN	Grain Size, mm.
As Rolled	423		435	
1,000	418		435	
1,600	407		423	
2,000	391		396	
2,250	376		391	
2,500	329		367	
2,750	317		317	
2,800	(RX)321	0.0158		
2,900	321	0.0308		
3,000	329	0.0323	349	
3,100			(RX)349	0.0159
3,200			349	0.0222
3,250	336	0.0625	341	
3,500	329	0.125	341	0.111

The one-hour recrystallization temperature of the rolled Alloy 10 was found to be 2800% F, and that of Alloy 11, 3100° F. The as-rolled condition of Alloy 10 contained moderate amounts of the face-centered-cubic (f.c.c.) dispersed phase (Hf, Ta) (C, O, N) together with a small amount of the monoclinic phase HfO_2. The (Hf, Ta) (C, O, N) phase underwent essentially complete dissolution above 2800° F. which was followed by rapid grain growth of the matrix, particularly above 3250° F. The as-rolled condition of Alloy 11 contained considerably more dispersed phase consisting almost entirely of hexagonal Ta_2C. Using optical microscopy, the dispersion underwent little change with annealing until about 3000° F. when the volume fraction of the dispersion appeared to have decreased as a result of dissolution and agglomeration. It may be noted from Table II that, after recovery and softening at the lower temperatures, the hardnesses of both Alloy 10 and Alloy 11 show a definite increase upon annealing at 3000° F. and above. Dissolution apparently proceeds in Alloy 11 rapidly above the recrystallization temperature of 3100° F. so that, at 3250° F, the dispersed phase was evident mainly along the grain boundaries. Grain growth occurred slowly in Alloy 11 up to 3200° F. but became accelerated upon annealing at 3500° F. as a consequence of complete dissolution of the dispersed phases.

Phase identification studies were performed on Alloy 10 and Alloy 11 by means of X-ray diffraction and emission analyses of residues extracted electrolytically from each alloy in various stages of processing as well as after selected heat treatments. In Alloy 10, the monoclinic HfO_2 and hexagonal Ta_2C were the predominant dispersed phases at temperatures above about 2800° F., as represented by the as-extruded condition and the as-cast, ex-

truded, or rolled conditions annealed at 3000° F. or above. At temperatures below about 2800° F., a new phase appeared in Alloy 10 having f.c.c. structure with a_0 equal to 4.58 angstrom units. This phase was found in the as-rolled condition as well as upon aging the extruded material in the range of 2200°–2750° F. Emission data showed that the phase was rich in hafnium, with the hafnium intensity decreasing when the f.c.c. phase disappeared upon aging at 3000° F. To further characterize the f.c.c. phase in question, residues extracted from two of the aged conditions in which the phase appeared were analyzed for carbon. The carbon contents were found to be between 1 and 2%, much higher than the 0.01% overall carbon content of the alloy, but still considerably less than the approximately 6% stoichiometrically required by the nominal compound (Hf, Ta)C. For this reason, and due to considerations of the stability of the various other f.c.c. phases which could be hypothesized in the present alloy system, and due to the fact that Alloy 11 having a higher carbon content did not contain as much of the monometal carbide (Hf, Ta)C, we have concluded that the f.c.c. phase in question is a complex (Hf, Ta) (C, O, N). The identification implies first, a hafnium-rich compound which contains a substantial amount of tantalum and little tungsten or rhenium (as indicated by the emission data), and second, the presence in the compound of substantial amounts of oxygen and perhaps nitrogen in addition to carbon.

In contrast to Alloy 10, Alloy 11 having a higher carbon content, lower oxygen and about the same nitrogen, has Ta_2C as its major dispersed phase at both high and low temperatures. The lattice parameter values and the emission data indicate little substitution of tungsten or rhenium for tantalum in the precipitate. A small amount of the (Hf, Ta) (C, O, N) phase was found in Alloy 11 below about 2800° F., as it has been detected in the as-rolled condition and upon aging at 2500° F. and 2750° F. The amount of this phase did not increase greatly with prolonged aging at the intermediate temperatures.

