Sept. 29, 1964 M. B. VORDAHL 3,151,003

MIXED PHASE, ALPHA-BETA TITANIUM BASE ALLOYS

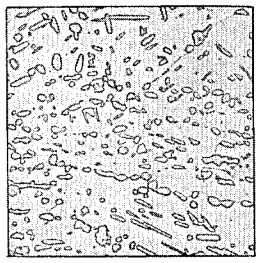
Filed Feb. 13, 1961 5 Sheets-Sheet 1

Fig. 1.



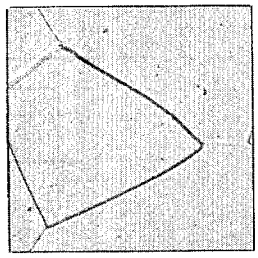
Ti-7% Mn Alloy
2 hrs. 1375°F, water quenched
Ult. Str. 160,000 psi
El. in 1", 19%
Rockwell "A" Hardness, 67.5

Fig. 2.



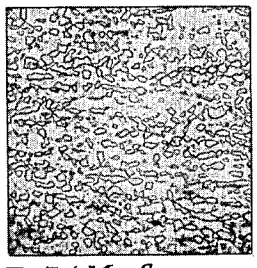
Ti-7% Mn Alloy
2 hrs. 1425°F, water quenched
Ult. Str 173,000 psi
El. in 1", 17%
Hardness R.A., 69.2

Fig. 3.



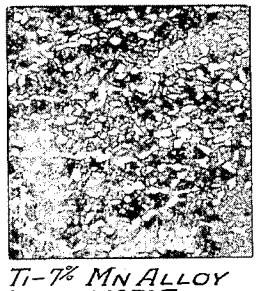
Ti-7% Mn Alloy
2 hrs. 1575°F, water quenched
Tensile properties-Glass Brittle
Hardness R.A., 73.0

Fig. 4.



Ti-7% Mn Alloy
2 hrs. 1425°F, air cooled
and aged 4 hrs. at 600°F.
Tensile properties-Glass Brittle
Hardness R.A. 73.6

Fig. 5.



Ti-7% Mn Alloy
2 hrs. 1425°F, air
cooled and aged 4 hrs at 800°F.
Ult. Str. 171,300 psi
El. in 1", 8%, Hardness R.A. 69.0

INVENTOR.
Milton B. Vordahl.

BY
Ward, Neal, Haselton, Orme & McElhannon
ATTORNEYS.

Sept. 29, 1964 M. B. VORDAHL 3,151,003

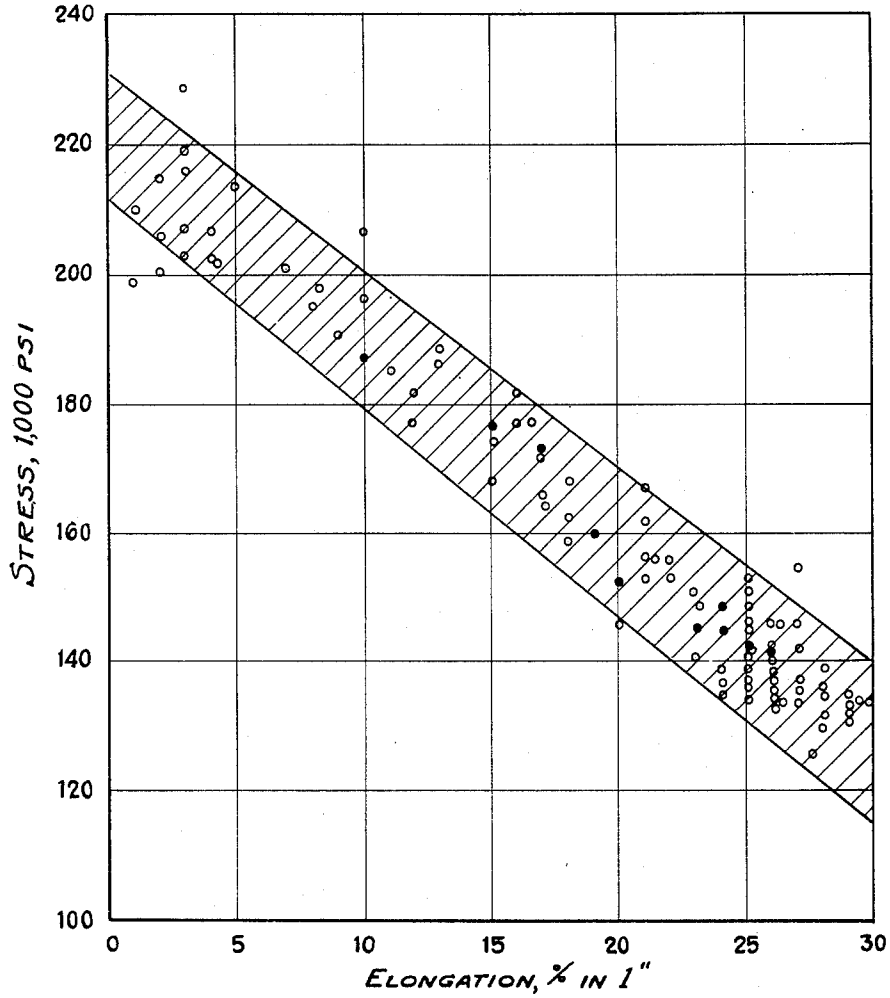
MIXED PHASE, ALPHA-BETA TITANIUM BASE ALLOYS

Filed Feb. 13, 1961 5 Sheets-Sheet 2

Fig. 6.

ULTIMATE STRENGTH VS. ELONGATION FOR Ti-7Mn ALLOY, QUENCHED FROM BELOW BETA TRANSUS AND AGED AT VARIOUS TEMPERATURES AND TIMES

LEGEND FOR PLOTS:-
SOLID DOTS "AS WATER QUENCHED"
CIRCLES "AS WATER QUENCHED AND AGED.



