

(19)



Europäisches Patentamt
European Patent Office
Office européen des brevets



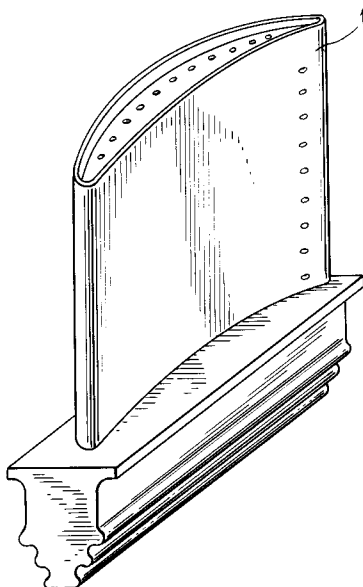
(11) Publication number:

0 561 179 A2

(12)

EUROPEAN PATENT APPLICATION(21) Application number: **93102779.1**(51) Int. Cl.⁵: **C22C 19/05**(22) Date of filing: **23.02.93**(30) Priority: **18.03.92 US 854142**(43) Date of publication of application:
22.09.93 Bulletin 93/38(84) Designated Contracting States:
BE CH DE ES FR GB IT LI NL SE(71) Applicant: **WESTINGHOUSE ELECTRIC
CORPORATION**
Westinghouse Building Gateway Center
Pittsburgh Pennsylvania 15235(US)(72) Inventor: **Narayana, Sastry Cheruvu**
3638 Jericho Drive
Casselberry, Florida 32707(US)(74) Representative: **Fleuchaus, Leo, Dipl.-Ing. et al**
Melchiorstrasse 42
D-81479 München (DE)(54) **Gas turbine blade alloy.**

(57) A nickel-based alloy suitable for use in making gas turbine components, such as rotating blades is provided. The alloy comprises the following elements in weight percent: Chromium 14.75 to 16.0, Cobalt 8.0 to 8.5, Aluminum 3.4 to 4.0, Titanium 3.4 to 4.3, Aluminum plus Titanium 7.7 to 8.3, Tantalum 1.75 to 2.7, Tungsten 2.0 to 4.0, Columbium up to .5, Molybdenum up to 2.0, Carbon .05 to .12 and the balance Nickel. The alloy of the current invention may also comprise impurities and incidental elements generally associated with nickel-based alloys, such as Zirconium up to .06 and Boron up to .015 percent by weight.

***Fig. 2*****EP 0 561 179 A2**

The current invention relates to an alloy suitable for use in making gas turbine components, such as the rotating blades in the turbine section of a gas turbine. More specifically, the current invention concerns a nickel-based alloy having a sufficiently high chromium content for good corrosion resistance yet maintaining high strength when used to make a directionally solidified turbine blade casting.

A gas turbine employs a plurality of rotating blades in its turbine section. Such blades are exposed to gas at temperatures in excess of 1100 °C (2000 °F) and subjected to high stress. Consequently, the alloys from which such blades are cast must, after suitable heat treatment, have very high stress rupture strength and sufficient metallurgical stability to maintain this strength for many thousands of hours of operation. Such alloys must also have sufficient ductility to withstand the large thermal stresses imposed on turbine blades.

In addition, as a result of impurities in the fuel and combustion air, the gases to which the blades are exposed contain corrosive compounds, such as sulfides and chlorides. Consequently, such blade alloys must also have good high temperature corrosion resistance, as well as oxidation resistance.

Generally, high temperature corrosion resistance is provided by the incorporation of substantial amounts of chromium into turbine blade alloys. High chromium content inhibits the basic fluxing of the alloy by forming a continuous chromia scale that is not susceptible to solution and reprecipitation from a Na₂SO₄ melt, thereby providing an effective barrier for the alloy. Moreover, chromium combines with sulfur to form high melting point sulfides, thereby inhibiting degradation due to sulfidation. As discussed in R. Streiff and D.H. Boone, "Corrosion Resistant Modified Aluminide Coating," Journal of Materials Engineering (1988), a minimum level of 15% chromium is considered necessary for good high temperature corrosion resistance.