Bend tests showed a minimum bend radius at room temperature and below of equivalent to "zero" T for both alloys in the rolled and recrystallized conditions. A radius of T indicates that the specimen bent through a full angle of 115° around a tool having a radius equal to the thickness of the sheet without cracking.

Tensile tests were performed at various temperatures between room temperature and 3500° F. on Alloys 10 and 11, the results of which are presented in Table III below for comparison with the results of similar tests on two alloys which are discussed and claimed in the above-identified copending application. In the table, SR indicates stress relieved material, and RX indicates recrystallized material. Ks.i. means thousands of pounds per square inch. For Alloys 10 and 11, the stress relief was one hour at 2000° F. The recrystallization treatment used on Alloy 10 was one hour at 2900° F.; the recrystallization treatment for Alloy 11 was one hour at 3150° F.; the alloy containing 8.9% tungsten and 2.6% rhenium was recrystallized at 2950° F. for one-half hour; the alloy containing 7.7% tungsten; 2.3% rhenium was recrystallized at 2900° F. for one-half hour. The Alloy 10 and Alloy 11 tests were performed on 0.05 inch thick sheet specimens having a gauge length of 0.5 inch long by ⅛ inch wide with the major surfaces being as-pickled in the rolled condition, and the testing conducted on a Instron apparatus at a nominal strain rate of 0.05 per minute. The tests on the two alloys presented for comparison were performed on specimens machined from sheet which had been rolled to approximately 0.090 inch thick and then ground from each side to about 0.07 inch, having a gauge length of ½ inch. Tests were conducted at a cross-head speed of 0.02 inch per minute to yield and then 0.05 inch per minute to failure. In all cases, testing performed above room temperature was done in a vacuum of about 10^{-5} torr. The material for tensile test specimens of Alloys 10 and 11 was that rolled by the simplified procedure described above using titanium sheathing.

TABLE III.—TENSILE PROPERTIES OF ROLLED ALLOYS
[Stress Relieved at 2,000° F./1 hr. Nominal Strain Rate: 0.05/min.]

Test Temp., ° F.	Condition	Tensile Strength, K.s.i.	Tensile/Density Ratio Inches X10-3	0.2% Yield, Strength, K.s.i.	Elongation, Percent
A. ALLOY 10					
78	SR	156.1	260	149.0	10.2
	RX	128.3	214	123.9	24.5
500	SR	136.2	227	131.3	11.2
	RX	93.5	156	82.3	23.4
1,000	SR	131.3	219	117.2	10.2
	RX	84.5	411	65.3	23.6
1,500	SR	117.8	196	103.1	6.4
	RX	104.7	175	54.0	21.3
2,000	SR	101.3	169	78.2	12.4
	RX	84.8	141	48.3	19.3
2,500	SR	66.7	111	55.9	30.4
	RX	51.4	86	38.6	25.8
3,000	SR	27.2	45	23.6	74.0
	RX	28.8	48	26.1	39.1
3,500	SR	19.2	32	17.5	54.2
	RX	19.6	33	14.7	41.4
B. ALLOY 11					
78	SR	176.8	295	170.2	4.8
	RX	147.8	246	131.3	11.9
500	SR	147.8	246	135.5	8.2
	RX	118.7	198	100.0	14.2
1,000	SR	147.3	246	133.1	7.1
	RX	110.2	184	87.8	10.2
1,500	SR	128.0	213	115.2	4.0
	RX	125.8	210	72.3	13.9
2,000	SR	114.8	191	106.1	7.5
	RX	97.4	162	56.3	13.5
2,500	SR	71.3	119	65.0	18.5
	RX	64.3	107	47.3	14.5
3,000	SR	30.8	51	29.5	35.8
	RX	37.8	63	31.7	23.4
3,500	SR	21.4	36	12.7	71.0
	RX	26.5	44	17.7	33.9
C. Ta, 8.9W, 2.6 Re					
75	RX	88.5	----------	79.6	14.0
3,000	RX	22.6	----------	15.4	60
3,500	RX	14.6	----------	12.3	30
D. Ta, 7.7W, 2.3 Re					
75	RX	107.5	----------	91.0	34
2,500	RX	30.6	----------	21.8	54
3,000	RX	20.0	----------	17.5	49
3,500	RX	14.0	----------	10.9	32

