United States Patent Office 2,873,187
Patented Feb. 10, 1959

2,873,187

AUSTENITIC ALLOYS

Wasil W. Dyrkacz, Newtonville, George Aggen, Colonie, and Edward E. Reynolds, Albany, N. Y., assignors to Allegheny Ludlum Steel Corporation, Brackenridge, Pa., a corporation of Pennsylvania

No Drawing. Application December 7, 1956
Serial No. 626,824

7 Claims. (Cl. 75—124)

This invention relates to austenitic alloys and in particular to precipitation hardening austenitic alloys containing iron, nickel and chromium as major elements, said alloys being suitable for use at elevated temperatures of up to 1500° F.

Engineers of today are designing power units for use especially in supersonic aircraft which require the power units to operate at higher and higher temperatures for increased power and efficiency. With the ever increasing temperature of operation of modern day power units new alloys must be developed which are suitable for use at the higher temperatures of operation. In particular, for such applications as turbine wheels and turbine buckets which are designed for use at temperatures of up to 1500° F., the alloy from which such turbine wheels and turbine buckets are formed must be suitable for use at temperatures of up to 1500° F. and must have an optimum combination of high strength, good ductility and resistance to corrosion and oxidation at the elevated temperatures encountered in operation. At the same time, these alloys must be capable of easy fabrication, be relatively cheap and contain a minimum amount of strategic alloying elements.

Heretofore most of the iron and iron-nickel base alloys suitable for use as turbine wheels and buckets have been limited to an operating temperature of not over 1300° F. These alloys have been effective for use in these operations, that is, they have sufficient strength, ductility and corrosion resistance to adequately meet the demands of a power unit operating at temperatures of no greater than 1300° F. Where, however, the temperature of the power unit is designed for operation at temperature in excess of 1300° F., and in particular, at temperatures of about 1500° F., recourse has almost been universally had to nickel base alloys or cobalt base alloys, commonly known as superalloys. When such parts are produced from these superalloys, they are extremely costly, are difficult to fabricate, and utilize a great quantity of strategic alloying elements. Thus, there is an apparent need for an iron-nickel base alloy to replace the superalloys, the iron-nickel base alloy being characterized by using a minimum of strategic alloying elements, and which can be easily fabricated and is relatively inexpensive, yet possesses physical properties comparable to those possessed by the superalloys.

An object of this invention is to provide a precipitation hardening iron-nickel base alloy having a minimum amount of costly and strategic alloying elements, said alloy being suitable for use at temperatures of up to 1500° F.

Another object of this invention is to provide a precipitation hardening austenitic iron-nickel base alloy having high strength, good ductility and adequate corrosion resistance when used at temperatures up to 1500° F.

A more specific object of this invention is to provide a precipitation hardening austenitic iron-nickel base alloy having as essential alloying elements nickel, chromium, aluminum, titanium, molybdenum and tungsten which is capable of being precipitation hardened by heat treatment and is suitable for use at temperatures of up to 1500° F.

Other objects will become apparent when read in conjunction with the following description.

In its broader aspects the alloy of this invention comprises up to about 0.15% maximum carbon, up to 1.5% maximum manganese, up to 1.5% maximum silicon, from about 10.0% to about 20.0% chromium, from about 30.0% to about 55.0% nickel, from about 0.5% to about 4.0% aluminum, from about 1.0 to about 4.5% titanium, from about 1.0% to about 8.0% molybdenum, from about 2.0% to about 15.0% tungsten, from traces to 1.5% vanadium, and the balance iron with not more than about 1.5% of incidental impurities such as nitrogen, phosphorus, cobalt, copper, sulfur and the like. Each of the essential alloying elements performs a specific function within the alloy of this invention.

The carbon is present as an impurity and must be limited to about 0.15% maximum in order to prevent the formation of deleterious carbides, especially titanium carbide, which, if formed, precipitates as an unwanted constituent and adversely affects the strength and ductility of the alloy. Manganese and silicon are used for the purposes of purifying the alloy but do not substantially contribute otherwise to the strength, ductility or corrosion resistance. Chromium, on the other hand, is necessary to provide the alloy with adequate resistance to corrosion and oxidation, especially when used at temperatures of up to 1500° F. At least 10.0% chromium is required in order to afford the alloy with the adequate resistance to corrosion and oxidation whereas chromium contents in excess of 20.0% adversely affect the strength of the alloy. Optimum combination of properties appears to be obtained when the chromium content is maintained within the range between about 13.0% and about 16.0% in that the alloy is afforded sufficient corrosion and oxidation resistance without adversely affecting the strength of the alloy.