One nickel-based alloy, used with some success for a number of years in gas turbine blades, is manufactured by the International Nickel Company and known commercially as IN-738. A typical composition of IN-738 in weight percent, as published in the American Society of Metal Handbook, 9th ed., volume 4, page 244, is as follows: Nickel 61, Chromium 16.0, Cobalt 8.5, Molybdenum 1.7, Tungsten 2.6, Tantalum 1.7, Niobium (aka Columbium) 0.9, Aluminum 3.4, Titanium 3.4, Carbon .17, Boron 0.01, and Zirconium 0.10. In addition, the manufacturer recommends that the electron vacancy number for this alloy not exceed 2.36. This alloy is disclosed in U.S. Patent No. 3,459,545 (Bleber), hereby incorporated by reference in its entirety.

It is known that directional solidification, whereby a uniaxial grain structure is produced, increases both the ability to withstand cyclical thermal stress and the stress rupture strength of many nickel-based alloys --for example, U.S. Patent No. 4,519,979 (Shaw) discloses that directional solidification of an alloy known commercially as IN-939 and having a composition in weight percent of Carbon 0.15, Chromium 22.5, Cobalt 19, Tungsten 2, Titanium 3.7, Aluminum 1.9, Tantalum 1.4, Niobium 1.0, Zirconium 0.1, Boron 0.01 and the balance Nickel, increased the stress the rupture life of the alloy from about 850 hours to 1370 hours at 870 °C and 200 N/mm². The inventor has confirmed, however, that directional solidification does little to increase the stress rupture strength of the IN-738 alloy. Figure 1 shows a graph of time to rupture versus stress at three temperature levels for both conventionally cast -- that is, having an equiaxed grain structure -- and directionally solidified IN-738 test specimens. As can be seen, data at both the 870 °C (1600 °F) and 925 °C (1700 °F) temperature levels indicates that the rupture life of the directionally solidified specimens is worse than that of the conventionally cast specimen above a stress level of about 275 N/mm² (40 KSI).

Generally, it has been observed that the alloys that derive the most improvement from directional solidification have three characteristics -- (i) a relatively low chromium content, (ii) a high gamma prime volume fraction and (iii) a high solution temperature. Consequently, it has previously been thought that the advantages of directional solidification could be obtained for alloys such as IN-738 only by reducing their chromium content -- see, for example, M. McLean, "Directionally Solidified Materials for High Temperature Service," published by Metals Society, London (1983), page 153. Thus, one popular alloy used in directionally solidified blades, known commercially as GTD-111, has a typical composition in weight percent of Chromium 14.0, Cobalt 9.5, Aluminum 3, Titanium 4.9, Tantalum 2.8, Tungsten 3.8, Molybdenum 1.5, Boron 0.01, Carbon 0.1 and the balance Nickel, as disclosed by R. Viswanathan in "Damage Mechanisms and Life Assessment of High-Temperature Components," published by the American Society of Metals (1989). As can be seen, in GTD-111 the benefits of directional solidification have been obtained by reducing the 16% chromium level used in IN-738 to only 14%. Unfortunately, as previously discussed, such relatively low levels of chromium result in inadequate corrosion resistance.

Accordingly, it would be desirable to provide a blade alloy that has high strength when directionally solidified yet sufficient chromium to achieve adequate corrosion resistance.

It is the principal object of the current invention to provide a nickel-based alloy suitable for use in casting the rotating blades of a gas turbine and having high stress rupture strength when directionally solidified, at least 15% by weight chromium for good corrosion resistance, and stability during long term service.

With this object in view, the present invention resides in a nickel-based alloy for a gas turbine blade, comprising the following elements in weight percent:

5	Chromium	14.75 to 16.0
	Cobalt	8.0 to 8.5
	Aluminum	3.4 to 4.0
	Titanium	3.4 to 4.3
	Aluminum plus Titanium	7.7 to 8.3
10	Tantalum	1.75 to 2.7
	Tungsten	2.0 to 4.0
	Carbon	.05 to .12
	Nickel	Balance.

15 The invention will become more readily apparent from the following description of a preferred embodiment thereof described with reference to the accompanying drawings, wherein:

Figure 1 is a graph of rupture time t , in hours, versus stress σ , in KSI (N/mm²), for both conventionally cast (CC) and directionally solidified (DS) specimens of IN-738, showing the effect of directional solidification on the stress rupture strength of IN-738.

