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(54) **New superalloys and the methods for improving the properties of superalloys.**

(57) This invention has provided a new technique of improving the properties of superalloys by reducing the segregation of the alloying elements. The characteristic specificity is to reduce the phosphorus content of the alloys to below 0.001 (in wt%) and, in case of Ni-base superalloys, to below 0.0005(wt%). It is needed to eliminate Zr from the superalloys. Meanwhile, it is also needed to control the proper content of B and to reduce the content of Si to below 0.005. The invention has also provided a series of low segregation superalloys.

EP 0 387 976 A2

New Superalloys and The Methods For Improving the Properties of Superalloys

Background of the Invention

The present invention relates to improved superalloys and methods for improving superalloys.

Superalloys are important materials used as hot components of gas turbines in aeroplanes, warships and industrial and traffic machines. Recently, they have also been widely used in aerospace nuclear reactors and chemical industries, etc. How to improve their resistant ability to oxidation and corrosion, and raise they strength and service temperature are of significant importance. And researches on these problems have been paid much attentions by many metallurgists and designers in the world. Since the late forties, superallloys have been developed rapidly. However, the progress of the development and improvement in composition design of superalloys has been slowed down after 1970's. the main reasons are as follows:

(1) The compositions of the persent superalloys are extremely complicated. In case of adding more alloying elements, it will lead to the formation of severe solidification segregation in the castings. For instance, on nickel base superalloys more addition of Al and Ti is beneficial to raising their strength at elevated temperature and Cr benefits to increasing their corrosion resistance. However, the excess of the these elements will make the superalloys unusable due to the appearance of brittle phase.

(2) With increasing the addition of alloying elements, the incipient melting temperature of the superalloys decreases acordingly. For instance, the starting forging temperature of Fe-Ni-Cr-base superalloys must be lower than 1140° C because of the limitation of the incipient melting temperature. Under such a low forging temperature, it is necessary to use huge forging machine and special technique to manufacture the gas turbine disks with tese superalloys due to the large deformation resistance. Therefore the production costs are increased correspondingly.

(3) In most cases, it is difficult to overcome the contradictions of the service properties at elevated temperature with the machinig and welding properties through adjustments of alloy composition. The aim of this invention is to advance new superalloys with the incipient melting temperature raised, the segration of alloying elements being reduced, precitation of harmful phases being eliminated and eventually the technological properties of the superalloys being improved effectively, and new methods for improving the properties of superalloys.

Summary of the Invention

The alloys of the present invention contain at least about 25% nickel (all percents expressed herein and in the claims are by weight unless otherwise specified) and up to about 0.001% phosphorus, for nickel-base superalloys the phosphorus content should be lower than 0.0005%, up to about 0.05% silicon, about 0.001 to about 0.015% boron. The balance of the alloy will consist of other elements which are conventionally alloyed with nickel to form superalloys such as elements selected from the group consisting of chromium, iron, cobalt, molybdenum, titanium, tantalum, tungsten, aluminum, niobium, carbon, vitriol and combinations thereof. These alloys will be zirconium-free.

Generally, the alloy for the cast consists essentially of about 6.0% to about 22.0% chromium, up to about 16.0% cobalt, up to about 8% molybdenum, about 2% to about 5.5% titanium, about 2% to about 6.5 aluminum, up to about 12.0% tungsten, about 10 to about 150 ppm boron, up to about 4.0 niobium, up to about 12.0% tantalum, and about 0.02% to about 0.22% carbon, up to about 0.0005% phosphorus, up to about 0.05 silicon, hafnium-free, zirconium-free and the balance nickel.

Generally the alloy for the wrought consists essentially of about 25 to 55% nickel, about 0.01 to 0.1% carbon, about 0.1 to about 0.5% vitriol, about 10 to 22% chromium, up to about 3% tungsten, up to about 7% molybdenum, up to about 6% niobium, up to about 6% tantalum, about 0.1 to about 3% aluminum, about 0.5 to about 3% titanium, about 0.001 to about 0.01% boron, up to about 0.001% phosphorus, up to about 0.05% silicon, zirconium-free and the balance iron.