Nickel is an essential element in the alloy of this invention and at least 30.0% is required in order to obtain the optimum conditions for precipitation hardening the alloy. The nickel content should be limited to 55.0% since nickel contents in excess of 55.0% do not appear to be of any further benefit in effecting precipitation hardening or enhancing the properties of the alloy, and in addition, contribute to the cost. Optimum combination of properties appears to be obtained where the nickel content is maintained within the range between 42.0% and 48.0%. This optimum range appears to give the optimum balance between strength and ductility without inducing difficulties in fabrication or without excessively increasing the cost of the alloy.

Aluminum and titanium are the elements which cooperate with nickel to form a precipitation hardening alloy. It appears that within the general range of elements set forth hereinafter in Table I, the aluminum and titanium cooperate with the nickel to form a precipitation hardening constituent, possibly Ni_3 (Al, Ti), and each may be partially substituted one for the other. It has been found, however, that in order to obtain the

optimum strength characteristics of this alloy both aluminum and titanium must be present. It has been found that a minimum of 0.50% of aluminum and about 1.0% titanium is necessary for the proper strengthening of the alloy through precipitation hardening although the sum of the aluminum and titanium contents must be not less than 2.5%. Aluminum contents in excess of 4.0% are not beneficial as they do not strengthen the alloy and have a deleterious effect on hot workability. While both aluminum and titanium are essential, the sum of the aluminum and titanium contents should not exceed 5% because of the danger of precipitating deleterious phases of a highly complex nature which adversely affect hot workability and cause reduced elevated temperature ductility. Optimum strength and ductility appears to be imparted to the alloy when the sum of the aluminum and titanium contents is within the range between 3.2% and 4.0%, with at least 0.5% of aluminum and at least 1.0% of titanium present therein.

Where the alloy of this invention is made by the well-known air melting procedure, it is found that vanadium in an amount of up to 1.5% is beneficial in offsetting the embrittling tendency of the titanium content on the resulting alloy. However where the alloy is produced by a vacuum melting procedure, the vanadium may be present only in the form of traces.

Not all of the hardening or strengthening is due to the precipitation hardening of this alloy since both molybdenum and tungsten are used as solid solution strengtheners. In this respect it is found that a molybdenum content of at least 1.0% is needed to show any effective strengthening of the solid solution. Molybdenum contents in excess of 8.0% do not appear to increase the strength to any substantial degree. A tungsten content of at least 2% is needed for solid solution strengthening whereas tungsten contents in excess of 15.0% do not appear to increase the strength to any marked degree. In all cases the tungsten content must be maintained at not less than the molybdenum content and preferably higher in order to obtain optimum solution strengthening of the alloy. High tungsten contents, that is, in excess of 15% increase the cost of the alloy materially. The balance of the alloy is substantially all iron ranging from about 20.0% to about 48.0% with not more than 1.5% of incidental impurities of the type usually found in such alloys, for example, nitrogen, phosphorus, cobalt, copper and sulfur.

Reference may be had to Table I illustrating the general range and the optimum range for the chemical composition of the alloy of this invention, it being noted that where the balance is reported as iron such balance includes the incidental impurities as set forth hereinbefore.

TABLE I

	General Range	Optimum Range
C	0.15 max.	0.12 max.
Mn	1.50 max.	1.50 max.
Si	1.50 max.	1.50 max.
Cr	10.0–20.0	13.0–16.0.
Ni	30.0–55.0	42.0–48.0.
Al	0.50–4.0	0.50–3.50.
Ti	1.0–4.50	1.0–3.5.
Al+Ti	2.50–5.0	3.20–4.0.
Mo	1.0–8.0	1.0–4.0.
W	2.0–15.0	4.0–9.0.
V	Traces—1.5	0.1–0.5.
Fe	Bal.	Bal.