20 Figure 2 is an isometric view of a gas turbine rotating blade.

Figure 3 is a bar chart of rupture time, t , in thousands of hours, versus stress σ , in KSI (N/mm²), showing the stress rupture life at 870 °C (1600 °F) and four stress levels of (i) four heats of the alloy according to the current invention, SAS1-SAS4, as directionally cast, (ii) conventionally cast IN-738 and (iii) directionally solidified IN-6203.

25 There is shown in Figure 2 a rotating blade 1 used in the turbine section of a gas turbine. The current invention is directed to an alloy, referred to as OM 200, from which such blades may be cast, especially using a directional solidification casting process. Generally speaking, the current invention is directed to an alloy comprising the following elements in weight percent: Chromium 14.75 to 16.0, Cobalt 8.0 to 8.5, Aluminum 3.4 to 4.0, Titanium 3.4 to 4.3, Aluminum plus Titanium 7.7 to 8.3, Tantalum 1.75 to 2.7,
30 Tungsten 2.0 to 4.0, Carbon .05 to .12, Columbium up to .5, Molybdenum up to 2.0, and the balance Nickel. The alloy of the current invention may also comprise impurities and incidental elements generally associated with nickel-based alloys, such as Zirconium up to .06 and Boron up to .015 percent by weight. Preferably, the alloy of the current invention consists essentially of the following elements in weight percent: Chromium about 15.5, Cobalt about 8.0, Aluminum about 4.0, Titanium about 3.8, Aluminum plus Titanium
35 about 7.8, Tantalum about 2.7, Tungsten about 2.6, Molybdenum 0.5, Carbon about 0.08 and the balance Nickel.

As previously discussed, both corrosion resistance and strength are important properties for blade alloys. Corrosion resistance in nickel-based alloys is provided primarily by chromium. Nickel-based alloys used for gas turbine components are strengthened by three mechanisms --(i) solid solution strengthening,
40 (ii) strengthening resulting from the presence of carbides and (iii) gamma prime strengthening. Solid solution strengthening is provided by molybdenum, chromium and tungsten and, to a lesser extent, by cobalt, iron and vanadium. Gamma prime strengthening is provided primarily by aluminum and titanium, which strengthen the austenitic matrix through the precipitation of Ni₃(Al and/or Ti), an fcc intermetallic compound. In addition, the aluminum in gamma prime can be replaced by tantalum and columbium.

45 Since the strength of nickel-based super alloys increases with increasing gamma prime volume fraction up to about 60%, in the alloy according to the current invention the content of aluminum, titanium, tantalum and columbium, which tend to increase the gamma prime volume fraction, have been controlled so as to achieve a high gamma prime volume fraction. The amount of gamma prime in an alloy can be determined as discussed below with respect to the determination of the electron vacancy number N_v . The amount of
50 gamma prime, in weight percent, in four heats of the alloy according to the current invention shown in Table I are approximately 52, 54, 56 and 54, respectively. The aforementioned preferred composition of the alloy according to the current invention has approximately 56% gamma prime. By comparison, the amount of gamma prime in IN-738 is approximately 50% by weight. (It should be noted that the aforementioned variation in the gamma prime content among the four heats of the alloy according to the current invention
55 did not adversely affect corrosion resistance or stability.)

As a result of operation at elevated temperature, nickel-based alloys undergo microstructural changes. Such changes include gamma prime coarsening, which adversely affects the strength of the alloy, and the transformation of gamma prime into unwanted topologically close-packed secondary phases, such as plate

or needle-like sigma, eta, etc. The formation of these plate-like phases adversely affects both strength and toughness. Consequently, in order to ensure that high strength and toughness are maintained for many thousands of hours of operation at elevated temperature, the composition of the strengthening elements must be carefully balanced, as explained below, so that the alloy has a certain degree of microstructural stability.