The invention provides a new method of raising the incipient temperature, reducing the segregation, eliminating technological properties of the superalloys effectively by means of strictly controlling the contents of P and B and eliminating Zr. The key points are: the P content in the present superalloys must be reduced below 0.001 (in wt%); for nickel-base superalloys the P content should be lower than 0.0005 (wt%); for the existing superalloys using Zr as the grain boundary strengthening element, besides reducing the P content less than 0.0005, it is also necessary to eliminate the Zr from the superalloys. By using this invention, the inventors developed a series of low segregration superalloys with excellent properties. For example, by using this new method the incipient melting temperature of Fe-Ni-Cr-base wrought superalloy GH901 (Namely Alloy IN 901) can be raised from 1140°C to 1260°C and the starting forging temperature can be raised to 1190-1225°C accordingly. At such temperature, the precision mould forging technology can be used very easily and the materials can be utilized very savingly without any loss in tensile and stress rupture properties, for cast superalloys the contents of Al and Ti can be added by 1% more without any loss in hot corrosion resistance and formation of harmful phases.

The comparison of 100h stress reupture strength data shows that the mechanical properties of the superalloys are greatly improved. What is more, the Cr content can be increased by 4% more in the low P content cast Ni-base superalloys, thus the resistance to oxidation and corrosion at elevated temperature of the alloys can be enhanced obviously.

This invention also provides the composition of new superalloy modified by the new method mentioned above. They are as follows: (1) an analogue of GH901 (IN 901) FE-Ni-Cr-base superalloys with low segregation. The characteristic specificity is to control the P content below 0.001. The composition of the alloy is: Ni 40-45%, Cr 11-14%, Mo 5.0-6.5%, Al</=0.3%, Ti 2.8-3.1%, B 0.001-0.004%, C 0.02-0.6%, P</=0.001% and Fe balance. (2) an analogue of M17(IN 100) Ni-base superalloy with low segregation. The characteristic specificity is to eliminate Zr and control P below 0.0005%. The composition is: Cr 8.0-10.0%, Co 8.0-10.0%, W 2.0-5.0%, Mo 1.0-3.0%, Ta 2.0-5.0%, Al 3.5-5.5%, Ti 3.5-5.0%, C 0.1-0.22%, P</=0.0005%, B 0.003-0.010%, Si</=0.05%, Fe</=0.3% and Ni balance. (3) an analogue of M40 (IN 792) Ni-base superalloy with low segregation. The characteristic specificities are (A) to eliminate Zr and reduced P below 0.0005%, and (B) to raise the Al and Ti contents by 1(%). The composition is: Cr 11.0-14.0%, Co 8.0-10.0%, W 2.0-5.0%, Mo 1.0-3.0%, Ta 2.0-5.0%, Al 3.8-4.5%, Ti 4.2-5.4%, C 0.1-0.22%, P</=0.0005%, B 0.003-0.010%, Si</=0.05%, Fe</0.3% and Ni balance. (4) an analogue of M38 (IN 738) Ni-base superalloys with low segregation and high Al and Ti contents. The characteristics specificities are (A) to eliminate Zr and control P below 0.0005%, (B) to raise the Al and Ti contents by about 1(%). The composition is: Cr 15.7-16.3%, Co 8.0-9.0%, W 2.4-2.8%, Mo 1.5-2.0%, Nb 0.6-1.1%, Ta 1.5-2.0%, Al 3.7-5.0%, Ti 3.7-5.0%, C 0.1-0.22%, P</=0.0005%, B 0.005-0.010%, Fe</=0.3%, Si</=0.05% and Ni balance. (5) an analogue of M 38(IN 738) Ni-base superalloy with low segregation and high Cr content. The characteristic specificities are (A) to eliminate Zr and control P content below 0.0005% and (B) to increase the Cr content by about 4(%). The composition is: Cr 18.0-22.0%, Co 7.5-9.5%, W 1.6-4.0%, Mo 2.5-3.0%, Nb 0.3-1.5%, Al 3.7-5.0%, Ti 3.5-5.0%, C 0.08-0.22%, P</=0.0005%, B 0.003-0.010%, Si</=0.05%, Fe</=0.3% and Ni balance.