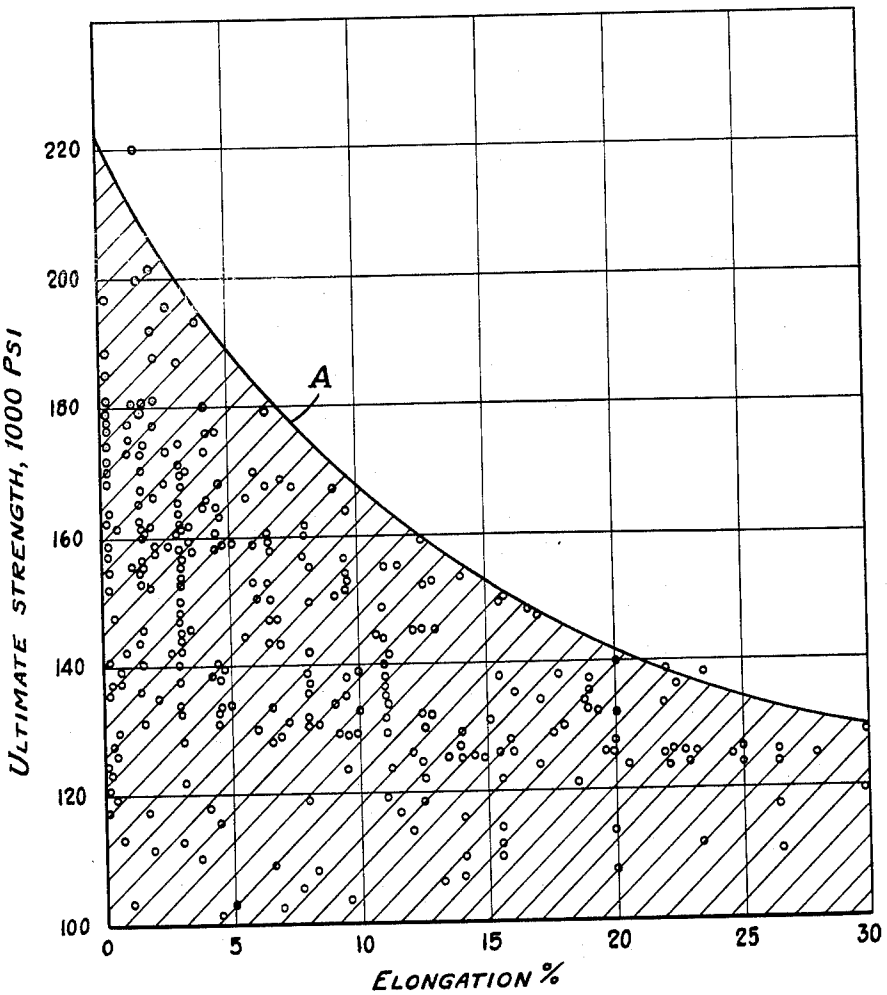
INVENTOR.
MILTON B. VORDAHL.
BY
Ward, Heal, Haselton, Orme & McCharnon
ATTORNEYS.

MIXED PHASE, ALPHA-BETA TITANIUM BASE ALLOYS

Filed Feb. 13, 1961 5 Sheets-Sheet 3

Fig. 7.

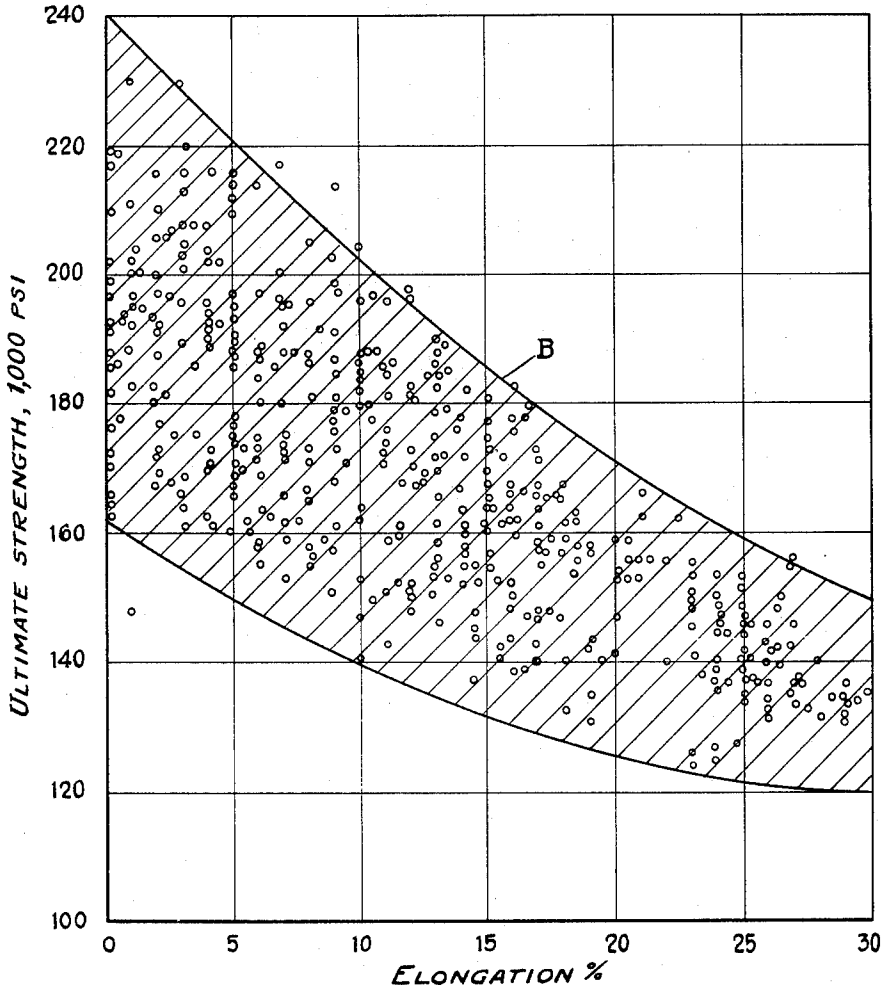
ULTIMATE STRENGTH VS. ELONGATION FOR VARIOUS MIXED PHASE, ALPHA-BETA TITANIUM BASE ALLOYS AS QUENCHED FROM ABOVE BETA TRANSUS TEMPERATURE AND THEREAFTER AGED BELOW BETA TRANSUS



INVENTOR.
MILTON B. VORDAHL.
BY
Ward, Neal, Haselton, Orner & McElhannon
ATTORNEYS.