It is well known that the electron vacancy number N_v is a measure of the microstructural stability of the alloy. The higher the value of N_v the greater the tendency for the formation of the aforementioned topologically close-packed secondary phases -- see for example, the American Society of Metal Handbook, 9th ed., volume 4, page 278. As disclosed in the brochure entitled "Alloy IN-738, Technical Data," published by The International Nickel Company and hereby incorporated by reference in its entirety, the electron vacancy number may be calculated from the equation $N_v = 0.66 \text{ Ni} + 1.71 \text{ Co} + 2.66 \text{ Fe} + 4.66 (\text{Cr} + \text{Mo} + \text{W}) + 6.66 \text{ Zr}$. The atomic percent of each element in the matrix to be substituted into this equation is determined by converting the composition from weight percent to atomic percent and assuming that (i) one-half the carbon forms MC in the order of TaC, CbC, TiC, (ii) the remaining carbon forms M_{23}C_6 with the M comprising twenty three atoms of Cr, (iii) boron is combined as Mo_3B_2 , (iii) gamma prime is $\text{Ni}_3\text{-(Al, Ti, Ta, Cb)}$, and (iv) the residual matrix consists of the atomic percent minus those atoms contained in the carbides, the boride and the gamma prime reaction so that the total of the remaining atomic percentages gives the atomic concentration in the matrix. Conversion of these atomic percentages to a 100 percent basis gives the atomic percentage of each element to be substituted into the equation above. Since it is desirable to maintain a low value of N_v in the alloy according to the current invention the composition is adjusted so that, in addition to obtaining high strength, the electron vacancy number of the alloy does not exceed about 2.4. In the preferred composition of the alloy the electron vacancy number is equal to about 2.4.

As previously discussed, the aluminum in gamma prime can be replaced by columbium and/or tantalum, as well as titanium. However, since replacement of Al by Ti, Cb or Ta in Ni_3Al adversely affects the misfit between the austenitic matrix and the gamma prime particles, the stability of these compounds in order of decreasing stability is Ni_3Al , Ni_3Ti and $\text{Ni}_3\text{Cb(orTa)}$. As a result, the titanium/aluminum ratio plays a major role in gamma prime coarsening. The titanium/aluminum ratio also plays a major role in the transformation of gamma prime into the aforementioned unwanted plate-like phases. Accordingly, in order to minimize gamma prime coarsening and to improve the stability of the gamma prime phase, the titanium/aluminum ratio is generally maintained below 2:1 (by weight). The transformation of gamma prime into unwanted needle or plate-like phases can also be retarded by the addition of tungsten.

To assess the stability of the alloy according to the current invention, the shoulder and grip sections of creep rupture specimens tested at 871°C (1600°F) and 206.9 N/mm^2 (30 KSI) were examined metallographically for sigma, eta and mu phases. It was found that the aging of these specimens at 871°C , under stress, for up to 4000 hours did not produce any undesirable phases, thereby suggesting that the alloy of the current invention is stable.

In the alloy according to the current invention, the amounts of aluminum, titanium, tantalum, columbium and molybdenum have been balanced so as to attain high strength when the alloy is directionally solidified while maintaining good microstructural stability. Significantly, this result has been achieved without the need to reduce the chromium content, and, therefore, without impairing corrosion resistance, as had heretofore been thought necessary by those skilled in the art. Specifically, in the alloy according to the current invention, the content of aluminum and titanium has been increased, when compared with IN-738, to a minimum of 3.4% for each, with the minimum combined aluminum plus titanium content being 7.7%. The maximum amounts of aluminum and titanium are 4.0% and 4.3%, respectively, with the maximum aluminum plus titanium content being 8.3%. In addition, the amounts of columbium and molybdenum have been reduced so that the optimum composition includes no columbium and only 0.5% molybdenum. At most only 0.5% columbium and 2.0% molybdenum are permitted in the alloy of the current invention. Chromium, however, has been maintained in the range of 14.75% to 16%, so that adequate corrosion resistance is maintained. Tungsten is maintained in the range of 2.0% to 4.0% and tantalum in the range of 1.75% to 2.7%. Zirconium and boron are limited to 0.06% and 0.015%, respectively, with none of either of these elements being present in the most preferred composition. Carbon is maintained in the 0.05% to 0.12% range. Moreover, as previously discussed, in the alloy according to the current invention, the elements are adjusted within the aforementioned ranges so that the electron vacancy number is maintained at no more than about 2.4, thereby ensuring that adequate microstructural stability is achieved.

As a result of the foregoing, the strength of the alloy according to the current invention when cast by a directional solidification process is high, despite its relatively high chromium content. Significantly, good microstructural stability of the alloy according to the current invention has been achieved by considerably

reducing the levels of columbium and molybdenum, when compared to IN-738, so that the amounts of aluminum and titanium can be increased without driving the electron vacancy number too high.

