RENE 220: 100°F IMPROVEMENT OVER ALLOY 718

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Abstract

A new precipitation-hardened cast superalloy, designated as Rene 220, has been successfully developed to satisfy the needs for structural component applications in advanced aircraft engines. With the same processing advantages (castability, weldability, workability) as alloy 718, Rene 220 exhibits an enhanced temperature capability 100° F over alloy 718. The alloy was designed to be strengthened by the bct γ'' phase instead of the fcc γ' phase observed in most superalloys. The sluggish kinetics of γ'' precipitation reduced the susceptibility of weld cracking. The precipitation behavior of strengthening phases was determined as a function of hardening element additions. Extensive investigations were performed to study the effects of alloying elements on the high-temperature properties of γ'' -hardened alloys. The Co addition offered a significant improvement on the stress rupture life, and the precipitation-hardening elements controlled the alloy strength at temperatures up to 700° C. In the comparison tests, Rene 220 demonstrated the best combination of tensile, rupture, and welding properties over other γ' -hardened alloys: Waspaloy, Rene 41, and Rene 63.

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Introduction

A good combination of strength and environmental resistance at elevated temperatures allows high-strength superalloys to be extensively employed as the primary material critical core components in current gas turbine engines. Among many commercially available superalloys, alloy 718 stands out as the most dominant alloy in superalloy production. In 1986, this grade achieved 45% of the total production of nickel-base alloy in wrought form, and 25% in cast form [1]. Through extensive field experience in both manufacture and service, many unique characteristics have been identified in alloy 718. High alloy strength and good resistance to low cycle fatigue (LCF) at elevated temperature are the important properties for engine disk applications [2]. From the metal processing point of view, alloy 718 offers good castability, workability, and most of all, weldability. There is a trend that many wrought components will be replaced by structural casting[3]. Because of its excellent resistance to post-weld heat treatment cracking, alloy 718 is called the "work-horse" alloy of large structural casting.

Increasing the maximum operating temperature has been one of the most vital design thrusts since the aero gas turbine has existed. Alloy 718 was developed in the late 1950s [4]. The turbine sections consisting of components made from this alloy have already reached the upper limit of their temperature capability in current gas turbine engines. The need for weldable superalloys, which are capable of operating at higher temperatures above alloy 718, have been forecast in the designing of the advanced aircraft engines.

In the past, many attempts on this subject were pursued, but any success has been extremely limited and constrained. Most alloy development programs adapted the approach of modifying the existing superalloy composition that had a proven temperature advantage over alloy 718. Minor chemistry changes were expected to enhance the alloy's weldability without significant reduction in temperature capability. Rene 41, which shows a definite increment (100°F) in creep rupture capability over alloy 718, was one of those examples. Lowering the level of trace-alloying elements or impurities, such as iron, silicon, manganese, sulfur and carbon, can reduce strainaging cracking in highly restrained Rene 41 weldments [5,6]. However, this type of approach has never been able to develop a superalloy whose weldability matched that of alloy 718.

This paper reports briefly on an alloy design project at GE that adapts a new approach for developing weldable superalloys with improved temperature capability. The alloy system is first selected to satisfy the welding criteria, and then the alloy chemistry is modified to achieve the maximum temperature capability. Using this revolutionary approach has successfully accomplished the design of a new nickel-base superalloy for the application of high-temperature structural components in advanced gas turbine engines. The new alloy, designated as Rene 220, provides the unique combination of weldability and high temperature capability with all good processing characteristics as alloy 718.

Experimental

More than 100 alloy compositions have been studied in this project. Each alloy with a selected chemistry was prepared by vacuum induction melting using high purity raw materials. Two sizes of heats, 5 lb and 40 lb, were used for different evaluation purposes; and the melts were cast into chilled copper molds under a partial argon atmosphere. All ingots were subjected to homogenization or hot isostatic pressing (HIP) processes.