Fig. 8.

ULTIMATE STRENGTH VS. ELONGATION FOR VARIOUS MIXED PHASE, ALPHA-BETA TITANIUM BASE ALLOYS AS QUENCHED FROM BELOW BETA TRANSUS TEMPERATURE AND THEREAFTER AGED.



INVENTOR.
MILTON B. VORDAHL.
BY
Ward, Neal, Haselton, Ormer & McElhannon
ATTORNEYS.

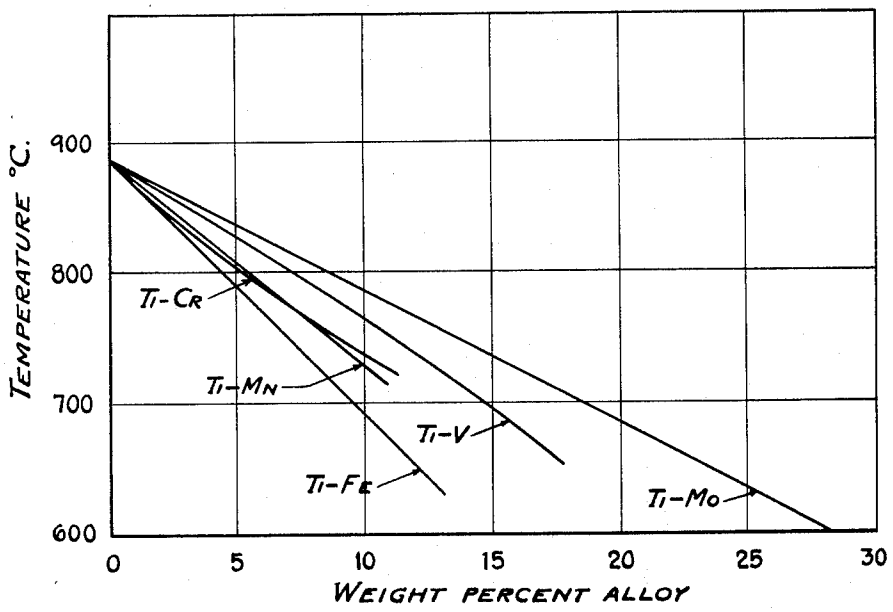
Sept. 29, 1964 M. B. VORDAHL 3,151,003

MIXED PHASE, ALPHA-BETA TITANIUM BASE ALLOYS

Filed Feb. 13, 1961 5 Sheets-Sheet 5

Fig. 9.

BETA TRANSUS TEMPERATURE VS. ALLOY CONTENT FOR VARIOUS TITANIUM BASE BINARY ALLOY SYSTEMS



INVENTOR.
MILTON B. VORDAHL.
BY
Ward, Heal, Haselton, Orme & McElhannon
ATTORNEYS.

United States Patent Office 3,151,003 Patented Sept. 29, 1964

3,151,003
MIXED PHASE, ALPHA-BETA TITANIUM BASE ALLOYS

Milton B. Vordahl, Beaver, Pa., assignor to Crucible Steel Company of America, Pittsburgh, Pa., a corporation of New Jersey

Filed Feb. 13, 1961, Ser. No. 88,943
5 Claims. (Cl. 148—32.5)

This invention pertains to methods of heat treating titanium base alloys having a mixed alpha-beta microstructure for imparting thereto exceptionally high strength and thermal stability, combined with adequate ductility for purposes of fabrication and for use under conditions of service. The invention also pertains to the resulting heat treated alloys.

This application is a continuation-in-part of my copending application Serial No. 435,754, filed June 10, 1954 (now Patent No. 2,974,076, issued March 7, 1961), which is in turn a continuation-in-part of my applications Serial No. 132,327, filed December 10, 1949 (now abandoned); Serial No. 229,143, filed May 13, 1951) now Patent No. 2,950,191, issued August 23, 1960); Serial No. 424,569, filed April 21, 1954 (now Patent No. 2,950,192, issued August 23, 1960); Serial No. 323,155, filed November 28, 1952 (now abandoned), and Serial No. 671,316, filed July 11, 1957 (now Patent No. 2,857,269).

In the first three of my above-mentioned applications referred to, I have described methods for imparting to titanium base alloys having a mixed alpha-beta microstructure, optimum combinations of strength and ductility, and high ductility at any given strength level, said methods consisting essentially in subjecting such alloys to prolonged plastic deformation at temperature within the mixed alpha-beta temperature range, i.e., at temperature below the beta transus temperature at which the microstructure of these alloys converts from mixed alpha-beta to all-beta on heating. As set forth in said applications, the preferred temperature range for carrying out said plastic deformation extends from about 400° C. or 750° F. up to not more than about 50° C. or 100° F. below the beta transus temperature.

As shown in said applications, the effect of plastically deforming such mixed phase alloys within the two phase temperature field, is to impart thereto a microstructure comprising a coherent admixture or interleaved dispersion of small bodies of alpha titanium and of beta titanium. Since the alpha phase of titanium is relatively weak and ductile, while the beta phase is relatively strong but less ductile, the result of this interleaving and dispersion of the two phase is to impart uniformly throughout the alloy a combination of high strength and high ductility.

The field of utility of the processes described in said application is, however, restricted to conditions of fabrication wherein the alloys can be plastically deformed to the extent required for imparting the aforesaid microstructure, such, for example, as rolling into sheet or rod or drawing into wire, etc. Where, however, such alloys must be fabricated into relatively large forgings, they cannot in many instances be worked or plastically deformed sufficiently in the two phase field to secure the full benefits of the processes described in said application. Accordingly, a primary object of the present invention is to provide methods of so heat treating such alloys as to duplicate or surpass the strength and ductility properties impartable thereto by the two phase field, plastic deformation procedures of my earlier applications above mentioned.