It is well known that during the recent two decades the air-cooling technique in cast superalloys has been widely used in many countries and some new techniques, such as directional solidification which can raise the service temperature of cast superalloys by about 20°C, have also been developed. However, the directional solidification technique not only makes the air-cooling difficult to be used, but also results in poorer properties along the transverse direction. To overcome this shortage very expensive metal Hf has to be added in the alloy. The new method based on this invention makes it possible that the transverse properties of the directionally solidified superalloys can be improved obviously without any addition of Hf. So, the invention benefits to the application of the directional solidification technology and makes it easy to combine with the air-cooling technique. By using both of the techniques, the service temperature of the exisiting cast superalloys can be raised by 100-300°C. Meanwhile, the invented new method is also very helpful to solve the difficult problem in hog working processing of wrought superatlloys and greatly improve the hot workability of the alloys.

The new method of the present invention is carried out as follows: Firstly, the P and Si contents of raw materials (Fe and Cr) must be strictly controlled. If the P and Si contents in Fe and/or Ce are a little high, it is necessary that to refine Fe into powder form by wet metallurgy and to purify Cr several times by electrolysis.

It is needed that the P content in raw materials of Fe and Cr is below 0.0005%, and Si content is below 0.05%. Secondly, the qualified raw materials are melted in a vacuum induction furnace and the added boron content is controlled at the range from about 0.001 to about 0.015%. Finally, when casting the superalloy blades, the alloy is remelted in the vacuum induction furnace with the same melting technology to the ordinary one.

<u>Preferred Embodiments</u>

Example I

I-1. A superalloy (our name: M17E) with low segregation for the cast was produced by melting a composition of about 0.08 to about 0.21% C, about 8.0 to about 10.0% Cr, about 10.0 to about 16.0% Co, about 2.5 to about 3.5% Mo, about 5.5 to about 6.5% Al, about 4.5 to about 5.5% Ti, about 0.005 to about 0.015% B, and the balance of Ni, controlling incidental impurities, such as: <0.0005% P, <0.05% Si, in a vacuum induction furnace. The melted alloy was cast into ceramic molds to form sticks. And cuting down the ends of the sticks, then remelting the sticks in a vacuum induction furnace. The melted alloy was cast into shaped molds to produce parts or specimens.

I-2. A superalloy (our name: M17F) with low segregation for the cast was produced by melting a composition of about 0.08 to about 0.21% C, about 8.0 to about 10.0% Cr, about 8.0 to about 10.0% Co, about 3.0 to about 5.0% W, about 1.0 to about 3.5% Mo, about 3.0 to about 5.0% Ta, about 4.5 to about 5.5% Al, about 4.0 to about 5.0% Ti, about 0.005 to about 0.015% B, and the balance of Ni, controlling incidental impurities,such as: ≦0.0005% P, ≦0.05% Si, in a vacuum induction furnace. The melted alloy was cast into ceramic molds to form sticks. And cuting down the ends of the sticks, then remelting the sticks in a vacuum induction furnace. The melted alloy was cast into shaped molds to produce parts or specimans.

I-3. A superalloy (our name: M40) with low segregation for the cast was produced by melting a composition of about 0.08 to about 0.21% C, about 12 to about 14% Cr, about 8 to about 10% Co, about 1.0 to about 3% Mo, about 3.2 to about 4.3% Al, about 4.2 to about 5.3% Ti, about 0.005 to about 0.015% B, about 3 to about 5% W, about 3 to about 5% Ta and the balance of Ni, controlling incidental impurities, such as: ≦0.0005% P, ≦0.05% Si, in a vacuum induction furnace. The melted alloy was cast into ceramic molds to form sticks. And cuting down the ends of the sticks, then remelting the sticks in a vacuum induction furnace. The melted alloy was cast into shaped molds to produce parts or specimans.