The precipitation behavior of various alloy compositions was investigated by transmission electron microscopy (TEM), which identifies the fine strengthening precipitates. Thin foils were prepared by the jet-polishing technique employing a 20% perchloric acid-methanol solution at -60°F. Some extraction replicas were made and examined under scanning transmission electron microscopy (STEM) to determine the precipitate chemistry by using X-ray spectroscopy. Optical metallography and mechanical tests were performed according to the standard laboratory procedures.

Weldability tests were conducted on 0.225 in. thick slices cut from 3.5 in. diameter ingots in both homogenized and HIP'd conditions. Grooves of 0.75 in. width were machined on both top and bottom faces leaving a 0.060 in. thickness between the grooves on each slice. Such a configuration can restrain the weld path and simulate the practical welding condition. A series of electron beam (EB) and tungsten-arc inert gas (TIG) welds were repeatedly applied on the groove of every alloy. Visual inspections counted the number of welding cracks after each welding and post-weld heat treatment.

Weldability Criteria

The cracking mechanisms that account for welding precipitation hardenable superalloys involve the kinetics of the precipitation reaction directly. The alloy could be effectively strengthened as soon as a certain volume of precipitates forms in any thermal history. When alloy strength is developed, the critical strain that causes cracks becomes sufficiently low. Since the rate of hardening during heat treatment is related in part to the content of precipitation-hardening elements, the tendency for cracking in the γ -strengthened superalloys is commonly reported to be a function of aluminum and titanium contents [7]. Figure 1 summarizes weldability of various high-strength superalloys. The margin line for acceptable weldability is found to be (Al wt% + 1/2 Ti wt%) <3.0. This empirical rule implies that all γ -hardened alloys that have an equivalent weldability as alloy 718 would not develop enough precipitates to achieve a strength comparable to that of alloy 718.

The crystal structure of strengthening precipitates is the key difference between alloy 718 and other precipitation-hardened superalloys. The addition of 5.1 wt% Nb in alloy 718 makes most coherent precipitates that develop during aging treatments form an ordered bct structure, though a small percentage of ordered fcc γ - precipitates can still be detected [8]. The unique precipitation phase, designated as γ ", exhibits a great lattice misfit along its tetragonal axis (c-axis) with the alloy matrix. Large coherency strains associated with the formation of γ " precipitates in nickel-base superalloy impose two significant effects with respect to the welding topic studied here. One effect is that γ "-hardening through coherency strengthening becomes so efficient that only a low volume fraction of precipitation phase is required to attain the desirable strength. The second effect is that the nucleation rate of γ " precipitates is comparably slow because of the additional activation energy from large elastic strain energy. Both effects provide the alloy sufficient advantages in the weldability viewpoint.

The first stage of this work is to identify the alloying parameters that determine the strengthening precipitation phases. Precipitation-hardening elements generally employed in nickel-base superalloys include four alloying elements: Al, Ti, Nb, and Ta. This class of elements has a common feature in the nickel-base alloys; their solubility increases with temperature and reaches to a maximum at the eutectic temperature. The intermetallic phases forming the eutectic with the fcc matrix have a nominal chemistry form of $N_{i3}X$ where X is the hardening element and their crystal lattice consists of ordered close packing planes similar to {111} planes in fcc lattice. During aging treatments, the hardening element atoms will segregate together by diffusion and will form the ordered structure on close packing planes, {111}, of fcc matrix. Depending upon the final equilibrium phase, the ordering pattern on {111} planes is either in triangle (for Al and Ti) or in rectangle (for Nb and Ta). The triangularly ordered fcc precipitate is designated as γ_{γ} ,

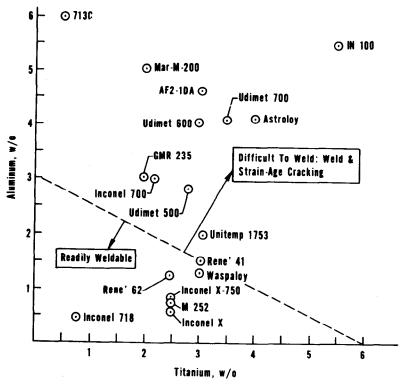


Figure 1 Weldability criteria for precipitate hardened superalloys [7].