Basically the heat treating processes of the present invention applicable to titanium base alloys having a mixed alpha-beta microstructure, consists in quenching such alloys from a temperature substantially below, i.e., at least

50–100° F. below, the beta transus temperature of the alloy, and thereupon subjecting the alloy to an aging treatment and preferably to a prolonged aging heat treatment at a somewhat lower temperature. Broadly stated, the quenching temperature range will extend from about 50° F. below the beta transus temperature down to about 400° F. below the beta transus, a more preferred range extending from about 150 to 300° F. below the beta transus temperature, but in no instance is the quenching carried out at a temperature below about 1100° F. The subsequent aging treatment is carried out within a temperature range broadly stated of about 750–1000° F. for a period of about 1 to 50 hours, a more preferred aging procedure being to age the alloys after quenching, within the temperature range of about 800–900° F. for a period of about 2 to 12 hours.

In order to render the sub-critical temperature quenching treatment of the invention most effective in strengthening the alloy, the heat treatment must consist in two main steps, first obtaining a good distribution or interdispersion of the alpha and beta phases, and secondly quenching in such manner as not to destroy the good alpha to beta distribution obtained by the preceding step.

The first step can, of course, be carried out by the procedure outlined in my above mentioned earlier applications, i.e., by plastic deformation of the alloy at temperature within the two phase field. Particularly good dispersions are thus obtained with the alpha and beta particles elongated and interleaved as results, for example, by rolling or drawing, since this results in greater strength in the direction of the applied stress.

An advantage of securing a good dispersion of the alpha and beta phases by plastic deformation in the two phase field, is that the history of the alloy prior to such plastic deformation is of little, if any, significance. The effect of the plastic deformation in the two phase field is to increase the ductility at a given strength level and this effect is quite independent of the prior history or prior heat treatment of the alloy. Whether the alloy has been hardened by quenching from the beta field or from a lower temperature, or has been previously fully annealed, as by slow or furnace cooling from a beta field temperature, or given any other treatment, extensive working in the two phase field still develops the optimum properties, i.e., the highest ductility at a given strength level.

However, it is not always possible to subject the alloy to such plastic deformation in the two phase field prior to subjecting the same to the quench hardening heat treatment of the present invention. A good example of this is the case of massive forgings requiring working in the beta temperature field, in which case an acceptable alternative for securing a good alpha to beta distribution, is obtained by slow cooling from above the beta transus temperature, i.e., from a temperature at which the alloy assumes an all-beta microstructure. Such slow cooling from above to below the beta transus temperature, precipitating the alpha phase by a process which may be termed solid state dendritic growth, more commonly known as a Widmanstaetten structure, to give a resulting microstructure which is generally lamellar in character.

The conventional practice for quench hardening alloys in general is to quench from above the critical temperature at which the alloy assumes a single phase microstructure. Thus, in the case of hardenable and temperable, martensitic steel, quench hardening is effected by quenching from the austenitizing temperature range, i.e., from above the temperature at which the steel converts from ferrite to austenite on heating. Depending on the rate of quenching and on the carbon content and the contents of other alloying elements present, any desired degree of hardening, within limits, may be thus obtained, and the hardness thereafter reduced to a desired degree

with accompanying increase in ductility by subsequent tempering at a sub-critical temperature.

In the case of the mixed-phase alpha-beta titanium base alloys, however, quenching from above the beta transus temperature results in general in an extremely brittle alloy which cannot thereafter be transformed into a strong and ductile alloy by subsequent heat treatment at a temperature below the beta transus temperature.

In contrast, however, and by employment of the process of the present invention, wherein the alloy has first imparted thereto a good alpha-beta distribution as by plastic deformation in the alpha-beta field, or by slow cooling from the all-beta field, and thereafter quenching from a temperature below the beta transus temperature, there is imparted to the alloy an excellent combination of strength and ductility, all as substantiated by the comparative test data hereinafter presented.

Applicant is aware of no other instance in metallurgy wherein it has been established as optimum to initiate hardening by quenching from within a two phase field rather than from a higher temperature single phase field; or wherein prior to quenching from within a two phase field, the two phases of the alloy are advantageously distributed, by either of the two procedures aforesaid, i.e., by prior plastic deformation in the two phase field, or by slow cooling from a higher temperature single phase field.

The following Table I shows the effect on ultimate strength and tensile elongation of heat treating various mixed alpha-beta titanium base alloys by (*a*) quenching from above the beta transus temperature followed by aging or tempering at a temperature below the beta transus, and (*b*) quenching from below the beta transus temperature followed by tempering or aging at a still lower temperature.

TABLE I

Composition percent (Balance Titanium)	Treatment	Ultimate (p.s.i.)	Elong. percent
(a) 6Mo	1,000° C. 15 min. quenched into water; tempered 800° F. for 10 min.	161,900	0
(a) 6Mo	do	188,800	>2
(b) 6Mo	850° C. 15 min. quenched into water; tempered 800° F. for 10 min.	189,200	4
		190,800	4
(a) 6Mo-0.1N	Quenched into water 10 min. 900° C.; tempered 20 min. 575° C.	191,700	0
		184,000	2
		191,300	2
		192,200	2
(b) 6Mo-0.1N	Quenched into water 10 min. 850° C.; tempered 20 min. 575° C.	191,700	3
		195,000	4
		197,800	2
		198,800	3
(a) 4Mn	Quenched into water 15 min. 1,000° C.; tempered 20 min. 550° C.	174,800	1
		173,900	0
		181,100	0
		177,300	2
(b) 4Mn	Cold rolled; quenched into water for 15 min. 725° C.; tempered for 40 min. 450° C.	175,200	6
		177,300	5
		173,700	5
		173,100	2

It will be seen from the above table that in each instance where the specimen was quenched from above the beta transus temperature and thereafter tempered below the same, the ultimate strength was less and the ductility was also less than for the corresponding alloy quenched from below the beta transus temperature and thereafter tempered.

The data in Table I are based on short time tempering following quenching. Such heat treatment is desirable because of the ease with which it may be conducted. Furthermore, it results in an ideal microstructure in that the distribution of the two phases, i.e., alpha and beta, is ideal as far as strength and ductility are concerned. The microstructure consists of a beta phase matrix which contains a random distribution of globular alpha phase islands. No better distribution with respect to ductility is conceivable. The objection to this short time tempering treatment following tempering, however, is that it is not thermally stable. On long time exposure to moderately elevated temperatures, transformation takes place, due to further conversion of the retained beta into the alpha phase, so that the mechanical properties are consequently altered resulting in a decrease in ductility. Hence the material in this condition is not suitable for applications where elevated temperature service is encountered.