The alloy of this invention may be made in any of the well-known steel mill manners, for example, carbon electrode electric arc furnace melting, induction furnace melting or vacuum melting. Predetermined quantities of scrap and/or hot metal are conveniently melted in an electric arc furnace and thereafter the melt as refined is cast into ingots. The ingots are thereafter hot formed to the desired semi-finished mill product by forging, pressing, rolling, extruding or in any other convenient manner. The alloy in this form can be solution heat treated to be referred to more fully hereinafter and thereafter quenched from the solution heat treatment temperature thereby softening the alloy to permit the alloy to be easily fabricated into the desired component, for example, a turbine wheel or a turbine bucket. Thereafter the fabricated part is subjected to the precipitation hardening heat treatment to be more fully described to impart to the finished article the optimum combination of strength, hardness, ductility, corrosion and oxidation resistance within the alloy. The alloy may also be used as a casting alloy for forming cast articles of predetermined shape, the physical properties being developed or imparted by the precipitation hardening heat treatment to be described hereinafter.

In order to show the effect of the alloying elements and the effect of the heat treatment on the alloys of this invention, reference may be had to Table II which contains the chemical analysis of a number of alloys having a chemical analysis both within and outside of the general range ase set forth hereinbefore in Table I. The alloys of Table II have not been grouped according to the effect of each element but rather according to heat number and the heat numbers have been grouped in Tables III and IV to be referred to hereinafter according to the chemical analysis with respect to the element or elements being varied.

TABLE II

[Chemical composition (wt. percent).]

Heat No.	C	Mn	Si	Cr	Ni	Al	Ti	Al+Ti	Mo	W	V	Fe
D-976	0.04	1.21	0.68	14.7	44.8	1.14	3.0	4.14	1.7	--------	0.3	Bal.
D-977	0.04	1.23	0.68	15.5	44.8	1.1	3.0	4.1	4.9	--------	0.3	Bal.
D-978	0.03	1.30	0.78	15.1	44.9	1.0	2.8	3.8	7.4	--------	0.3	Bal.
D-979	0.03	1.36	0.62	15.0	45.3	1.1	2.7	3.8	3.9	4.04	0.3	Bal.
D-980	0.03	1.34	0.66	14.8	44.3	1.0	3.0	4.0	--------	7.53	0.3	Bal.
D-981	0.03	1.38	0.70	14.7	44.3	1.1	2.8	3.9	--------	11.04	0.3	Bal.
H-151	0.03	1.17	0.59	14.3	44.7	0.16	3.2	3.36	1.5	--------	0.3	Bal.
H-152	0.06	1.34	0.62	14.1	45.5	0.22	2.6	2.82	3.9	3.5	0.3	Bal.
H-153	0.09	1.36	0.69	14.1	45.3	1.07	2.8	3.87	3.9	3.7	0.4	Bal.
H-233	0.12	1.56	0.12	15.0	43.1	0.9	2.9	3.8	3.6	3.6	0.3	Bal.
H-400	0.04	0.25	0.56	14.7	44.3	0.95	2.96	3.91	3.6	2.95	0.4	Bal.
H-402	0.06	0.25	0.68	13.9	44.1	1.36	2.29	3.65	3.8	3.5	0.3	Bal.
H-403	0.03	0.25	0.70	14.4	45.1	1.8	1.92	3.72	4.04	3.51	0.3	Bal.
H-404	0.03	0.23	0.64	14.0	44.0	2.1	1.4	3.5	3.74	3.17	0.3	Bal.
H-405	0.05	0.28	0.52	12.5	46.3	2.68	1.34	4.02	3.9	3.81	0.3	Bal.
H-513	0.03	0.14	0.24	15.5	45.0	3.12	0.75	3.87	3.93	3.58	0.3	Bal.
H-520	0.02	0.23	0.32	15.2	45.0	0.8	3.1	3.9	3.93	3.58	0.4	Bal.
V-039A	0.12	--------	0.08	15.3	44.7	0.7	2.7	3.4	0.11	7.9	Trace	Bal.
V-039B	0.12	--------	0.08	14.9	45.3	0.6	2.7	3.3	1.87	7.88	Trace	Bal.
V-039C	0.10	--------	--------	15.2	44.8	0.6	2.7	3.3	3.4	7.85	Trace	Bal.
4X-374	0.09	0.32	0.24	14.2	42.6	0.9	3.0	3.9	3.82	3.65	0.2	Bal.
62918	0.05	1.35	0.95	15.5	26.1	0.2	1.95	2.15	1.25	--------	0.3	Bal.
M-252	0.17	--------	--------	19.0	Bal.	0.5	2.25	[1]10.0	10.0	--------	Trace	3.0

[1] Co.