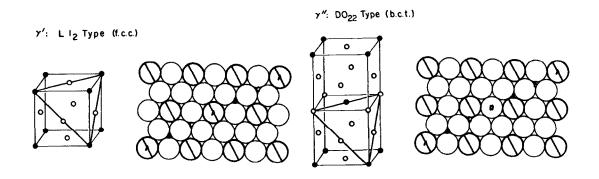


Figure 2 Schematic plots of crystal unit cell and closed packing plane for strengthening precipitate phases in superalloys.

which has a L1₂ crystal lattice. The precipitate with a rectangular fcc ordering is called γ ", having a DO₂₂ crystal structure. Figure 2 shows the unit cell and the closed packing planes of two precipitate crystals.

An alloy series (P1 - P5) was specifically designed to discover the transition of precipitate crystal structure from γ to γ'' . The major alloying constituents including Cr, Co, Mo and the total atomic content of precipitation-hardening elements (Al+Ti+Ta+Nb \simeq 6.6 at %) are maintained at constant. The γ'' -forming elements, Ta and Nb, were added to substituting γ -forming elements Al and Ti with the increment \sim 15% of total hardening elements from alloy to alloy. Table 1 lists the variation of hardening elements in this alloy series.

TEM investigation on the strengthening precipitate phase developed after aging treatments in this alloy series identified both ordered Ni_3X structures, γ , and γ . The precipitate structure and quantity depend upon the percentage of (Ta+Nb) substitution. Low (Ta+Nb) percentage forms γ , which has a spherical-to-cuboidal morphology as seen in alloy P1, P2 and P3 (Figure 3(a)). When sufficient (Ta+Nb) replaces (Al+Ti), such as alloy P4 whose (Ta+Nb) percentage is 44.9%, plate-like γ " precipitates start to appear. Almost equal volume fraction of γ , and γ " particles are observed in alloy P4. Further increases in (Ta+Nb) percentage above 62% can convert all precipitation phases into γ ". The dark field image taken from a superlattice reflection reveals only γ " precipitate existing in alloy P5 (Figure 3(b)).

As mentioned above, γ and γ " are the transition precipitates having an ordered structure determined by the equilibrium phase, which is a function of the (Ta+Nb) percentage in the total hardening element content. Also these two transition precipitates, γ and γ ", can coexist over a range of (Ta+Nb) percentage. Prolonged aging of alloy P4 coarsens both γ and γ " without any indication that one precipitate will replace or consume another (Figure 4). The idea of a two-phase region in a phase diagram can be applied to alloy P4, and the "transition" phase boundaries can be obtained by analyzing γ and γ " precipitate chemistries under analytical STEM. These results are given in Table 2; the method of microanalysis work is reported elsewhere [9].

TABLE 1
Precipitation Hardening Element Content in Alloy Series P

Alloy	C	C_H^*	R**			
Number	Al	Ťi	Ta	Nb	at.%	%
P1	1.5	3.0	-	-	6.71	0.0
P2	1.5	2.0	3.0	-	6.62	14.5
P3	1.0	2.0	6.0	-	6.68	29.6
P4	0.5	2.0	6.0	1.5	6.61	44.9
P5	0.5	1.0	6.0	3.0	6.38	62.4

^{*} C_H is the total hardening element content, (Al+Ti+Ta+Nb), in at.%.

^{**} R is the (Ta+Nb) percentage of C_H; i.e., R = (Ta+Nb)/(Al+Ti+Ta+Nb) by at.%.