For imparting thermal stability, the mixed phase alloys as quenched from the alpha-beta field are, in accordance with a further aspect of the invention, subjected to an aging heat treatment in the temperature range of about 750–1000° F. and preferably, for optimum results, at about 800–900° F. Although relatively short time aging up to about 1 hour or so substantially increases the thermal stability as compared to the as quenched condition, for imparting maximum thermal stability, prolonged aging or overaging is required. Thus, duration of the aging cycle may vary over a broad range of about 1 to 50 hours, a preferred aging cycle for good thermal stability being from 2 to 12 hours.

The data contained in the following Table II is illustrative of the results obtainable by application of the aforesaid heat treatment according to the invention, to a typical mixed phase, alpha-beta titanium base alloy containing about 6.5% manganese and the balance titanium of commerical purity. An ingot of this alloy as produced by arc melting in a cold mold furnace in an inert or argon atmosphere, was hot rolled at 1300° F. and a series of test specimens prepared therefrom. Groups of these specimens were then heated for 2 hours at 1100° to 1500° F. in an inert atmosphere, and quenched in water to room temperature. Individual specimens were then aged at temperatures of 700–1000° F. for the periods indicated in the table, and tested for room temperature mechanical properties with results as set forth.

TABLE II

Quenched and Aged Tensile Properties on Ti-6.5 Mn *Alloy*

Water Quench Time and Temp.	Aging Temp., ° F.	Aging Time, hrs.	Tensile Properties			
			Ult. (1,000 p.s.i.)	Y.S. (1,000 p.s.i.)	El., percent	R.A., percent
1,100° F., 2 hrs.	700	2	134.3	126.9	26.0	54.0
		16	134.8	123.7	30.0	58.0
		96	137.2	128.0	27.0	62.2
	800	1	136.6	130.6	28.0	59.4
		16	133.5	124.9	29.0	62.8
		96	131.4	121.8	29.0	60.5
	900	2	134.0	122.7	29.0	59.1
		16	134.2	123.2	25.0	56.7
		64	132.4	124.1	26.0	58.5
	1,000	1	133.9	124.8	26.0	58.7
		6	134.4	122.2	27.0	57.4
		24	134.5	123.1	29.0	59.1
1,200° F., 2 hrs.	700	2	140.6	134.3	26.0	53.1
		16	141.7	138.2	26.0	59.8
		96	151.5	144.0	25.0	58.1
	800	1	142.6	137.7	27.0	55.4
		16	154.8	135.5	27.0	48.7
		96	148.7	134.3	25.0	51.5
	900	2	146.2	134.9	25.0	55.4
		16	141.3	131.1	25.0	46.4
		64	137.3	130.5	24.0	53.2
	1,000	1	139.0	133.1	24.0	54.8
		6	130.3	123.2	28.0	56.4
		24	132.2	122.9	29.0	59.5
1,300° F., 2 hrs.	700	2	199.6	186.8	1.0	0.8
		8	207.7	183.3	3.0	2.4
		16	207.7	185.7	3.0	1.6
		24	207.7	184.9	4.0	1.6
		48	206.1	182.7	2.0	2.4
		96	200.4	178.8	2.0	2.4
	800	1	195.5	173.9	8.0	7.9
		4	191.4	161.1	9.0	17.1
		8	186.0	154.6	13.0	29.4
		16	182.5	156.7	16.0	37.9
		48	177.6	160.9	16.0	42.2
		96	168.2	151.3	18.0	47.5
	900	½	174.5	157.8	15.0	39.2
		2	166.0	151.4	17.0	43.3
		8	158.8	145.8	18.0	52.5
		16	155.6	142.4	21.0	48.1
		32	150.7	141.4	23.0	48.3
		64	145.0	136.3	24.0	57.0

TABLE II—*Continued*

Water Quench Time and Temp.	Aging Temp., ° F.	Aging Time, hrs.	Tensile Properties			
			Ult. (1,000 p.s.i.)	Y.S. (1,000 p.s.i.)	El., percent	R.A., percent
1,300° F., 2 hrs.	1,000	¼	167.2	153.2	21.0	49.3
		½	162.7	150.5	21.0	49.5
		1	156.4	143.9	21.0	52.5
		2	156.4	148.1	22.0	53.2
		6	149.3	141.5	23.0	54.7
		24	141.0	135.2	23.0	54.1
1,400° F., 2 hrs.	700	2	Brittle			
		16				
		96				
	800	1	215.9	--------	2.0	1.6
		4	210.6	--------	1.0	.8
		8	214.1	188.3	5.0	4.8
		16	206.9	191.4	10.0	18.9
		48	196.0	177.0	10.0	26.7
		96	188.1	175.4	13.0	29.0
	900	½	201.2	185.1	7.0	13.1
		2	185.7	171.5	11.0	34.8
		8	177.2	163.9	12.0	24.6
		16	168.4	158.5	15.0	32.3
		32	162.1	150.9	18.0	39.1
		64	153.0	141.8	25.0	54.0
	1,000	¼	182.6	171.5	12.0	27.3
		½	177.8	166.1	16.0	48.7
		1	172.1	160.9	17.0	48.1
		2	165.7	153.5	17.0	43.8
		6	153.5	143.4	20.0	43.2
		24	146.6	136.5	20.0	43.4
1,500° F., 2 hrs.	700	2	Brittle			
		16				
		96				
	800	1				
		4				
		8	224.4	--------	--------	--------
		16	197.6	--------	--------	--------
		48	229.7	227.8	3.0	1.6
		96	216.4	211.5	3.0	5.5
	900	½	230.3	227.4	1.0	1.6
		2	220.0	211.1	3.0	5.5
		8	202.5	191.1	4.0	6.4
		16	190.7	183.9	5.0	13.1
		32	180.2	177.9	7.0	13.8
		64	171.1	165.2	7.0	16.9
	1,000	¼	203.7	196.3	3.0	7.9
		½	202.4	194.1	4.0	7.1
		1	193.1	182.9	4.0	9.3
		2	179.2	171.9	6.0	16.1
		6	165.6	159.2	7.0	18.4
		24	151.9	144.6	14.0	31.4

