Jan. 12, 1971 E. R. THOMPSON 3,554,817

CAST NICKEL-COLUMBIUM-ALUMINUM ALLOY

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FIG. 1

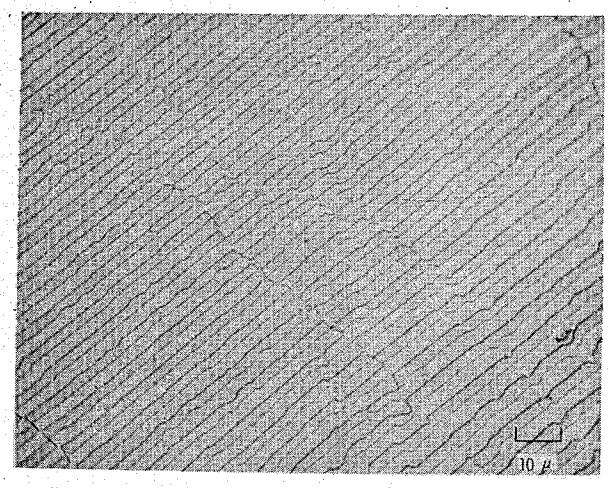
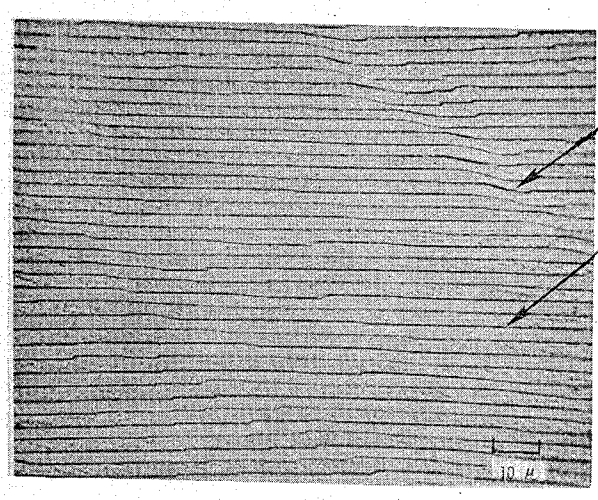


FIG. 2



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CAST NICKEL-COLUMBIUM-ALUMINUM ALLOY

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FIG. 3

ULTIMATE TENSILE STRENGTHS OF Ni_3Al-Ni_3Cb EUTECTIC & CONVENTIONALLY CAST ALLOY

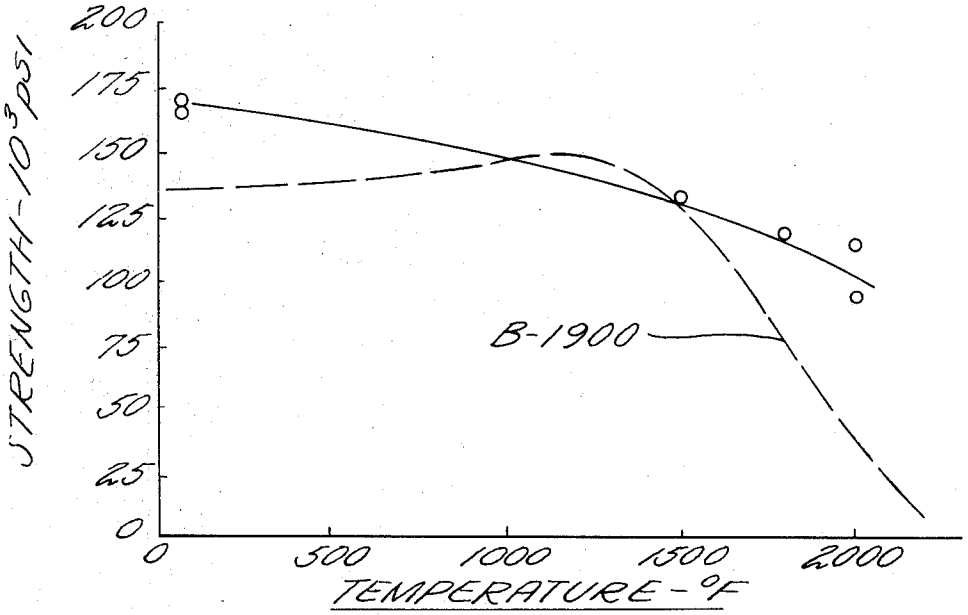
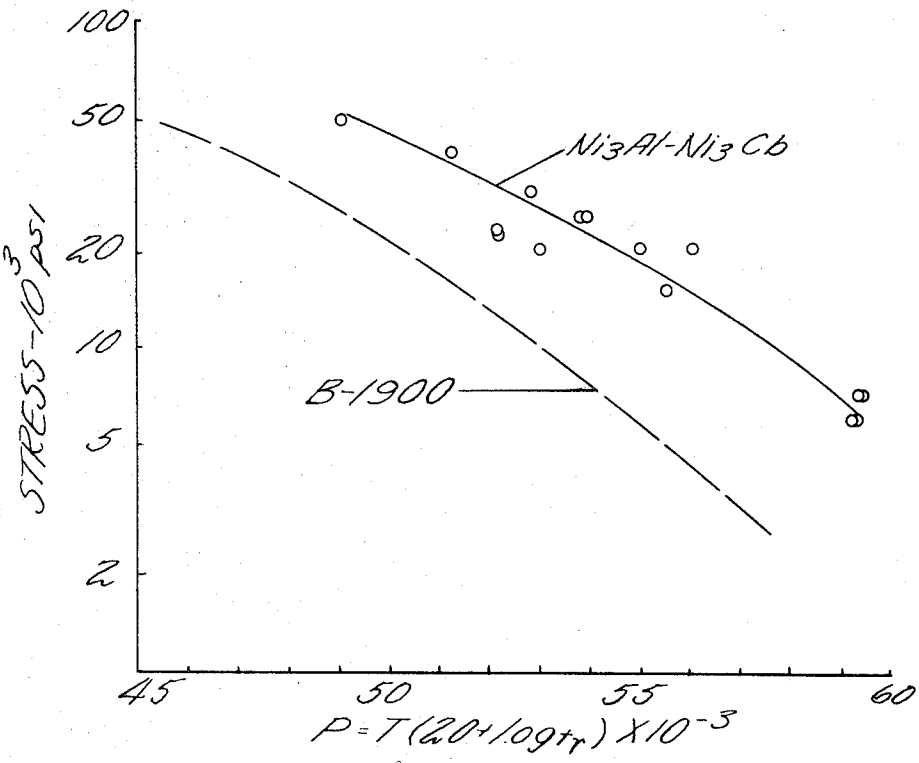