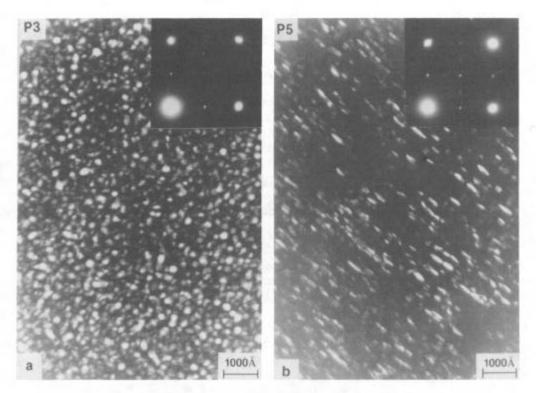


Figure 3 Transmission electron microscopy of strengthening precipitates (a) γ' in alloy P3; (b) γ'' in alloy P5.

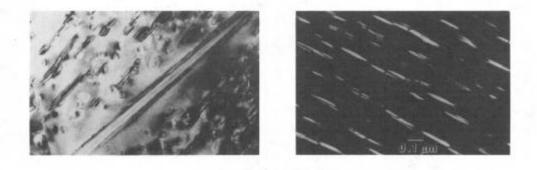


Figure 4 Co-existence of γ' and γ'' precipitates in overaged alloy P4.

TABLE 2 γ and γ " Transition Precipitate Chemistries (at %)

	Ni	Cr	Со	Мо	Al	Ti	Та	Nb	C_H	R
overall	60.0	20.1	12.9	2.4	1.3	2.5	2.1	1.0	6.9	44.9
7.	68.9	1.0	6.7	0.2	6.7	10.4	3.8	2.4	23.3	26.6
γ "	64.0	5.2	5.4	3.3	4.5	5.4	6.1	6.1	22.1	55.2

^{*} See Table 1 for the definitions of \mathbb{C}_H and \mathbb{R} .

The total amount of hardening elements, Al+Ti+Ta+Nb, in the γ - and γ " precipitates is equal to 23.3. and 22.1 at%, respectively. These values approach the stoichiometric composition 25 at%. The critical (Ta+Nb) percentages for γ - and γ " precipitates, which may define the "transition" phase boundaries, are 26.6% and 55.2%, respectively. The accuracy of experimental data is expected to be within 5%, and the results of the present study have an error of no more than 10%.

To assure the strengthening precipitates being γ'' in nickel-base superalloys, the hardening elements should be controlled so that the (Ta+Nb) percentage with respect to the total hardening element content would be above 62.5%. The precipitation behavior might also be influenced by alloying elements other than hardening elements, but the observed effects are secondary.

Temperature Capability Enhancement

A disadvantage of using γ "-strengthening in nickel-base superalloys is the phase stability after prolonged exposure at high temperature. The γ " precipitation phase is not the equilibrium phase, and it will be replaced, when overaged, by the stable δ phase (DO_a type), which can degrade the alloy properties. Very limited literature is available that discusses how to improve the phase stability of γ "-hardened alloys.

Through the exclusive evaluation of alloy series, two ideas were identified as the most effective methods to improve the high-temperature properties of γ "-strengthened superalloys. With respect to the phase stability, an optimum alloying addition of major elements (Cr, Co) can further slow down the precipitation kinetics and can retard the formation of large δ particles. Figures 5 and 6 illustrate the influence of Co and Cr alloying content on tensile and stress rupture properties. Both tensile and stress rupture tests were performed at 704°C; the intial stress level for stress rupture tests was set at 621 MPa.

The Co addition does not affect alloy strength until Co content reaches 20%, and the rupture life shows a maximum value at 12% of Co addition. Cobalt is believed to have a similar chemical behavior as nickel and can substitute for nickel in the precipitation reaction up to a certain amount. Too much Co may reduce the volume fraction of strengthening precipitate because of the shortage in Ni. As a result, both alloy strength and rupture life are decreased. On the other

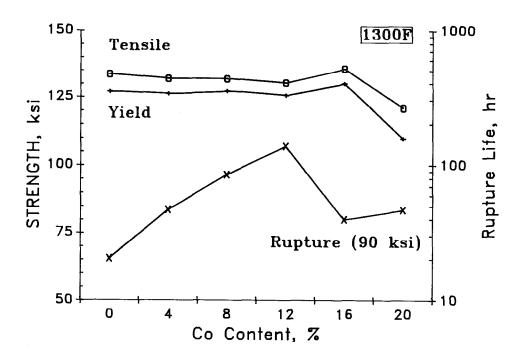


Figure 5 The effect of Co content on high-temperature properties of γ "-hardened superalloys.