It will be noted that with this alloy, no particular increase in strength, regardless of aging time and temperature, was obtained for specimens quenched below 1300° F. Specimens quenched from 1300° F. and up, however, are shown to have undergone marked strengthening on subsequent aging. The data also show at a 700° F. aging temperature is too low for aging to take place in a reasonable time, and that at upwards of 1000° F. the beta-to-alpha transformation occurs too rapidly for convenient control. The best quenched and aged properties are obtained by quenching in the vicinity of 1400° F., which, for this alloy, is about 150° below the beta transus, and thereafter aging at about 800° to 900° F. Tensile strengths up to 200,000 p.s.i. with 10% elongation are obtainable by this treatment, as shown by the data. This alloy as quenched in water after 2 hours at 1400° F., had the following properties: ultimate strength 177,000 p.s.i., yield strength 171,800 p.s.i., elongation 15%, reduction in area 32.4%.

Various of the Table II quenched and aged conditions were also investigated for notch sensitivity of this alloy. For these tests, standard one-fourth inch notched tensile specimens were employed. The notched ultimate vs. plain ultimate ratios observed on these specimens are given in the following Table III, together with the ultimate strength and reduction in area for the plain and notched specimens. This data shows that the alloy in the quench-overaged conditon is not notch sensitive at stress concentration up to 3.4.

TABLE III

Notched Tensile Properties on Heat Treated Ti-6.5 Mn *Alloy*

Condition	P.s.i.× 1,000 Ult. Strength		Red. in Area Percent		N. Ult./ P. Ult. Ratio
	Plain	Notched	Plain	Notched	
2 hrs. 1,300° F., H_2O Q: +8 hrs. at 800° F.	186.0	238.1	29.4	5.0	1.28
2 hrs. 1,400° F., H_2O Q:					
+16 hrs. at 800° F.	206.9	249.0	18.9	7.4	1.21
+2 hrs. at 900° F.	185.7	251.7	34.8	8.7	1.35
+14 hrs. at 900° F.	168.4	248.4	32.3	11.8	1.49
+¼ hr. at 1,000° F.	182.7	252.4	27.3	5.3	1.38
+½ hr. at 1,000° F.	177.8	248.2	48.7	8.1	1.40
2 hrs. 1,500° F., H_2O Q:					
+8 hrs. at 900° F.	202.5	230.0	6.4	7.8	1.13
+16 hrs. at 900° F.	190.7	235.9	13.1	8.7	1.23
+32 hrs. at 900° F.	180.2	227.2	13.8	8.3	1.26

Table IV below gives the results of a series of quenching and aging tests on a mixed phase, alpha-beta alloy containing about 4% aluminum, 4% manganese and the balance titanium of commercial purity. These tests were conducted in the same manner as those for the above Table II. The beta transus temperature for this alloy is about 1700° F.

TABLE IV

Quenched and Aged Tensile Properties of Ti-4Al-4Mn *Alloy*

Water Quench Time and Temp.	Aging Temp., ° F.	Aging Time, hrs.	Tensile Properties			
			Ult. (1,000 p.s.i.)	Y.S. (1,000 p.s.i.)	El., percent	R.A., percent
1,350° F., 2 hrs.	--------	--------	157.6	148.4	18.0	42.0
	700	2	158.7	148.7	20.0	46.5
		16	162.9	151.7	16.0	40.9
		48	167.7	153.8	12.0	36.2
	800	2	160.9	148.4	17.0	44.0
		16	188.2	157.1	7.0	9.3
		48	189.8	165.8	13.0	27.5
	900	1	163.5	151.4	15.0	42.9
		16	183.3	162.3	13.0	31.4
		24	182.1	164.4	14.0	30.1
	1,000	2	165.6	150.8	13.0	36.2
		16	161.3	148.6	15.0	39.7
		24	161.5	151.4	18.0	38.7
1,450° F., 2 hrs.	--------	--------	150.1	149.1	21.0	42.5
	700	2	178.6	160.3	13.0	14.8
		15	215.9	171.7	4.0	4.1
		24	205.7	168.5	2.0	1.6
		48	167.7	153.8	12.0	36.2
	800	2	213.8	170.5	6.0	10.2
		8	220.1	178.8	5.0	3.9
		24	202.9	170.3	9.0	20.5
		48	215.9	177.4	5.0	10.2
	900	2	205.3	173.2	8.0	24.0
		4	211.8	172.1	5.0	13.2
		24	200.8	176.0	10.0	24.8
	1,000	1	186.3	169.9	11.0	30.2
		2	184.8	162.8	11.0	30.2
		15	167.3	152.5	14.0	38.1
		24	169.5	154.3	15.0	38.4
1,550° F., 2 hrs.	--------	--------	156.9	150.5	17.0	36.8
	700	2	154.4	--------	--------	--------
		4	195.9	--------	--------	--------
		16	187.4	180.5	1.0	--------
		48	199.2	--------	--------	--------
	800	2	191.4	--------	--------	--------
		16	209.8	--------	--------	--------
		48	215.9	--------	--------	--------
	900	1	199.6	--------	--------	--------
		2	202.0	189.9	--------	--------
		8	216.5	191.5	7.0	15.5
		16	209.9	188.4	5.0	12.4
		24	196.3	179.4	7.0	11.0
	1,000	1	215.9	191.6	--------	--------
		2	214.4	195.0	9.0	17.7
		4	203.7	186.3	10.0	19.7
		16	169.6	157.3	12.0	34.9
		24	177.6	164.7	9.0	6.3

TABLE IV—*Continued*

Water Quench Time and Temp.	Aging Temp., ° F.	Aging Time, hrs.	Tensile Properties			
			Ult. (1,000 p.s.i.)	Y.S. (1,000 p.s.i.)	El., percent	R.A., percent
1,650° F., 2 hrs			178.4	90.2	14.0	10.2
	700	2	Brittle			
		16				
		48				
	800	2	61.1			
		8	123.2			
		24	19.59			
		48	194.8			
	900	1	137.6			
		2	204.5			
		4	212.8			
		24	211.8	201.1	3.0	11.5
	1,000	1	204.0			
		4	215.6	197.1	5.0	24.0
		24	177.4	162.0	7.0	13.2
1,750° F., 2 hrs			132.1	111.4	3.0	1.7

It will be seen from Table IV that the quenching and aging response of this Ti-4Al-4Mn alloy is generally similar to that of the Ti-6.5Mn alloy of Table II, except that optimum properties are in this instance obtained by quenching from about 1350–1450° F. and aging at about 900–1000° F. Here again, tensile strengths up to about 200,000 p.s.i. with accompanying elongation of about 10% are obtained.