Reference may be had to Table III illustrating the effect of some of the alloying elements on the stress rupture properties on the alloy of this invention. It is to be noted that in certain engineering applications such as, for example, where the alloy is to be subjected to high stresses at temperatures of up to 1500° F., the stress rupture properties afford a closer criterion of evaluating the properties of the alloy for that purpose.

TABLE III

Stress rupture properties

[Temperature—1500° F. Stress—25,000 p. s. i.]

A. INFLUENCE OF MANGANESE AND SILICON

Heat No.	Mn	Si	Rupture Life (Hrs.)	Elongation (percent)	Reduction of Area (percent)
D-979	1.36	0.62	235	5.0	5.1
H-153	1.36	0.69	237	4.3	7.8
H-233	1.56	0.12	240	10.8	15.3
4X-374	0.32	0.24	308	4.4	7.0
H-520	0.23	0.32	200	3.4	5.5

B. INFLUENCE OF ALUMINUM

	Al				
H-152	0.22		86	13.9	18.0
H-153	1.07		237	4.3	7.8
H-151	0.16		26	8.0	18.4
D-976	1.14		103	6.9	14.8

C. INFLUENCE OF TITANIUM AND ALUMINUM

	Ti	Al			
H-400	2.96	0.95	395	9.4	15.3
H-402	2.29	1.36	388	9.8	17.4
H-403	1.92	1.80	342	12.8	32.4
H-405	1.34	2.68	283	19.1	43.9
H-513	0.75	3.12	57	2.0	8.2

D. INFLUENCE OF MOLYBDENUM AND TUNGSTEN

	Mo	W			
D-796	1.7	0	103	6.9	14.8
D-979	3.95	4.04	235	5.0	5.1
D-981	0	11.04	162	7.8	16.7
V-039A	.11	7.9	230	10.7	16.4
V-039B	1.87	7.88	390	18.8	19.7
V-039C	3.4	7.85	247	20.1	37.8

[Temperature—1500° F. Stress—18,000 p. s. i.]

D-976	1.7	0	386	4.5	4.7
D-977	4.96	0	562	4.5	5.1
D-978	7.40	0	236	2.7	8.0
D-979	3.95	4.04	978	5.7	10.4

The influence of manganese and silicon on the alloy of this invention is clearly illustrated by reference to subsection A of Table III. Comparing the test results recorded for Heats D–979, H–153 and H–233 it is seen that by increasing the manganese content from 1.36% to 1.56% and decreasing the silicon content from 0.62% to 0.12%, the balance of the alloying elements remaining substantially unchanged, there is no substantial effect on the rupture life of these alloys. Substantially the same results are obtained when comparing Heats 4X–374 and H–520 with the foregoing three heats, it being noted that the latter two contain about 0.27% manganese and about 0.27% silicon whereas the three foregoing heats contain about 1.4% to 1.5% manganese and about 0.6% silicon. It is thus clearly seen that manganese and silicon exert little influence on the stress rupture properties of these alloys. While it is desirable to maintain the manganese and silicon contents as low as possible consistent with good deoxidation and purification of the alloy, these elements which are usually introduced unintentionally through the use of master alloys and scrap, do not affect the physical properties of the steel.

Referring now to subsection B of Table III, the test results recorded therein clearly illustrate the influence of aluminum on the alloy of this invention. Thus comparing Heat No. H–152 with Heat No. H–153 it is seen that increasing the aluminum content from 0.22% to 1.07%, the balance of the alloying elements remaining substantially the same, has produced a large increase in the rupture life of from 86 hours to 237 hours. It is, therefore, apparent that the aluminum content has a great influence on this alloy. It is for this reason also that an aluminum content of not less than 0.5% must be maintained in this alloy. Similar results are obtained when comparing Heat H–151 with Heat D–976. Thus for the latter two heats which have substantially the same chemical composition of the foregoing heats with the exception of the tungsten content it is clearly illustrated that increasing the aluminum content of from 0.16% as in Heat H–151 to 1.14% in Heat D–976 has produced a corresponding increase in the rupture life of from 26 hours to 103 hours, respectively. While the rupture life in the latter two alloys is substantially lower than the rupture life of Heats H–152 and H–153 such difference can be explained on the basis that Heats H–152 and H–153 have about 3.6% tungsten and 3.9% molybdenum where Heats H–151 and D–976 have about 1.6% molybdenum and are completely devoid of tungsten. It is thus apparent that both within and outside of the general range of alloying elements set forth in Table I, increasing the aluminum content has a definite beneficial effect for increasing the rupture life of these alloys. However, the beneficial effect of aluminum is only apparent when the aluminum content is at least 0.50%.