FIG. 4

RUPTURE STRENGTH OF Ni_3Al-Ni_3Cb EUTECTIC AND CONVENTIONALLY CAST ALLOY



United States Patent Office 3,554,817 Patented Jan. 12, 1971

3,554,817

CAST NICKEL-COLUMBIUM-ALUMINUM ALLOY

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Filed Mar. 20, 1969, Ser. No. 808,956
Int. Cl. B22d 21/00; C22c 15/00
U.S. Cl. 148—32 8 Claims

ABSTRACT OF THE DISCLOSURE

The pseudo-binary eutectic alloy occurring between the intermetallic compounds Ni_3Al and Ni_3Cb is unidirectionally solidified into a casting characterized by an aligned, lamellar microstructure.

BACKGROUND OF THE INVENTION

The present invention relates in general to the nickel-base alloys and, more particularly, to a nickel-columbium-aluminum alloy at about the eutectic composition, as directionally solidified to provide a lamellar microstructure.

The demands of today's technology require a continuous search for new alloys adapted to high temperature structural applications. One of the areas to which considerable research and development activity has been furnished is that of the fibrous composite materials. Strengthening of metal matrices has been accomplished utilizing a variety of reinforcing materials, including vapor grown oxide, carbide and nitride whiskers, carbon and refractory metal wires.

It is known that many of the eutectic-type alloys respond to plane front solidification to produce phase-aligned microstructures, as described in the patent to Kraft 3,124,452 which shares a common assignee with the present invention. In some cases, casting of the eutectic alloys as described by Kraft has produced in situ strengthening of a metal matrix by a stiffer, less ductile phase and has avoided many of the difficulties normally encountered in the fabrication and use of composite materials produced by other methods.

SUMMARY OF THE INVENTION

The present invention makes available the strongest nickel-base alloy known. It includes, as its essential component, the eutectic alloy existing between the intermetallic compounds Ni_3Al and Ni_3Cb at about 33 weight percent Ni_2Al, as directionally solidified to provide a controlled lamellar microstructure.

BRIEF DESCRIPTION OF THE DRAWINGS

FIG. 1 is a photomicrograph of a transverse section of a Ni_3Al-Ni_3Cb eutectic alloy unidirectionally solidified at a rate of 2.2 cm./hr.

FIG. 2 is a photomicrograph of the same alloy in longitudinal section, clearly revealing the pronounced anisotropy of the ingot as evidenced by the lamellar microstructure thereof.

FIG. 3 is a graph comparing the tensile strength of the Ni_3Al-Ni_3Cb alloy as directionally solidified with an advanced cast nickel-base superalloy.