Referring now to the annexed drawings for a further description of the invention:

FIGS. 1 to 5, inc., are photomicrographs showing the microstructure of a Ti-7Mn, mixed phase, alpha-beta alloy after being subjected to various heat treatments according to the invention as discussed below.

FIG. 6 is a graph comparing the strength versus ductility of the Ti-7Mn alloy as quenched from the alpha-beta field and also after subsequent aging.

FIG. 7 is a plot of tensile strength versus ductility of numerous specific analyses of mixed phase, alpha-beta alloys, as quenched from above the beta transus temperature; while FIG. 8 is a comparative plot of such alloys as quenched from below the beta transus temperature.

FIG. 9 is a plot showing the beta transus temperature versus alloy content for various binary alloys of titanium and a beta stabilizing element.

Referring to the photomicrographs FIGS. 1 to 3, inc., show as quenched microstructures of the Ti-7Mn alloy as water quenched from 1375, 1425 and 1575° F., respectively. As shown by FIG. 9, the beta transus temperature of this alloy is about 780° C. or 1440° F. Thus FIGS. 1 and 2 show the alloy as quenched from the alpha-beta field and FIG. 3 as quenched from the beta field. It will be noted that increasing quenching temperature results in decreasing amounts of alpha titanium (darker etching islands) in the beta titanium matrix (light background). This, as shown in the drawing, is accompanied by increasing tensile strength with decreasing ductility and increasing hardness, as measured on the Rockwell A scale and designated RA. It will further be noted that whereas quenching from the alpha-beta field gives combinations of high strength and ductility, quenching from the beta field rendered the alloy glass brittle.

FIGS. 4 and 5 illustrate the aging response pattern of specimens of the alloy as air cooled in one instance from 1425° F. and aged 4 hours at 600° F., FIG. 4, and as air cooled from 1425° F. in the other instance and aged 4 hours at 800° F., FIG. 5. Note that the FIG. 4 structure is glass brittle and quite hard, RA 73.6. This material has been aged but not overaged. The structure of FIG. 5, on the other hand, has been aged at a higher temperature and is in the overaging condition as indicated by the dark etching background. The dark etching background is the result of a very fine dispersion of transformed alpha titanium in the beta titanium matrix, the white etching islands being the alpha titanium present prior to aging.

As a result of this overaging, it will be noted that the FIG. 5 structure has regained considerable ductility and is much softer than the FIG. 4 structure.

Referring to FIG. 6, the solid dots are plots of ultimate strength versus tensile elongation for quenched specimens of the Ti-7Mn alloy as quenched from various temperatures in the alpha-beta field; while the circles are similar plots of this alloy as quenched from various alpha-beta field temperatures and subsequently aged at varying temperatures and times. The purpose of the plot is to show that the as quenched mechanical properties can be duplicated in the quenched and aged specimens as regards any desired combination of strength and ductility. This is of prime importance because, as shown above, the distribution of the alpha and beta phases in the alloy as quenched from the alpha-beta field, is ideal as regards strength and ductility. Hence the ability to duplicate these as quenched properties in the alloys as quenched and aged, results in the attainment of maximum strength-ductility combinations combined with high thermal stability.

As shown above in Table I for various typical mixed alpha-beta alloys, quenching from the beta field in general renders the alloys weak and brittle; whereas quenching from the alpha-beta field results in general in greatly increased strength with retention of good ductility. That these observations are rules of general application is shown by the test results plotted in FIGS. 7 and 8.

In FIG. 7 is plotted ultimate strengths versus percent tensile elongations for numerous mixed alpha-beta alloys as heat treated in the beta field and quenched. Test results are given for such alloys containing about 2 to 15 atomic percent of one or more of Mo, Mn, Cr, Fe. FIG. 8 is a similar plot of such alloys as heat treated in the alpha-beta field and quenched.

It will be seen from the envelope curves A, FIG. 7, and B—B, FIG. 8, that for any given ductility, the strength level is much higher for these alloys as heat treated in the alpha-beta field than when treated in the beta field. Based on data similar to FIGS. 7 and 8 for the Ti-7Mn and Ti-4Al-4Mn alloys, the advantage of heat treating in the alpha-beta field as compared to heat treating in the beta field prior to aging or overaging is shown by the following:

Strength Level	Attainable Ductility Alpha-betatized, percent	Attainable Ductility Betatized, percent
140,000	27	20
150,000	25	15
160,000	21	11
170,000	17	6
200,000	10	1

It will be seen from the above data that at any given strength level, greater ductility is secured by heat treating the alloy in the alpha-beta field in accordance with the present invention, i.e., by quenching from below the beta transus temperature followed by aging, than is obtainable by quenching from above the beta transus temperature followed by aging.

Although best results are generally obtainable by rapid quenching of the alloys from below the beta transus temperature, as by quenching into water, other quenching media may be employed such as oil quenching and even cooling in air, the latter as shown by some of the examples above discussed.

Referring to FIG. 9 there is shown a plot of the beta transus temperature versus alloy content for titanium binary alloys containing, respectively, the beta promoter additions chromium, vanadium, manganese, iron and molybdenum. Corresponding plots for other beta promoter additions have been published in the literature. The beta transus temperature for any particular alloy may be determined by quenching specimens from pro-

gressively higher temperatures, until a completely martensitic or retained beta microstructure is obtained.