Referring now to subsection C of Table III, the test results recorded therein clearly illustrate the effect of both the aluminum content and the titanium content within the alloy of this invention. It is to be noted that in Heats H–400, H–402, H–403, H–405 and H–513 that the sum of the aluminum and titanium contents is within the range between about 3.6% and about 4.0%. In these same heats, the aluminum content varies between 0.95% and 3.12% whereas the titanium content varies between 2.96% and 0.75%. It clearly appears from the test results recorded in subsection C of Table III that increasing the aluminum content and decreasing the titanium content has substantially no effect so long as at least 0.5% aluminum and 1.0% titanium is maintained within the alloy and the sum of the aluminum and titanium contents is within the range between 2.5% and 5.0%. Thus Heats H–400, H–402 and H–403 have a rupture life of 395 hours, 388 hours and 342 hours, respectively, as the aluminum content is increased from 0.95% to 1.80% and the titanium content is decreased from 2.96% to 1.92%. However, as clearly illustrated by the rupture life recorded for Heat H–513, a minimum titanium content of about 1.0% is needed for the optimum combination of physical properties in the alloy. Thus by increasing the aluminum content from 2.68% to 3.12% and decreasing the titanium content from 1.34% to 0.75% as in Heats H–405 and H–513, respectively, a corresponding decrease in the rupture life is noted of from 283 hours to 57 hours, respectively. It, therefore, becomes apparent that at least 1.0% titanium is necessary for obtaining the desired physical properties in the alloy. While it appears that titanium and aluminum are partially interchangeable, it is also evident from Table III that both aluminum and titanium must be present, the aluminum content being at least 0.5% and the titanuim content being at least 1.0%, the sum of the aluminum and titanium contents being at least 2.5%. Increasing the sum of the aluminum and titanium contents to more than 5.0% does not appear to increase the properties over the properties obtained when the sum of the aluminum and titanium contents is within the range between 2.5% and 5.0%. In addition, there is a distinct possibility of precipitating highly complex compounds of both aluminum and titanium which compounds, if precipitated, detrimentally affect the physical properties of the alloys and the hot workability.

Referring now to subsection D of Table III and the test results recorded therein, it is clearly seen that both molybdenum and tungsten substantially contribute to the properties of the alloy. By comparing Heat No. D–976 with Heat No. D–979, it is seen that increasing the molybdenum content from 1.7% to 3.9% and adding 4% of tungsten to the latter alloy is sufficient for increasing the rupture life from 103 hours to 235 hours, the other elements remaining substantially the same. However, if the tungsten content is increased to 11% and the molybdenum excluded, it is seen that the rupture life of the alloy has been decreased to 162 hours as illustrated by Heat D–981. The effect of the molybdenum is clearly illustrated by reference to Heats V–039A, V–039B and V–039C. In these three heats, having a tungsten content of about 7.8%, it was found that an increase in the molybdenum content of from 0.11% to 1.87% and 3.4% produced corresponding rupture life in these alloys of 230 hours, 390 hours and 247 hours, when they are tested at a temperature of 1500° F. and at a stress of 25,000 p. s. i. It thus clearly appears that increasing the molybdenum content up to about 2% with about 7.8% tungsten present therein is effective for obtaining a substantial increase in the rupture life. These alloys also possessed excellent short time elevated temperature tensile properties.