Four heats of the alloy according to the current invention, designated SAS1 through SAS4, were prepared using the compositions in weight percent shown in Table I. Specimens were cast from these heats so as to be directionally solidified and were given the standard heat treatment suggested for IN-738 -- that is, a 1121 °C (2050 °F) solution heat treatment for two hours followed by a 843 °C (1550 °F) aging treatment for twenty four hours. These specimens were subjected to a variety of tests and the results compared to conventionally cast IN-738 and directionally solidified IN-6203. IN-6203 is a nickel-based alloy having a nominal composition in weight percent of Chromium 22.0, Cobalt 19.0, Aluminum 2.3, Titanium 3.5, Tantalum 1.10, Columbium 0.80: Tungsten 2.00, Boron 0.01, Carbon 0.15, Zirconium 0.10, Hafnium 0.75 and the balance Nickel.

TABLE I

Element	SAS1	SAS2	SAS3	SAS4
Cr	16.00	16.00	14.75	16.00
Co	8.50	8.00	8.00	8.00
Al	3.40	3.40	3.60	4.00
Ti	4.30	4.30	4.30	3.40
Al + Ti	7.70	7.70	7.90	7.40
W	2.60	2.60	2.60	2.60
Mo	1.75	0.90	1.00	0.60
Ta	1.75	2.70	2.70	2.70
Cb	0.00	0.00	0.50	0.00
C	0.09	0.09	0.09	0.09
Zr	0.04	0.04	0.04	0.04
B	0.01	0.01	0.01	0.01
Ni	Bal	Bal	Bal	Bal

The results of low fatigue testing at 871 °C (1600 °F) of specimens from heats SAS1, SAS3 and SAS4 are shown in Table II and indicate that the cyclic life of the alloy according to the current invention is superior to IN-738. The results of impact testing of three specimens from heat SAS1 -- taken from the tip, mid-section and bottom of the cast test bar, respectively -- are shown in Table III and reveal that the impact strength of the alloy of the current invention is comparable to that of IN-738. Note that IN-738 is provided in both low and high carbon forms and the data is reported accordingly where appropriate. In its low carbon form the carbon content is reduced from 0.17% to 0.11% and zirconium is reduced from 0.10% to 0.05%, by weight, compared to conventional high carbon IN-738.

The results of stress rupture tests are shown in Tables IV and V and Figure 3, wherein the rupture life of specimens from each of the four heats are compared with those of conventionally cast IN-738 and directionally solidified IN-6203. As can be seen, the data shows that the stress rupture life of the alloy according to the current invention, when directionally solidified, is significantly superior to both that of both IN-738 and IN-6203 -- for example, at 69 N/mm² (10 KSI), the alloy of the current invention can provide the same rupture life as IN-738 and IN-6203 at a temperature about 56 °C (100 °F) higher. When applied to a gas turbine blade, such increased metal temperature capability allows the turbine to operate at higher gas temperatures, thereby significantly improving its performance.

TABLE II

Total Strain Range	Cycles to Failure			
	IN-738LC	SAS1	SAS3	SAS4
0.012	-	1412	1442	1061
0.01	317	2990	3973	2852
0.0085	862	5342	6673	11200
0.0075	-	16220	50173 +	10504
0.007	1510	-	-	-
0.006	1819	-	-	-
0.005	2503	-	-	-
0.004	10153	-	-	-

TABLE III

Temp., °C	Impact Energy, Joules				
	IN-738HC	IN-738LC	SAS1		
			Tip	Mid	Bot
24	6.3	8.8	6.8	-	8.1
260	6.9	-	7.9	11.3	16.7
538	11.7	-	7.5	9.6	12.1
649	7.6	10.0	6.3	8.7	14.1
760	7.1	8.5	6.0	7.3	9.9
871	8.1	9.8	9.6	9.1	14.4

Table IV

Temp., °C	Time to Rupture, Hours @ 69 N/mm ² (10 KSI)		
	IN-738	IN-6203	SAS1
954	4043	-	-
968	2696	-	-
982	1437	2752	8400 +
996	820	1359	6450 +
1010	-	485	3745

Table V

Stress N/mm ² (KSI)	Time to Rupture, Hours @ 871 °C				
	IN-738	IN-6203	SAS1	SAS3	SAS4
275.8 (40)	188	169	403	633	317
241.3 (35)	420	683	1165	1397	776
206.9 (30)	1143	1944	3272	4211	2676
172.4 (25)	3177	-	8800 +	8800 +	8247

Turbine blades cast from the alloy of the current invention are advantageously made by vacuum-induction melting and vacuum casting using a directional solidification process. Directional solidification causes the grain boundaries to be oriented substantially parallel to the principal stress axis of the blade with

almost no grain boundaries oriented normal to the principal stress axis. Techniques for directional solidification are well known in the art -- see, for example, U.S. Patent Nos. 3,260,505 (Ver Snyder), 3,494,709 (Piercey) and 3,897,815 (Smashey), hereby incorporated by reference in their entirety.