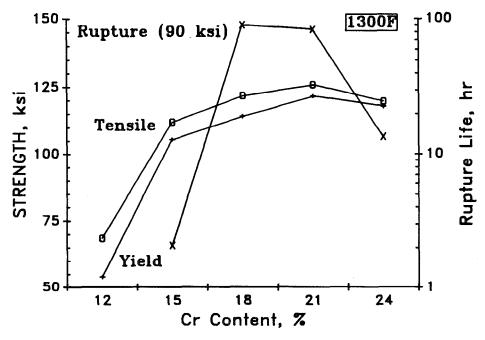


Figure 6 The effect of Cr content on high-temperature properties of γ "-hardened superalloys.

hand, Co alloying is expected to influence the creep rate by slowing down the precipitation kenitics. The effect of Cr addition on the mechanical properties is directly associated with the formation of strengthening precipitates. Increase in Cr content can substantially reduce the solubility of hardening elements at aging temperatures. For a given amount of total hardening elements, the strengthening precipitates will increase with Cr content, which results in the improvement of strength and rupture life. However, the improvement becomes saturated after 18% of Cr addition, and the exceeding chromium seems to degrade alloy properties.

Controlling the precipitate chemistry and volume fraction is a second approach to enhancing high temperature properties. Usually a high precipitate content can benefit alloy strength and rupture life as long as there is no detrimental effect on phase stability. The thermodynamic data suggests that high level Nb can drop the melting point of Ni-base alloys dramatically [10], and melting experience indicates the relationship between Nb and the incipient melting problem of alloy 718 [11]. Theoretically, a high Ta addition is preferred. From the practical point of view, material cost and alloy density limit the Ta content in γ "-strengthened superalloys.

Though the strengthening precipitates are selected for γ'' , the (Ta + Nb) percentage in the total hardening element content can vary from 62.5% to 100%. A series of alloys with a fixed total hardening elements of 7.5 at% was prepared to investigate the effect of (Ta + Nb) percentage on high temperature properties. The measured results are plotted in Figure 7. A high (Ta + Nb) percentage in γ'' precipitates is expected to increase the coherency strains. As a consequence, a high alloy strength is attained while the rupture life becomes shortened. The latter observation is related to the phase stability problem.

Based on sufficient experimental data, a new γ "-hardened superalloy has been developed. The total hardening elements are selected to be about 6.5 at.%, and the (Ta + Nb) percentage is set at 65%. Major alloying additions, including Co and Cr, are chosen at the optimum level individually. The new alloy is designated as Rene 220.

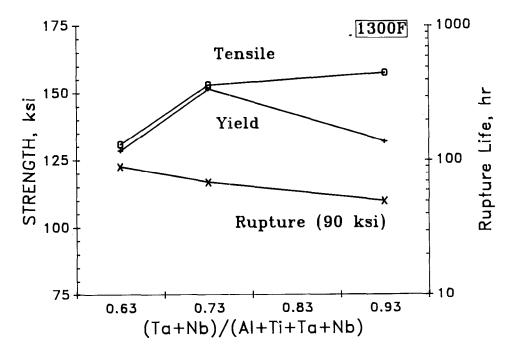


Figure 7 The effect of (Ta+Nb)percentage on high-temperature properties of γ "-hardened superalloys.