When Heats D–976, D–977, D–978 and D–979 are tested at 1500° F., and a stress of 18,000 p. s. i., it is seen that increasing the molybdenum content from 1.7% to 4.9% produces a great increase in the rupture life of from 386 hours to 562 hours as shown by Heats D–976 and D–977 whereas a further increase in the molybdenum content to about 7.4% produces a corresponding decrease in the rupture life to 236 hours as illustrated for Heat D–978. Optimum results appear to be obtained when the molybdenum and tungsten are present in equal amounts of about 4% each. Thus Heat D–979 having about 4% of each of molybdenum and tungsten has a rupture life of about 978 hours when tested at 1500° F. and 18,000 p. s. i. From the foregoing, it is apparent that both molybdenum and tungsten are necessary for obtaining optimum physical properties in the alloy. At least 1.0% of molybdenum and 2.0% tungsten are necessary and optimum results appear to be obtained where the molybdenum content is in the range between 2% and 4% with the corresponding tungsten content between 4% and 8%.

As was stated hereinbefore, the alloy of this invention is a precipitation hardening alloy which is suitable for use at temperatures of up to 1500° F. However, it becomes necessary to design a heat treatment which is capable of imparting optimum properties to the alloy. It has been found that two separate heat treatments may be employed to impart the desired physical properties to the alloy of this invention depending upon the temperature at which the article is intended to be used. It is preferred to heat treat the alloy of this invention by a two stage precipitation hardening heat treatment where the alloy is intended to be used at a temperature of about 1200° F. On the other hand, if the intended temperature of operation is at 1500° F., high rupture strength can be obtained through the use of a single stage heat treatment.

More particularly, the two stage precipitation hardening heat treatment consists of solution heat treating the alloy at a temperature in the range between 2000° F. and 2100° F. for one half to two hours followed by rapid quenching in water, oil or air. The solution heat treatment is necessary to place all of the precipitation hardening elements in solution within the matrix of the alloy in order to obtain the full effect of these elements thereby obtaining the optimum balance of physical properties. The alloy in the solution heat treated form is soft and ductile and can be easily fabricated into the finished product. Thereafter, the alloy in the form of the finished product is subjected to the initial aging step which consists of an aging heat treatment at a temperature in the range from 1500° F. to 1600° F. for 18 to 30 hours and preferably about 24 hours and thereafter air cooling. This is followed by a second aging heat treatment at a temperature of 1350° F. to 1450° F. for 12 to 20 hours and preferably about 16 hours and air cooling. This two stage precipitation hardening heat treatment is designed to obtain high ductilities in the alloys of this invention when the alloy in the form of the finished wrought or cast product is used at a temperature of up to about 1200° F. On the other hand, if the alloy is to be used at a temperature of about 1500° F., it has been found that a single stage precipitation hardening heat treatment is effective for developing the optimum stress rupture properties. The single stage heat treatment consists of a solution heat treatment at a temperature in the range between 2000° F. and 2100 F. for about one half to two hours followed by a rapid quench in water, oil or air. Thus the same solution heat treatment temperatures and times are used in each instance. Thereafter, the alloy in the form of the finished wrought or cast product is precipitation hardened by heating the alloy at a temperature in the range between 1275° F. and 1375° F. for about 14 to 24 hours and preferably about 16 hours and air cooling.

In order to more clearly demonstrate the effect of the heat treatment on the stress rupture properties, reference may be had to Table IV illustrating the stress rupture properties for wrought alloys treated by each of the heat treatments and thereafter tested under the conditions of temperature and stress as indicated therein.

TABLE IV

Influence of heat treatment—stress rupture

Melt	Test Temp. (° F.)	Test Stress (p. s. i.)	H. T. 2,050° F.—1 hr.—oil+1,325° F.—16 hr.—air			H. T. 2,050° F.—1 hr.—oil+1,550° F.—24 hr.—air+1,400° F.—16 hr.—air.		
			Rupture Life, Hr.	Percent Elong.	Percent Reduction of Area	Rupture Life, Hr.	Percent Elong.	Percent Reduction of Area
H-400	1,200	80,000	42	0.0	4.1	86	12.7	19.1
H-400	1,500	25,000	395	9.4	15.3	468	13.3	21.4
H-402	1,200	80,000	48	0.0	5.1	133	10.0	12.7
H-402	1,500	25,000	388	9.8	17.4	331	8.8	16.3
H-403	1,200	80,000	66	1.8	6.0	154	5.5	8.6
H-403	1,500	25,000	342	12.8	32.4	298	16.1	22.6
H-404	1,200	80,000	33	1.4	7.0	243	3.2	5.0
H-404	1,500	25,000	279	12.2	24.6	154	11.6	24.8
H-405	1,200	80,000	325	2.1	6.0	890	4.1	7.6
H-405	1,500	25,000	283	19.1	43.9	295	12.9	28.7
V-039B	1,200	80,000	127	3.1	7.0	69	11.7	36.5
V-039B	1,500	25,000	390	18.8	19.7	281	10.6	27.6
4X-374	1,200	80,000	31	0.3	3.0	148	8.4	13.3
4X-374	1,500	25,000	308	4.4	7.0	330	4.4	4.4