As can be seen by an inspection of Table III, alloys of the present invention compare quite favorably with those described and claimed in the above-identified application, having higher strength at elevated temperature, and still retaining satisfactory ductility at room temperature. Alloy 11, which has a somewhat higher carbon content than does Alloy 10, showed a superiority in strength over Alloy 10. The ductility minimum appears to exist for both Alloys 10 and 11 in the intermediate temperature range of 1500°–2000° F. It may be noted that recrystallization, having deprived both Alloy 10 and Alloy 11 of the work hardening, resulted in lowered strength below 2500° F. Above 2500° F., the properties of the recrystallized conditions are similar to those of the worked condition. Alloy 11, in fact, appeared to be substantially stronger in the recrystallized condition at 3000° F. Also to be noted is the improved ductility in the recrystallized condition over that of the stress-relieved condition at room temperature.

Table III also shows that the room temperature tensile elongation ductility of Alloys 10 and 11 is more than doubled by recrystallization. The elongation of these alloys is substantially lowered at 3000° and 3500° F. by recrystallization as compared to the stress-relieved condition. The increased ductility at room temperature is obviously beneficial, and the tensile data in the table indicates that the reduced elongation at elevated temperatures is generally accompanied by increased strength as would be expected, probably also beneficially increasing creep strength in the recrystallized condition.

Since weldability is a primary criterion of the extent of usefulness of refractory metal alloys, tests were made to determine whether the favorable properties of the re-

crystallized alloys would carry over into the welded condition. Electron beam fusion beads were made on initially stress-relieved speciments of Alloy 10 using 130 kilovolts, 4 milliamps, 30 inches per minute travel speed and focusing the beam about 0.5 inches above the surface for about 75% fusion penetration. The welding did not appear to affect significantly either the bend ductility or the yield strength of the alloy at room temperature.

The foregoing is a description of illustrative embodiments of the invention, and it is applicants' intention in the appended claims to cover all forms which fall within the scope of the invention.

What is claimed is:

1. A tantalum-base alloy consisting essentially of about, by weight: 1–4% rhenium; 5–10% tungsten; said tungsten and rhenium being in amounts such that their total is from at least 8% to about 12% of the total alloy composition; at least one metal selected from the group consisting of 0.5–4% hafnium, 0.25–2% zirconium, and 0.25–4% of mixtures of hafnium and zirconium, said mixtures containing at least 0.5 atomic percent and no more than 4 atomic percent hafnium plus zirconium; 0.005–0.2% carbon; from zero to about 0.1% of at least one element selected from the group consisting of oxygen and nitrogen; balance essentially tantalum; said alloy being capable upon heat treatment of forming precipitates of TaC modified with and stabilized by at least one of hafnium and zirconium and also modified with and stabilized by from zero to a small but effective amount of at least one of oxygen and nitrogen.

2. The alloy of claim **1** in which the modified and stabilized precipitate include (Ta, Hf) (C, O, N).

3. A tantalum-base alloy of claim **1** consisting essentially of about, by weight: 7–9% tungsten; 2–3% rhenium; at least one metal selected from the group consisting of 0.5–4% hafnium, 0.25–2% zirconium, and 0.25–4% of mixtures of hafnium and zirconium, said mixtures containing at least 0.5 atomic percent and no more than 4 atomic percent hafnium plus zirconium; 0.005–0.2% carbon; balance essentially tantalum.

4. The alloy of claim **3** in which the modified and stabilized precipitates includes (Ta, Hf) (C, O, N).

5. A tantalum-base alloy of claim **1** consisting essentially of about, by weight: 7.7% tungsten; 1.5% rhenium; 1.4% hafnium; 0.01% carbon; and at least one element selected from the group consisting of oxygen and nitrogen in total amount of 0.005–0.1%; balance essentially tantalum.

6. A tantalum-base alloy of claim **1** consisting essentially of about, by weight: 6.2% tungsten; 2.5% rhenium; 1.5% hafnium; 0.05% carbon; and at least one element selected from the group consisting of oxygen and nitrogen in total amount of 0.005–0.1%; balance essentially tantalum.

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CHARLES N. LOVELL, *Primary Examiner.*