The gamma prime distribution depends on heat treatment, as well as composition. The standard heat treatment for nickel-based alloys such as IN-738 -- i.e., a solution treatment followed by an aging treatment -- produces duplex gamma prime comprised of coarse, cuboidal primary gamma prime and fine, spherical gamma prime in approximately equal amounts. The coarse gamma prime is undissolved gamma prime that did not go into solution during the solution treatment. Hence, the amount of coarse gamma prime present in the alloy depends on the degree by which the solution temperature is below the gamma prime solvus temperature, at which all of the gamma prime goes into solution -- that is, the lower the solution temperature, the greater the amount of coarse gamma prime. The fine gamma prime forms during the aging treatment, the amount depending on the amount of gamma prime that did not go into solution during solution treatment.

Directionally solidified components are cooled at a slower rate than conventionally cast components. As a result, directional solidification produces coarser primary gamma prime so that the volume fraction of cuboidal gamma prime is higher and that of spherical gamma prime is lower than in directionally cast components. Unfortunately, a high volume fraction of coarse gamma prime has a deleterious effect on strength. This discrepancy in gamma prime distribution can be minimized by optimizing the heat treatment, particularly the solution treatment, based on the gamma prime solvus and incipient melting temperatures, both of which depend on alloy composition. Generally, it is desirable that the heat treatment be performed at a temperature between the solvus temperature and the incipient melting temperature. Hence, in order to obtain the full benefits of heat treatment, the solvus temperature should be considerably below the incipient melting temperature. Unfortunately, although aluminum, titanium, tantalum and columbium increase the gamma prime volume fraction, and therefore, strength, as previously discussed, they also have the effect of raising the solvus temperature and decreasing the incipient melting temperature, thereby narrowing the heat treatment window.

The gamma prime solvus and incipient melting temperatures for three melts of the alloy according to the current invention are shown in Table VI. These temperatures were determined using the differential thermal analysis and gradient bar -- i.e., metallography -- method, in which the bar was exposed to various temperatures in the 1066°C (1950°F) to 1427°C (2300°F) temperature range for four hours and then fan cooled. As can be seen, the solvus temperature varies from 1211°C to 1229°C. By comparison, the solvus temperature for IN-738 is approximately 1204°C (2200°F). As previously mentioned, IN-738 is typically given a 1121°C (2050°F) solution heat treatment for two hours followed by a 843°C (1550°F) aging treatment for twenty four hours. Various heat treatment regimes for the alloy according to the current invention, based on its incipient melting and solvus temperatures, are discussed further below.

TABLE VI

	SAS1	SAS3	SAS4
Incipient melting temp. C	1241	1229	1229
Gamma prime solvus temp., C	1211	1211	1229

According to the current invention, the as-cast blades may be heat treated in any of four ways -- (i) solution treating for 2 hours at 1121°C (2050°F), followed by aging for 24 hours at 843°C (1550°F), (ii) solution treating for 4 hours at 1149°C (2100°F), followed by aging for 24 hours at 843°C (1550°F), (iii) solution treating for 4 hours at 1204°C (2200°F), resolution treating for 2 hours at 1121°C (2050°F), followed by aging for 24 hours at 843°C (1550°F), and (iv) solution treating for 4 hours at 1204°C (2200°F), resolution treating for 4 hours at 1149°C (2100°F), followed by aging for 24 hours at 843°C (1550°F). These heat treatment optimizes mechanical strength by creating a duplex gamma prime structure in a gamma matrix and discrete chrome-carbides at the grain boundaries. The use of higher solution temperatures will decrease the amount of coarse gamma prime and increase the amount of fine gamma prime, hence, further strengthening the alloy.