I-4. A superalloy (our name: M38G) with low segregation for the cast was produced by melting a compositon of about 0.08 to about 0.21% C, about 15 to about 17% Cr, about 7 to about 9% Co, about 2 to about 4% W, about 1 to about 3% Mo, about 1 to about 2.5% Ta, about 3.5 to about 4.5% Al, about 3.3 to about 4.3% Ti, about 0.005 to about 0.015% B,about 0.5 to about 1.5% Nb and the balance of Ni, controlling incidental impurities,such as: ≦0.0005% P, ≦0.05% Si, in a vacuum induction furnace. The melted alloy was cast into ceramic molds to form sticks. And cuting down the ends of the sticks, then remelting the sticks in a vacuum induction furnace. The melted alloy was cast into shaped molds to produce parts or specimans.

I-5. A superalloy (our name: M36) with low segregation for the cast was produced by melting a composition of about 0.08 to about 0.21% C, about 19 to about 21% Cr, about 7 to about 9% Co, about 2 to about 4% W, about 1 to about 3% Mo, about 3.5 to about 4.5% Al, about 3.5 to about 4.5% Ti, about 0.005 to about 0.015% B,about 0.5 to about 1.5% Nb and the balance of Ni, controlling incidental impurities,such as: ≦0.0005% P, ≦0.05% Si, in a vacuum induction furnace. The melted alloy was cast into ceramic molds to form sticks. And cuting down the ends of the sticks, then remelting the sticks in a vacuum induction furnace. The melted alloy was cast into shaped molds to produce parts or specimans.

The most preferred version of the superalloys of the present invention were produced by following compositions of elements in Table I-A.

The conventional superalloys for the cast containing essentially about 0.004 to about 0.009% P, about 0.03 to about 0.15% Zr, about 0.005 to about 0.02% B, about 0.05 to about 0.3% Si. Because of the segregation, the microstructure of the conventional superalloys are found to have stable laves phase on solidification, to form 3-5% (gamma+gamma$'$) phase, and to produce delta phase after exposure in the elevated temperatures.

The suprealloys of the present invention for the cast containing about ≦0.0005% P, eliminating Zr, strictly controlling the contents of B and Si, reducing the segregation, and the microstructures of the said superalloys are found not to have stable laves on solidification.

Specimens of the said superalloys of Example I were evaluated to determine their mechanical properties at different temperatures.The results are:

TABLE I-B shows 100-hours stress Rupture strength (MPa) at different temperature and Hot corrosion resistance.

TABLE I-C shows stress Rupture time (hr)

As may be noted from Table I-B, Table I-C, the nickel-base superalloys in present invention showed improved elevated temperature strength properties over their proir art and these properties were even further improved by use of the preferred heat treatment.

Example II

II-1. A superalloy (our name :LSDS738) with low segregatin for the directional solidification was produced by melting a composition of about 0.08% C, about 16% Cr, about 8.5% Co, about 2.6% W, about 1.7% Mo, about 0.7% Nb, about 1.7% Ta, about 3.5% Al, about 3.3% Ti, about 0.008% B, Hf-free and the balance of Ni, controlling incidental impurities, such as: ≦0.0005% P, ≦0.05% Si, in a vacuum induction furnace. The melted alloy was cast into ceramic molds to form sticks. And cutting down the ends of the sticks, then remelting the sticks in a vacuum induction furnace. The melted alloy was cast into shaped molds to produce parts or specimans.

The conventional superaloys for the directional solidification containing essentially Hf and about 0.08% C, about 0.012% B, about 0.10% Zr, about 0.005% P, about 0.1% Si.

For improving the transverse properties of the directional solidified superalloys, the very expensive metal Hf have to be added in the said alloy.

The present invention make it possible that the transverse properties of the directional solidified superalloys can be improved obviously without any addition of Hf. So, the invention benefits to the application of the directional solidification technology and makes it easy to combine with the air-cooling technique.

Specimens of the said superalloys were evaluated to determine their mechanical properties at different temperatures. The results are:

TABLE II-A shows Stress Rupture Time (hr)

TABLE II-B shows 100-hour stress Rupture strength (MPa) at different temperature.