As clearly illustrated in Table IV where the alloy of this invention is to be used, for example, as turbine wheels and blades operating at a temperature of 1200° F., the two stage heat treatment is necessary to impart thereto sufficient ductility at this temperature. It is also apparent that the two stage heat treatment does not produce any substantially detrimental effect in the rupture life. What little decrease is apparent from the test results recorded in Table IV is clearly offset by the gain in ductility. On the other hand, if the alloy is to be used at temperatures of about 1500° F., it is seen that the single stage heat treatment is effective for imparting thereto the greatest rupture life consistent with the greatest ductility.

In order to more clearly demonstrate the advantage of the alloy of this invention, Heat No. 62918 was made and tested having the composition as set forth in Table II. This is one of the well-known commercial austenitic iron base alloys presently used in the form of a turbine wheel operating at temperatures of no greater than 1300° F. at high stress levels. Comparing the test results of this heat with those of Heat No. 4X–374, it is found that when the alloys are tested at a temperature of 1500° F. the alloy of Heat No. 62918 has a rupture life of 46 hours under a stress of 15,000 p. s. i. whereas Heat No. 4X–374 has a rupture life of 330 hours when tested at 25,000 p. s. i. and 1500° F. It is thus apparent that an enormous increase is noted both in the rupture life and the stress which was far in excess of the rupture life and strength of known prior art austenitic substantially iron base alloys. When the test results of Heat No. 4X–374 are compared with Heat No. M–252, a well-known nickel-cobalt base super alloy having the analysis set forth in Table II, substantially similar results are obtained. Thus when Heat No. M–252 was tested at 1500° F. and at a stress of 26,000 p. s. i., it had a rupture life of 100 hours, whereas Heat No. 4X–374 when tested at 1500° F. and 25,000 p. s. i., had a rupture life of 330 hours. It is thus apparent that the alloy of this invention possesses physical properties substantially comparable to those of some of the presently known and used super alloys. In addition, it is noted that when compared to the super alloys, the alloy of this invention contains a substantially lesser amount of strategic alloying elements thereby decreasing the cost and at the same time increasing the ease with which the alloy of this invention may be fabricated.

The alloy of this invention requires no special skills or knowledge in producing the alloy or in the heat treatment applied thereto. All operations can be performed on existing equipment by any one skilled in the metallurgical art. The alloy has excellent properties suitable for use at temperature of 1500° F. and uses a minimum amount of strategic alloying elements thereby rendering the alloy more readily available while substantially reducing the cost thereof in relation to presently known and used super alloys, yet the alloy has substantially similar properties to the presently known and used super alloys.

We claim:

1. An austenitic precipitation hardening alloy consisting of, up to 0.15% carbon, up to 1.5% manganese, up to 1.5% silicon, from about 10.0% to about 20% chromium, from about 30% to about 55% nickel, from about 2.5% to about 5.0% of the sum of aluminum and titanium, the aluminum being present in an amount of at least 0.5% and the titanium being present in an amount of at least 1.0%, from about 1.0% to about 8.0% molybdenum, from about 2.0% to about 15.0% tungsten, traces to 1.5% vanadium, and from about 20.0% to about 48.0% iron with incidental impurities.

2. An austenitic precipitation hardening alloy consisting of, up to 0.12% carbon, up to 1.5% manganese, up to 1.5% silicon, from 13.0% to 16.0% chromium, from 42% to 48% nickel, from 3.2% to 4.0% of the sum of aluminum and titanium, the aluminum being present in an amount of at least 0.5% and the titanium being present in an amount of at least 1.0%, from 1.0% to 4.0% molybdenum, from 4.0% to 9.0% tungsten, from 0.1% to 0.5% vanadium, and from about 20.0% to about 48.0% iron with incidental impurities.