Although the current invention has been described with reference to an alloy for casting turbine blades, it should be understood that the alloy according to the current invention may also be used to form other components that require high strength at elevated temperature and good corrosion resistance. Accordingly, the present invention may be embodied in other specific forms without departing from the spirit or essential attributes thereof and, accordingly, reference should be made to the appended claims, rather than to the

foregoing specification, as indicating the scope of the invention.

Claims

- 5 1. A nickel-based alloy for a gas turbine blade, comprising the following elements in weight percent:

10

15

Chromium	14.75 to 16.0
Cobalt	8.0 to 8.5
Aluminum	3.4 to 4.0
Titanium	3.4 to 4.3
Aluminum plus Titanium	7.7 to 8.3
Tantalum	1.75 to 2.7
Tungsten	2.0 to 4.0
Carbon	.05 to .12
Nickel	Balance.

20

2. An alloy according to claim 1, wherein the electron vacancy number of said alloy is no more than about 2.4.

3. An alloy according to claim 1, wherein the weight percents of tungsten and carbon are about 2.6 and about 0.09, respectively.

25

4. An alloy according to claim 1, further comprising the following elements in weight percent:

30

Columbium	0.0 to .5
Molybdenum	0.0 to 2.0
Zirconium	0.0 to .06
Boron	0.0 to .015.

35

5. An alloy according to claim 4, wherein the weight percent of tungsten is about 2.6, carbon is about 0.09, molybdenum is 0.60 to 1.75, zirconium is about 0.04 and boron is about 0.01.

6. An alloy according to claim 1, wherein the alloy comprises the following elements in weight percent:

40

45

50

Chromium	about 15.5
Cobalt	about 8.0
Aluminum	about 4.0
Titanium	about 3.8
Aluminum plus Titanium	about 7.8
Tantalum	about 2.6
Tungsten	about 2.6
Carbon	about 0.08
Molybdenum	about 0.5
Nickel	Balance.