As is evident, the directional solidified superalloys for the cast of the present invention exhibit substantially increased elevated temperature strength and stress Rupture Time, especially increased the transverse properties of the directional solidified suporalloys.

Example III

III-1: A superalloy (our name: LSIN901) with low segregation for the wrought was produced by melting a composition of 0.03% C, 12% Cr, 5.7% Mo, 0.2% Al, 3% Ti, 43% Ni, 0.003% B, and the balance of Fe, controlling incidental impurities, such as: ≦0.0005% P, ≦0.05% Si, in a vacuum inductiong furnace. The melted alloy was cast into ceramic molds to form slabs. The slabs were remelted by vacuum drip melting to form ingot.

III-2. A superalloys (our name:LSA286) with low segregation for the wrought was produced by melting a compositiong of about 0.08% C, about 13.5 to about 16.0% Cr, about 24% to about 27% Ni, about 1.0% to about 1.5% Mo, about 1.9% to about 2.35% Ti, ≦0.35% Al, about 0.1% to about 0.5% V, about 0.001% to about 0.004% B, and the balance of Fe, controlling incidental impurities, such as:≦0.001% P, ≦0.05% S, in a vacuum induction furnace the malted alloy was cast into ceramic molds to form slabs. The slabs were remelted by vacuum drip melting to form ingot.

Specimens of the two alloys (LSIN901 and IN901) were evaluated to determin their mechanical properties at both room temperature (20°C) and at elevated temperature (650°C). The results are showed at Table III-A.

Specimens of the two alloys (LSA286 and A286) were evaluated to determine their mechanical properties their mechanical properties at different temperatures (20°C). The results are showed at Table III-B(1)

TABLE III-B(2) shows stress Rupture Time (hr) of superalloys (LSA286,A286) in stick shape.

TABLE III-B(3) shows the mechanical properties of superalloys (LSA286, A286) in flat shape.

Comparison of the mechanical properties of the superalloys of the present invention (LSIN718) with the prior art (IN718).The results are showed at Table III-C.

As may be noted from Table III-A, Table III-B, III-C, the superalloys of the present invention (LSIN901, LSA286, LSIN718) showed improved elevated temperature strength properties over the prior art and these properties were even further improved by the use of the preferred heat treatment.

The new method of the present invention is suitable to both Fe-Ni-Cr-base superalloys and Ni-base superalloys. It is obvious that the characteristics of the low segregation superaloys of the present invention lie in: (1) no addition of Zr; (2) P content is below 0.001 or 0.0005%; (3) contents of Al and Ti increased by 1%; (4) Cr content increased by 4%. The properties of the superalloys provided by this invention are shown in Table IV, V,VI and VII.

Table IV shows the comparison of the properties of GH901 (IN901) alloy with those of the low segregation GH90; alloy. It can be seen that the tensile strength at room temperature, stress repture property and incipient temperature of the low segregation GH901 alloy are raised obviously. Table V shows the comparison of 100 hour stress repture strength data in the M38 (IN738) and M17(IN100) alloys with those in the corresponding low segregation ones. The properties of the low segregation alloys are obvously higher than those in conventional ones. Table VI gives the comparison of the stress rupture life (hour) and hot corrosion resistance in both series of the alloys (M38 & M36 with low segregation M38 & M36 alloys). Table VII lists the data of the stress reputre strength at elevated temperatures in both IN 792 and low segregation IN 792 alloys. It is clear that the properties of low segregation superalloys are obviously higher than those of the corresponding conventional ones.

By using the new method provided by this invention, it is quite possible to develop a series of new Fe-base and Ni-base superalloys with excellent properties. It is more important that the properties of the superalloys can be greatly improved by this new method without varying their composition range of the superalloys which are widely used and have stable properties. As well known, it takes very long time to develop kind of new superalloy since it can not be put into production until the stable property data is obtained by long-term and systematic tests, such as stress rupture and creep property tests Therefore, it is of significant importance that to improve obviously the properties of the superalloys only by means of controlling the content of one or two elements without varying their composition ranges. In summary, the invention provides a new technique which is easy to perform and has fantastic benefits to the improvement of the properties of superalloys. Its wide application in indusry will certainly bring about huge economic benefit.