Demonstration and Evaluation

In order to illustrate the success of the new approach for alloy design, a series of candidate alloys was prepared for demonstration. This alloy series consisted of Rene 220 and three of the most promising commercial alloys: Waspaloy, Rene 41 and Rene 63. Three commercial compositions were strengthened by γ precipitates; they represented the summary of past efforts following the old approach. As listed in Table 3, some minor modifications from their standard chemistry were made to upgrade their weldability or alloy strength. Rene 220 (S1) is the only alloy in the series strengthened by γ " precipitates. Alloy S2 is a Nb-modified Waspaloy whose lower volume fraction of γ precipitates may provide acceptable weldability. Alloy S3 and Alloy S4 are the low C versions of Rene 41 and Rene 63, respectively. A high content of refractory metal elements in Rene 41 and a large amount of precipitation hardeners in Rene 63 suggest better temperature capability.

Four alloy ingots were processed and evaluated side-by-side. The top half of each cast ingot was homogenized at 2150°F for 4 hours; the bottom half was HIP'd at 2150°F/15 ksi for 2 hours. Based on the solvus temperature determined by the metallography and differential thermal analysis (DTA) techniques, the individual heat treatment was specified for each alloy (Table 4). Tensile properties were evaluated at room temperature (RT) and 1300°F, and stress rupture life was measured at 1300°F/90 ksi. The results obtained from homogenized (H) and hot isostatically pressed (P) specimens are also summarized in Table 4. The newly designed alloy S1 offers the best tensile properties at both RT and 1300°F among four compositions in alloy series S. Yield strength shows at least a 15 ksi difference between alloy S1 and the second strongest alloy (S4). Also alloy S1 exhibits one of the longest rupture lives in this alloy series.

TABLE 3
Chemical Compositions (wt%) of Alloy Series S

		· . — — — — — — — — — — — — — — — — — —		
Alloy				
Number	S1	S2	S3	S4
Ni	Bal.	Bal.	Bal.	Bal.
Cr	18	19	19	14
Co	12	13	11	15
Mo	3.0	4.0	9.8	6.0
Al	0.5	1.0	1.5	3.8
Ti	1.0	2.0	3.2	2.5
Ta	3.0			
Nb	5.0	3.0		
Zr		0.05	0.05	0.05
В	0.01	0.01	0.01	0.01
<u>C</u>	0.015	0.025	0.02	0.02
C_H	6.57	6.42	7.11	10.84
R	64.1	29.1	0.0	0.0
Reference	Rene'220	Waspaloy	Rene'41	Rene'63

^{*} See Table 1 for the definitions of C_H and R.

TABLE 4
Heat Treatments and Mechanical Properties of Alloy Series S

Alloy	<u>S1</u>	S2	S3	<u>S4</u>
Reference	Rene'220	Waspaloy	Rene'41	Rene'63
Heat Treatments	1950°F/1hr + 1400°F/5hr + 1200°F/3hr	1950°F/1hr + 1400°F/5hr + 1200°F/3hr	2050°F/1hr +1600°F/4hr	2150°F/1hr +1600°F/4hr +1400°F/16hr
	HIP'ped	at 2150°F/15ksi f	for 2hr	- · · · · · · · · · · · · · · · · · · ·
room temperature	strength (ksi)			
YIELD	141.8	108.4	97.8	110.2
TENSILE	157.4	147.3	118.8	137.1
1300°F strength (ks	i)			
YIELD	118.7	92.3	88.0	105.5
TENSILE	125.9	104.9	133.0	146.8
1300°F/90ksi ruptu	re			
LIFE (hr)	97.1	27.5	56.1	232.5
	Homoge	nized at 2150°F f	or 4hr	
room temperature	strength (ksi)			
YIELD	145.0	107.9	96.3	113.8
TENSILE	160.9	145.3	114.7	137.5
1300°F strength (ks	si)			
YIELD	116.0	84.7	89.9	102.0
TENSILE	127.4	99.8	117.2	128.4
1300°F/90ksi ruptu	re			
LIFE (hr)	54.7	26.5	95.0	22.8