FIG. 4 is a plot of the Larson-Miller parameter curves for rupture comparing the aligned Ni_3Al-Ni_3Cb eutectic with an advanced cast nickel-base superalloy.

DESCRIPTION OF THE PREFERRED EMBODIMENTS

A eutectic existing in the nickel-columbium-aluminum alloy system occurs at about the composition, by weight, 72.5 percent nickel, 23.1 percent columbium and 4.4 percent aluminum. This alloy exhibits pseudo-binary behavior between the Ni_3Al and the Ni_3Cb intermetallic compounds at about 33 weight percent Ni_3Al and, upon plane front solidification, produces a controlled lamellar microstructure containing about 44 volume percent of Ni_3Cb within the Ni_3Al matrix. The phase alignment is clearly evident from the photomicrographs of FIGS. 1 and 2.

It is well recognized in the high temperature alloy development field that nickel-base superalloys may be strengthened by precipitates of γ', a phase based on the Ni_3A intermetallic. Such a precipitate is found in the typical nickel-base superalloy such as B–1900 which has a nominal composition, by weight, of 8 percent chromium, 10 percent cobalt, 1 percent titanium, 6 percent aluminum, 6 percent molybdenum, 0.11 percent carbon, 4.3 percent tantalum, 0.15 percent boron, 0.07 percent zirconium, balance essentially nickel.

For the sake of brevity, the alloy of the present invention is hereinafter referred to as the eutectic alloy. It will be understood, however, that, particularly in connection with the mechanical property considerations, this term is utilized to refer to the Ni_3Al-Ni_3Cb eutectic alloy as unidirectionally solidified to yield an aligned, lamellar microstructure.

The eutectic temperature of the present alloy lies at about 2336° F., and the solid solubility of the Ni_3Cb phase within the γ' phase extends to about 40 weight percent columbium, while the solubility of Ni_3Al in the Ni_3Cb phase is less than about 4 weight percent. The eutectic exhibits coupled plane front growth when solidified unidirectionally at 2 cm./hr., growing with a lamellar morphology at a lamellar spacing of approximately 4 microns. Numerous off-eutectic castings were made to demonstrate that a lamellar morphology could be attained, as suggested by Kraft, supra, and an examination of these off-eutectic specimens showed that the nucleating phase for the eutectic is the Ni_3Cb phase and that dendrites of the γ' phase are "haloed" by the Ni_3Cb phase from which the eutectic grains grow.

Master melts of eutectic compositions were made in alumina crucibles and chill cast in copper molds. The resulting ingots were unidirectionally solidified vertically within a resistively heated graphite tube furnace under a dynamic argon atmosphere. The molten eutectic alloy was contained in a ½ inch diameter 99.7 percent recrystallized alumina, cylindrical crucible held in graphite sleeve which separated the crucible bottom from a water cooled brass plug by one inch of graphite. Crucible lowering rates of about 2 cm./hr. and liquid superheats of about 300° C. were generally employed, with a resultant thermal gradient in the liquid of approximately 70° C./cm.

Tensile and creep-rupture specimens were ground from the unidirectionally solidified ingots such that the load would be applied parallel to the direction of growth. The specimens typically were formed with gage sections having a diameter and length of 0.140 and 1.1 inch, respectively. Tensile tests were performed with a Tinius-Olsen four screw machine. Creep rupture testing was carried out in air under constant load.

The longitudinal tensile properties of the eutectic alloy were measured as a function of temperature and the results are reported in Table I. The tensile strength of the alloy is strain-rate dependent at 2000° F.

TABLE I

Temp., ° F.	Elastic modulus, 10^6 p.s.i.	Tensile strength p.s.i.	Percent elongation at fracture	Loading rate, in./min.
75	32.2	169,800	0.56	0.01
75	35.8	170,100	0.60	0.01
1,500		131,000	5.8	0.015
1,800		119,900	1.4	0.07
2,000		94,500	10.0	0.01
2,000		115,300	3.0	0.07

The tensile strength of the eutectic alloy is plotted in FIG. 3 as a function of temperature where it is compared with B–1900, an advanced cast nickel-base superalloy. The marked superiority of the present alloy at the higher temperatures is evident. At 2000° F. it is in fact more than twice as strong as the B–1900 alloy. Furthermore, its tensile strength is remarkably insensitive to temperature.

Of perhaps greater importance in the usual applications of this type of material are its creep-rupture properties. These are summarized in Table II.