TABLE I-A

ALLOYS COMPOSITION (w%)													
Superalloys	Ni	C	Cr	Co	W	Mo	Nb	Ta	Al	Ti	B	P	Si
M17E	balance	0.17	9.0	15.0	-	3.0	-	-	6.0	5.0	0.008	<0.0005	<0.05
M17F	balance	0.17	9.0	9.0	3.9	2.0	-	3.9	5.0	4.5	0.008	<0.0005	<0.05
M40G	balance	0.18	12.7	9.0	3.9	2.0	-	3.9	3.8	4.8	0.008	<0.0005	<0.05
M38G	balance	0.17	16.0	8.5	2.6	1.7	0.7	1.7	4.0	3.8	0.008	<0.0005	<0.05

TABLEI-B

100-hours Stress Rupture Strength (MPa) at different temperature and Hot Corrosion Resistance							
Superalloys	750 °C	800 °C	850	900 °C	950 °C	1000 °C	Hot Corrosion resistance
IN100	686	540	421	312	206	147	bad
M17E	706	569	466	333	235	157	bad
M17F	745	608	480	349	255	176	bad
IN792	686	540	421	312	206	142	good
M40	696	573	451	343	245	162	good
N738	598	451	363	255	176	118	good
M38G	666	529	402	299	206	135	good
M36	627	500	363	260	186	123	excelent

TABLE I-C

Stress Rupture Time (hr)			
Superaeloys	815°C		932 °C
	51 (Kg/mm²)	39 (Kg/mm²)	155 (Kg/mm²)
IN792	100	1000	100
M40	300	2500	300

TABLE II-A

Stress Rupture Time (hr)				
		815°C	43Kg/mm²	
alloys	IN738	DS 738	LSDS738	
		Vertical	Vertical	transverse
Time (hr)	100~120	80~100	300	200

TABLE II-B

100-hour Stress Rupture Strength (mpa) at different T.						
alloys	750°C	800°C	850°C	900°C	950°C	1000°C
IN738	598	451	363	255	176	118
LSM38G	666	529	402	299	206	135
LSDS 38G Vertical	710	570	456	335	235	142
Transverse	670	540	430			

TABLE III-A

Alloys	20°C				650°C/617MPa	
	σb (mpa)	$\sigma 0.2$ (mpa)	δ (%)	ψ (%)	Stress Rupture Time (hr)	
					smotr	with gap
LSIN901	1340	1000	22	27	131	617
IN901	1226	821	19	20	64	63

TABLE III-B (1)

Alloys	20° C			
	σb (mpa)	σ0.2 (mpa)	δ (%)	ψ (%)
LSA286 stick	1150	803	24	43
flat	1049	706	29	54
A286 stick	892	676	28	43
flat	891	675	28	44
Alloys	550° C			
	σb (mpa)	σ0.2 (mpa)	δ (%)	ψ (%)
LSA286 stick	933	686	20	50
flat				
A286 stick	853	633	22	50
flat				
Alloys	550° C/666 (MPa)	650° C451 (MPa)		
Stress Rupture Time (hr)				
	smoth	with gap	smoth	with gap
LSA286 stick	6545	>984		
flat			243	332
A286 stick	66	84		
flat			64	67

TABLE III-B (2)

Stress Rupture Time (hr) of Superalloys (LSA286, A286) in stick shape							
Alloys	Begining T. of wrought (° C)	550° C, 68Kg/mm² Time (hr)		650° C, 46Kg/mm² Time (hr)		700 ° C, 32Kg/ mm² Time (hr)	
		smoth	with gap	smoth	with gap	smoth	with gap
LSA286	1240	652	>984	475	>650	304	
		659	>985	287			
A286	1120	53		64			
		67		67		68.7	

TABLE III-B (3)