Wedability tests were performed on both homogenized and HIP'd specimens for each alloy by electron beam (EB) and tungsten-arc inert gas (TIG) welding. The number of cracks developed after each weld pass and after the subsequent post-weld heat treatment was counted visually. Table 5 lists the results of the weldability test. As expected, the new alloy S1 stands out as the best weldable alloy. No cracks were observed until the third welding and heat treating cycle, after which only one crack appeared. The next weldable alloy is alloy S2, but it shows the lowest strength and shortest rupture life. A great number of cracks were detected in alloys S3 and S4 as soon as they were welded.

EB welding usually provides a better weld cracking resistance over TIG welding. The same observation can be found in Table 5, which includes different alloy chemistries and processing conditions. Overall properties of each alloy do not reveal a significant difference between the homogenized and HIP'd conditions except for 1300°F properties of alloy S4.

TABLE 5
Weldability Test Results: Number of Cracks

Alloy	S1	S2	S3	S4	S1	S2	S3	S4	
	EB Welding						elding		
		HIP'pe	d at 2150	°F/15ksi	for 2hr				
1st WELD 1st H.T.	none none	1 1	none none	1 1	none none	4 4	none none	2 10	
2nd WELD 2nd H.T.	none none				none none				
3rd WELD 3rd H.T.	none none				none 2				
	Homogenized at 2150°F for 4hr								
1st WELD 1st H.T.	none none	none none	none none	none none	none none	none none	4 5	9 10	
2nd WELD 2nd H.T.	none none	none none			none none	none none			
3rd WELD 3rd H.T.	none none	none none			1	5 11			

^{*} See Table 4 for the heat treatment (H.T.) of each alloy

A general observation suggests that the γ "-hardened superalloy exhibits a much better weldability than most γ -hardened alloys do. Lowering the hardening element content in γ -hardened alloys could improve the alloys weldability with a payoff in reduction of rupture strength and temperature capability.

Having established the superiority of Rene 220 relative to γ -hardened alloys, a property comparison was made with alloy 718 by processing and testing these two cast alloys in parallel. Both alloy ingots received a HIP'g treatment (2100°F/15 ksi/4 hrs) after casting. Rene 220 samples were solutioned at 1965°F for one hour followed by a double aging treatment at 1380°F for 8 hours plus 1200°F for 10 hours. A standard heat treatment cycle was applied to alloy 718: 1750°F/1 hr + 1325°F/8 hrs + 1150°F/10 hrs. Tensile tests were conducted at 1000°F and 1200°F, while the stress rupture tests were performed at 1200°F and 1300°F. To attain a meaningful rupture life within a reasonable time, Rene 220 was tested at a higher stress level than alloy 718 at both temperatures.

Table 6 lists the measured results of tensile and stress rupture properties for two alloys. Rene 220 demonstrated improved strength over alloy 718 at both testing temperatures. Most significantly, the rupture life of Rene 220 is much longer (at least 4 times) than that of alloy 718, even tested at a high-stress level. The data definitely indicated superior temperature capability of this new γ "-hardened superalloy, Rene 220. Also included in Table 6 are the results measured on some Rene 220 samples that had been exposed at 1300°F for 1000 hours.

TABLE 6
Mechanical Properties of γ "-Hardened Alloys

Tensile Test					
ALLOY	TEST TEMP.(°F)	YIELD (ksi)	TENSILE (ksi)	ELONG. (%)	R.A. (%)
alloy 718	1000	111	121	16	19
	1200	115	118	12	62
Rene'220	1000	126	133	7	60
	1200	135	139	13	43
overaged*	1300	127	129	7	38
Rupture Test				*****	
ALLOY	TEST CONDITION	Ŋ	LIFE (hr)	ELONG. (%)	
alloy 718	1200°F/90ksi		214	6.7	
	1300°F/75ksi		20	5.1	
Rene'220	1200°F/100ks	si	811	runout	
	1300°F/90ksi		118	6.0	
overaged*	1300°F/90ksi		177	4.2	

^{*}overaged at 1300°F for 1000 hr.