3. An austenitic precipitation hardening alloy consisting of, about 0.03% carbon, about 0.62% silicon, about 1.36% manganese, about 15.0% chromium, about 45.3% nickel, about 3.9% molybdenum, about 4.0% tungsten, about 2.7% titanium, about 1.1% aluminum, about 0.3% vanadium, and the balance iron with incidental impurities.

4. An article of manufacture for use in highly stressed moving parts operating at temperatures of up to 1500° F. formed from an austenitic precipitation hardening alloy consisting of, up to 0.15% maximum carbon, up to 1.5% silicon, up to 1.5% manganese, from 10% to 20% chromium, from 30% to 55% nickel, from 2.5% to 5.0% of the sum of titanium and aluminum, the titanium being present in an amount of at least 1.0% and the aluminum being present in an amount of at least 0.5%, from 1.0% to 8.0% molybdenum, from 2.0% to 15.0% tungsten, from traces up to 1.5% vanadium, and from about 20.0% to about 48.0% iron with incidental impurities.

5. An article of manufacture for use as a moving part subject to high stresses at elevated temperatures of up to 1500° F. formed from an austenitic precipitation hardening alloy consisting of, up to 0.12% carbon, up to 1.5% manganese, up to 1.5% silicon, from 13.0% to 16.0% chromium, from 42% to 48% nickel, from 3.2% to 4.0% of the sum of aluminum and titanium, the aluminum being present in an amount of at least 0.5% and the titanium being present in an amount of at least 1.0%, from 2% to 4% molybednum, from 4.0% to 8% tungsten, from 0.1% to 0.5% vanadium, and from about 20.0% to about 48.0% iron with incidental impurities.

6. A precipitation hardened article of manufacture suitable for use as a moving part subject to high stress at temperatures of up to 1200° F. and formed of an austenitic alloy consisting of, up to 0.15% carbon, up to 1.5% manganese, up to 1.5% silicon, from 10% to 20% chromium, from 30% to 55% nickel, from 2.5% to 5.0% of the sum of aluminum and titanium, the aluminum being present in an amount of at least 0.5% and the titanium being present in an amount of at least 1.0%, from 1% to 8% molybdenum, from 2% to 15% tungsten, from traces up to 1.5% vanadium, and from about 20.0% to about 48.0% iron with incidental impurities, and which has been quenched from a solution temperature between 2000° F. and 2100° F., the solution temperature being maintained for a time period of between ½ hour and 2 hours, aged at a temperature between 1500° F. and 1600° F. for a time period of between 18 hours and 30 hours, followed by another aging at a temperature between 1350° F. and 1450° F. for a time period of between 12 hours and 20 hours.

7. A precipitation hardened article of manufacture suitable for use as a moving part subject to high stress at temperatures of up to 1500° F. and formed of an austenitic alloy consisting of, up to 0.15% carbon, up to 1.5% manganese, up to 1.5% silicon, from 10% to 20% chromium, from 30% to 55% nickel, from 2.5% to 5.0% of the sum of aluminum and titanium, the aluminum being present in an amount of at least 0.5% and the titanium being present in an amount of at least 1.0%, from 1% to 8% molybdenum, from 2% to 15% tungsten, from traces up to 1.5% vanadium, and from about 20.0% to about 48.0% iron with incidental impurities, and which has been quenched from a solution temperature between 2000° F. and 2100° F., the solution temperature being maintained for a time period of between ½ hour and 2 hours, and aged at a temperature between 1275° F. and 1375° F. for about 14 to 24 hours.

(References on following page)

References Cited in the file of this patent

UNITED STATES PATENTS

2,515,185	Bieber et al.	July 18, 1950
2,688,536	Callaway et al.	Sept. 7, 1954
2,781,264	Gresham et al.	Feb. 12, 1957

FOREIGN PATENTS

286,367	Great Britain	Mar. 5, 1928
980,191	France	Aug. 20, 1945
583,841	Great Britain	Jan. 1, 1947

583,845	Great Britain	Jan. 1, 1947
160,838	Australia	Sept. 10, 1953
1,103,237	France	May 18, 1955
1,114,287	France	Dec. 19, 1955

OTHER REFERENCES

Canadian Metals and Metallurgical Industries, January 1947, vol. 10, pages 22–25, 31.

Journal of Metals, February 1954, Nordheim and Grant, pages 211–218.