The mechanical properties of superalloys (LSA286, A286) in flat shape					
Alloys	Begining T. of Wrought (°C)	20°C			
		σb (mpa)	σ0.2 (mpa)	δ (‰)	ψ (‰)
LSA286	1180	1068	770	24.6	43.6
(flat)		1070	789	25.4	42.8
LSA286	1220	1053	761	28.2	50.8
(flat)		1055	807	24.7	47.2
A286	1120	1066	707	24.4	56.8
(flat)					
Alloys	Begining T. of Wrought (°C)	650°C			
		σb (mpa)	σ0.2 (mpa)	δ (‰)	ψ (‰)
LSA286	1180	832	669	27.2	50.9
(flat)		837	634	27.1	52.6
LSA286	1220	837	687	24.6	47.6
(flatwt)		835	699	26.6	45.6
A286	1120				
(flat)					
Alloys	Begining T. of Wrought (°C)	650°C, 40Kg/mm^2		Crystal size	
		τ (hr)	δ/ψ (%)		
LSA286	1180	2007	15.2/25.5	4~6	
(flat)		1335	13.1/19.6		
LSA286	1220	1971	8.6/21.9	4~6	
(flat)		2078	8.3/22.8		
A286	1120	187	7.6/15.7	5~8	
(flat)		261	8.3/13.1		

TABLE III-C

Alloys	20 °C				650 °C			
	σb (mpa)	σ0.2 (mpa)	δ (%)	ψ (%)	σb (mpa)	σ0.2 (mpa)	δ (%)	ψ (%)
IN718	1553	1393	13.5	22.3	1233	1083	28.4	56.5
LSIN718	1517	1325	14.2	21.7	1232	1087	28.2	55.6
Alloys	650 °C Stress Rupture Time (hr)							
	Stress (mpa)	Stress rupture Time (hr)	δ (%)	stress rupture Time (hr) (the specimens With gap)				
IN718	686	37.3	38.8	231				
		53.6	30.8	450				
		62.5	28.3	421				
		64.0	32.0	493				
LSIN718	700	114.2	16.8	168				
		111.3	16.8	>261				
		68.7	27.6	>261				

Table IV

The Properties of The Alloys (Tensile properties at room temperature)							
Alloys	σ0.2 (kg/mm^2)	σb (kg/mm^2)	δ (%)	ψ (%)	α_K (kg/mm^2)	650 °C, 63Kg/mm² time (hours)	Incipient MeltingPoint (°C)
GH901	120/130	84/94	17/21	18/22	5.5/5.6	56/72	1140
LSGH901	134/136	100/101	16/18	31/32	7.5/8.0	109/136	1260

Table V

100hour Stress Rupture Strength (Kg/mm^2) at Different Temperature						
Alloys	700 °C	800 °C	850 °C	900 °C	950 °C	1000 °C
M38	61	46	37	26	18	12
LSM38	68	54	41	30.5	21	13.8
M17(In100)	70	56	43	32	21	15
LSM17F	76	62	49	36	26	18

Table VI

Stress Rupture Time(h) and Hot Crrosion Resistance				
Alloys	900C, 26Kg/mm^2	850°C, 37Kg/mm^2	750°C, 60Kg/mm^2	hot corrosion resistance
M38	100	100	100	good
LSM38	300	300	400	good
(M38G)				
LSM36		77	155/176	excellent

Table VII

Stress Rupture Time(h)			
Alloys	815°C		932°C
	51 (kg/mm^2)	39(kg/mm^2)	15.5 (kg/mm^2)
IN-792	100	1000	100
LSIN-792(M40)	300	2500	300
LS: low segregation.			

Claims

1. An alloy which comprises nickel, up to about 0.001% phosphorus up to about 0.05% silicon, about 0.001 to about 0.015% boron, no zirconium.

2. The alloy of claim 1 wherein the balance of the alloy comprises one or more elements selected from the group comprising of carbon, iron, chromium, cobact, tungsten, molybdenum, titanium niobium, aluminum and tantalum.