TABLE II

Temp., ° F.	Stress, p.s.i.	Time to 1%, hr.	Time to rupture, hr.	Final elongation, percent
1,800	40,000	340	475	
1,800	50,000	38	56	4.0
2,000	15,000	178	409	8.8
2,000	20,000	20	37	6.8
2,000	20,000	67	224	11.9
2,000	20,000	166	310	12.0
2,000	20,000		634	
2,000	22,000		18	2.7
2,000	22,500	8	15	5.3
2,000	25,000		78	
2,000	25,000	48	89	
2,000	30,000		32	
2,200	6,000	63	200	
2,200	7,000		214	4.1
2,200	7,000	20	223	

Time-temperature rupture data for the present alloy and the B–1900 alloy is presented in FIG. 4 as a Larson-Miller plot assuming a constant of 20 for both alloys. The superiority of the eutectic alloy to B–1900 is evident. Moreover, the directionally-solidified Ni_3Al-Ni_3Cb eutectic alloy exhibits rupture strengths in excess of those of any reported nickel-base alloy.

The room temperature elastic modulus and 0.2 percent offset compressive yield strength have been established at 34×10^6 p.s.i. and 200,000 p.s.i., respectively. Failure in compression occurs at a strain of approximately 10 percent and a stress of 340,000 p.s.i., both ductile flow and interfacial cleavage occurring in the process.

The lamellar morphology of the alloy is thermally stable at least to temperatures as high as 2200° F. Specimens which were heat treated for up to 200 hours at 2200° F. showed no measurable coarsening of the lamellae.

In assessing the strength characteristics of the Na_3Al-Ni_3Cb eutectic alloy, it will be remembered that the testing was performed in air. In extended testing at high temperatures in air an oxide scale was produced which at times was significant in terms of the loss in specimen cross-section, amounting to approximately ten percent for 500 hours at 2000° F. Accordingly, the superiority in strength of this alloy, while far surpassing that of the advanced nickel-base superalloys in air, nevertheless was diminished for comparative purposes by the effects of the oxidation. Of course, it is normal to provide surface protection to the vast majority of the nickel-base superalloys in dynamic oxidizing environments at high temperature.

Those alloys which deviate from the true eutectic compositions may nevertheless be capable of solidification to produce the desired lamellar microstructure. Experience with the unidirectional solidification of eutectics has shown that such alloy frequently comprise a eutectic portion and a proeutectic portion, i.e., a portion which does not undergo a eutectic reaction. The resulting castings of such alloys often comprise a lamellar microstructure, as described herein, having distributed therein relatively large proeutectic crystals throughout the lamellar microstructure in either random or uniform distribution.

In general, deviations from or additions to the basic eutectic composition may be made as long as they do not interfere with the basic coupled plane front growth mechanism by which the lamellar microstructure is formed. Such deviations may be advisable in certain applications to impart or improve a specific property of the basic alloy. Even in such cases, however, the strength advantages to be achieved from a major portion comprising the Ni_3Al-Ni_3Cb eutectic will be desirable.

While the invention has been described in detail with relation to specific preferred embodiments and examples, in its broader aspects it is not limited thereto, for obvious modifications will occur to those skilled in the art.

What is claimed is:

1. A cast Ni base alloy containing Al and Cb contains a multiphase lamellar microstructure, a first phase consisting essentially of the Ni_3Al intermetallic compound, a second phase consisting essentially of the Ni_3Cb intermetallic compound, the first and second phases consisting of lamellae arranged in alternation in the cast article and in substantial alignment over a major portion of the length thereof.

2. A metallic article according to claim 1 wherein the lamellae are arranged in substantial alignment with each other and with the major tensile strength axis of the article.

3. A cast Ni base alloy containing Al and Cb characterized by an aligned lamellar microstructure over the major portion of the length thereof formed from an alloy which consists essentially of the pseudo-binary eutectic between the Ni_3Al and Ni_3Cb intermetallic compounds.

4. A metallic article according to claim 3 wherein the alloy consists essentially of about 33 weight percent Ni_3Al, balance Ni_3Cb.

5. A cast gas turbine engine blade or vane which comprises an alloy consisting essentially of about 33 weight percent Ni_3Al, balance Ni_3Cb unidirectionally solidified parallel to the tensile axis of the blade or vane to produce a lamellar microstructure wherein the Ni_3Al and Ni_3Cb components comprises respective lamellar phases which are arranged in alternation in the microstructure.

6. The method of forming gas turbine engine components which comprises:

melting an alloy which consists essentially of about 72 percent nickel, 23 percent columbium and 4 percent aluminum; and

unidirectionally solidifying the alloy utilizing a solidification rate about 2 cm./hr. to produce a lamellar microstructure.

7. The method according to claim 6 wherein:

the alloy is solidified in a direction substantially parallel to the designed tensile axis of the component.

8. The metthod according to claim 6 wherein:

about 300° C. superheat is provided in the alloy in the solidification process.

References Cited

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CHARLES N. LOVELL, Primary Examiner

U.S. Cl. X.R.

75—135, 175.5; 148—3; 164—122