The heat treatment process of the invention is applicable in general to titanium base alloys having a mixed alpha-beta microstructure. Such alloys are generally those containing about 2 to 15 atomic percent of one or more beta promoters. In such alloys the isomorphous beta promoters, molybdenum, vanadium, columbium and tantalum may be present singly or in aggregate up to the upper limit of about 15 atomic percent. The same is also true of the sluggishly eutectoid beta promoters chromium and tungsten. Manganese may be present up to about 10 percent by weight and iron up to about 7 percent by weight, consistent with the retention of adequate ductility in the alloy. The rapidly eutectoid beta promoters cobalt, nickel and copper may be present up to about 5 percent by weight singly or in aggregate. For all such additions, the lower effective limit is about 0.5% and preferably about 1% by weight.

The mixed phase alloys to which the invention is applicable may also contain one or more of the alpha promoters, the more important of which are, aluminum up to about 8 percent by weight, tin up to about 23 percent by weight, and antimony up to about 18 percent by weight, for the ductile alloy range. Here again, the lower effective limit for these additions is about 0.5 percent and preferably about 1 percent by weight singly or in aggregate.

The interstitials carbon, oxygen and nitrogen, which also function as alpha promoters, should not be present in these alloys in amount exceeding about 0.2 percent each by weight.

It was pointed out at the outset that for securing a good alpha-beta distributon, plastic deformation in the alpha-beta field cannot always be resorted to, as for example in the case of massive forgings, and that in such cases the distribution is alternatively obtained by slow cooling from above the beta transus temperature, an appropriate cooling rate being about 250° F. per hour.

Even in the case of products formed of rolled sheet or rod or drawn wire, which as rolled and/or drawn in the alpha-beta temperature range, has a good alpha-beta distribution, it is frequently not possible to use the material in the "as rolled" or drawn conditions, owing to fabrication operations to which this semi-finished product must, in many instances, be subjected, such, for example, as blanking, punching and forming operations. It is often necessary that such rolled or drawn material have considerably lower properties for these operations than are required in the finished articles. The material must, therefore, be converted to the annealed condition for the final finishing operations as by slow cooling from the beta field, in which case it must thereafter be hardened and strengthened, this being accomplished by quenching from below the beta transus temperature and subsequently aging in accordance with the heat treatment of the invention herein.

What is claimed is:

1. A wrought and age hardened titanium base alloy containing about: 2–15 atomic percent of at least one element which stabilizes the beta phase at atmospheric temperatures, up to 23% by weight of at least one element which stabilizes the alpha phase at atmospheric temperatures, and up to 0.2%, each of carbon, oxygen and nitrogen balance titanium, characterized by a microstructure consisting of a beta matrix containing a dispersion of alpha and transformed alpha particles, and by an ultimate strength of at least 160,000 p.s.i. and a tensile elongation of at least 5%.

2. A wrought and age hardened titanium base alloy containing about: 2–15 atomic percent of at least one element which stabilizes the beta phase at atmospheric temperatures, up to 23% by weight of at least one element which stabilizes the alpha phase at atmospheric temperatures, and up to 0.2% each of carbon, oxygen and nitrogen balance titanium, characterized by a microstructure consisting of a beta matrix containing a dispersion of alpha and transformed alpha particles, and by an ultimate strength of at least 180,000 p.s.i. and a tensile elongation of at least 5%.

3. A wrought and age hardened titanium base alloy containing about: 2–15 atomic percent of at least one element which stabilizes the beta phase at atmospheric temperatures, 0.5 to 23% by weight of an alpha stabilizing metal selected from the group consisting of tin and aluminum and mixtures thereof, but not to exceed 8% aluminum, and up to 0.2% each of carbon, oxygen and nitrogen balance titanium, characterized by a microstructure consisting of a beta matrix containing a dispersion of alpha and transformed alpha particles, and by an ultimate strength of at least 180,000 p.s.i. and a tensile elongation of at least 5%.

4. A wrought and age hardened titanium base alloy containing about: 2–15 atomic percent of a beta promoting metal selected from the group consisting of vanadium, molybdenum, columbium, tantalum, chromium, tungsten, manganese and iron, and mixtures thereof, but not to exceed 10% by weight of manganese and 7% by weight of iron, up to 23% by weight of alpha promoting elements, but not to exceed 8% by weight of aluminum and 18% by weight of antimony when present, and up to 0.2% each of carbon, oxygen and nitrogen balance titanium, characterized by a microstructure comprising a fine dispersion of transformed alpha titanium in a beta matrix, together with islands of alpha titanium, and further characterized by an ultimate strength of at least 180,000 p.s.i. and a tensile elongation of at least 5%.

5. A wrought and age hardened titanium base alloy containing about: 2–15 atomic percent of at least one beta promoting element selected from the group consisting of vanadium, molybdenum, columbium, tantalum, chromium, tungsten, manganese and iron, but not to exceed 10% by weight of manganese and 7% by weight of iron, 0.5 to 23% by weight of alpha promoting elements, but not to exceed 8% by weight of aluminum and 18% by weight of antimony when present, and up to 0.2% each of carbon, oxygen and nitrogen balance titanium, characterized by a microstructure comprising a fine dispersion of transformed alpha titanium in a beta matrix, together with islands of alpha titanium, and further characterized by an ultimate strength of at least 180,000 p.s.i. and a tensile elongation of at least 5%.

References Cited in the file of this patent

UNITED STATES PATENTS

2,821,475 Jaffee et al. ---------- Jan. 28, 1958

UNITED STATES PATENT OFFICE

CERTIFICATE OF CORRECTION

Patent No. 3,151,003 September 29, 1964

Milton B. Vordahl

It is hereby certified that error appears in the above numbered patent requiring correction and that the said Letters Patent should read as corrected below.

Column 1, line 22, for "May 13, 1951) now" read -- May 31, 1951 (now --; column 5, line 50, for "at" read -- that --; column 6, TABLE IV, under the column heading "Ult. (1,000 p.s.i.)", line 14 thereof, for "150.1" read -- 156.1 --; column 7, same TABLE IV, same column heading, line 4 thereof, for "19.59" read -- 195.9 --; column 9, line 34, for "distributon" read -- distribution --.

Signed and sealed this 30th day of March 1965.

(SEAL)
Attest:

ERNEST W. SWIDER
Attesting Officer

EDWARD J. BRENNER
Commissioner of Patents