3. An alloy which consists essentially of about 0.02 to about 0.22% carbon, about 6 to about 22% chromium, up to about 8% molybdenum, up to about 4% niobium, up to about 12% tantalum, about 2 to about 6.5% aluminum, about 2 to about 5.5% titanium, about 0.003 to about 0.015% boron, up to about 0.0005% phosphorus, up to about 0.05% silicon zirconium-free and the balance nickel.

4. The alloy of claim 3 wherein if the alloy is to be directional solidified,carbon is about 0.05 to about 0.13% and no hafnium is added, if the alloy is to be casted, carbon is about 0.17% to about 0.18%.

5. An alloy which consists essentially of about 0.08 to 0.21% carbon, about 8 to about 10% chromium, about 10 to about 16% cobalt, about 2.5 to about 3.5% molybdenum, about 5.5 to about 6.5% aluminum, about 4.5 to about 5.5% titanium, about 0.005 to about 0.015 boron, up to about 0.0005% phosphorus, up to about 0.05% silicon, zirconium-free and the balance nickel.

6. An alloy which consists essentially of about 0.08 to about 0.21% carbon, about 8 to about 10% chromium, about 8 to about 10% cobalt, about 3 to about 5% tungsten, about 1 to about 3% molybdenum, about 3 to about 5% tantalum, about 4.5% to about 5.5% aluminum, about 4 to about 5% titanium, about 0.005 to about 0.015% boron, phosphorus</=0.0005%, silicon</=0.05%, zirconium-free and the balance nickel.

7. An alloy which consists essentially of about 0.08 to about 0.21% carbon, about 12 to about 14% chromium, about 8 to about 10% cobalt, about 3 to about 5% tungsten, about 1 to about 3% molybdenum, about 3 to about 5% tantalum, about 3.2 to about 4.3% aluminum, about 4.2 to about 5.3% titanium,about 0.005 to about 0.015% boron, phosphorus</=0,0005% silicon</=0.05%, zirconium-free and the balance

nickel.

8. An alloy which consists essentially of about 0.08 to about 0.21% carbon, about 15 to about 17% chromium, about 7 to about 9% cobalt, about 2 to about 4% tungsten, about 1 to about 3% molybdenum, about 0.5 to about 1.5% niobium, about 1 to about 2.5% tantalum, about 3.5 to about 4.5% aluminum, about 3.3 to about 4.3% titanium, about 0.005 to about 0.015% boron, phosphorus</=0.0005%, silican< =0.05%, zirconum-free and the balance nickel.

9. An alloy which consists essentially of about 0.08 to about 0.21% carbon, about 19 to about 21% chromium, about 7 to about 9% cobalt, about 2 to about 4% tungsten, about 1 to about 3% molybdenum, about 0.5 to about 1.5% niobium, about 3.5 to about 4.5% aluminum, about 3.5 to about 4.5% titanium, about 0.005 to about 0.015% boron, phosphorus</=0.0005%, silicon</=0.05%m, zirconium-free and the balance nickel.

10. The alloy of claims 5 to 9 wherein hafnium is free.

11. An alloy which consists essentially of about 0.01 to about 0.05% carbon, about 11 to about 13% chromium, about 5.3 to about 6.3% molybdenum, about 0.1 to about 0.5% aluminum, about 2.5 to about 3.5% titanium, about 40 to about 45% nickel, about 0.001 to about 0.005% boron, phosphorus</=0.001%, silicon</=0.05% and the balance iron.

12. An alloy which consists essentially of about 0.01 to about 0.05% carbon, about 14 to about 16% chromium, about 0.1 to about 0.5% aluminum, about 1.8 to about 3% titanium, about 24 to about 28% nickel, about 0.001 to about 0.005% boron, phosphors</=0.001%, silicon</=0.05% and the balance iron.

13. A method for making an improved superalloy comprising steps of:

(a) controlling phosphorus content in iron and chromium of raw materials, said phosphorus content in iron and chromium is less than 0.0005%,

(b) melting the raw materials with added boron in the range from about 0.001 to about 0.015% and eliminating zirconium.

14. The method of claim 13 wherein silicon content in iron and chromium is less than about 0.05%.

15. An alloy made by the method of claim 13 or 14.