Such a long-term exposure at high temperatures can determine the phase stability of an alloy. Alloy 718 has been reported to suffer a significant degradation of mechanical properties after this long-term exposure. In contrast, Rene 220 after 1000-hour exposure at 1300°F showed a somewhat improved rupture life in comparison to that of the unexposed samples. Apparently, the new γ "-hardened superalloy exhibits excellent thermal stability at temperatures up to 1300°F.

Now Rene 220 has been certified for use as a structural casting alloy in advanced aircraft engines. The new alloy offers at least 100°F advantage in temperature capability over 718 alloy as illustrated by Figure 8. The weldability and the castability were demonstrated through many engine hardware trials (Figure 9). The first uses will be in a combustion case and in a turbine frame, a major part of the engine's structural skeleton. Until now, the frame had been made from several parts welded together. But the superior castability and properties of Rene 220 makes it possible to fabricate the frame in a single casting. This new process incorporating the new alloy Rene' 220 sharply reduces the fabrication cost.

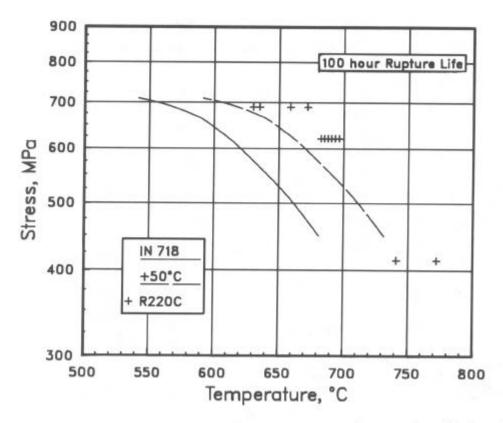


Figure 8 Rupture strength of new casting alloy Rene 220C in comparison with that of IN 718.

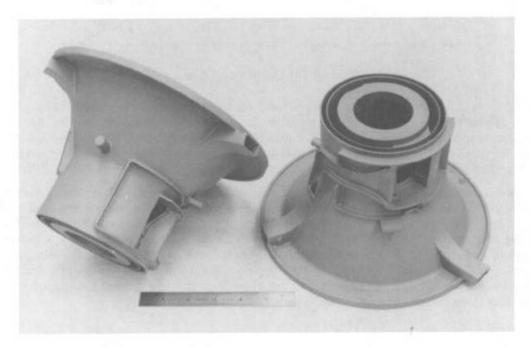


Figure 9 Engine bearing houses made from the investment casting of new alloy Rene 220.

Conclusions

Through an unconventional approach, a successful alloy design project has been accomplished to develop a weldable high-strength cast superalloy with improved temperature capability. The new alloy Rene 220 has superior high-temperature properties over alloy 718 and retains the advantage of metal processing characteristics of alloy 718. The successful alloy design is the result of several evolutions in establishing the knowledge of superalloy metallurgy:

- 1. The formation of strengthening precipitates in nickel-base superalloys consists only in γ'' phase requiring above 62.5% of (Ta+Nb) percentage with respect to total hardening element content (Al+Ti+Ta+Nb) in atomic percentage. Less than 30% of (Ta+Nb) percentage will form γ precipitates. The co-existence of γ and γ'' may occur when the (Ta+Nb) percentage falls between the above two ranges.
- 2. Cobalt addition can remarkably improve the rupture life of γ "-hardened superalloys. The optimum amount is determined to to be 12 wt%.
- 3. The precipitate chemistry can influence both alloy strength and phase stability of γ "-hardened superalloys. A high (Ta+Nb) percentage offers better alloy strengths with the payoff of alloy stability at high temperatures.
- 4. High-strength superalloys strengthened by γ'' precipitates provide superior weldability advantage over those hardened by $\gamma \sim$ precipitates.

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