55

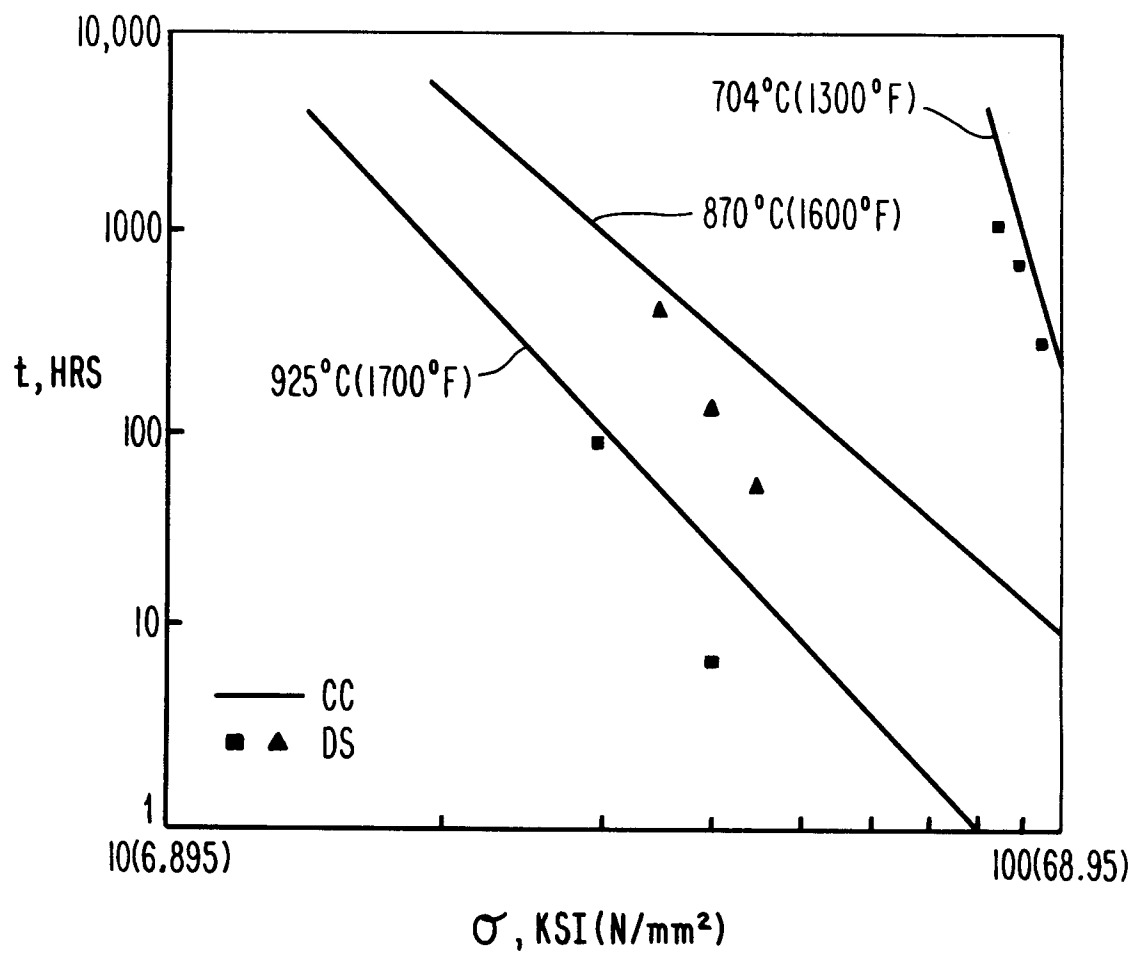


Fig. 1

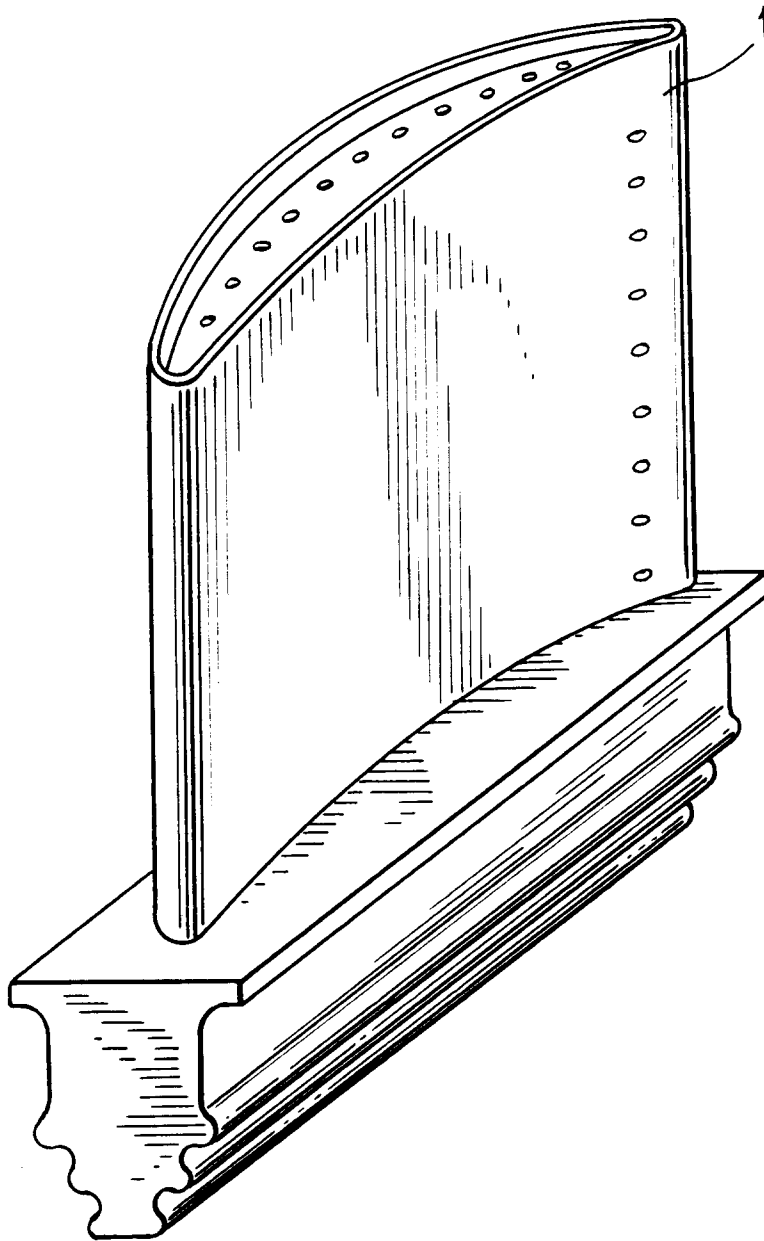


Fig. 2

Fig. 3

