DEVELOPMENT AND INTRODUCTION OF A DAMAGE TOLERANT HIGH

TEMPERATURE NICKEL-BASE DISK ALLOY, RENE'88DT

D.D. Krueger, R.D. Kissinger, and R.G. Menzies

General Electric Aircraft Engines
1 Neumann Way, Cincinnati, Ohio 45215

To address the need for an improved disk material, General Electric Aircraft Engines initiated an alloy development program which resulted in the production introduction of a powder metallurgy nickel-base disk alloy, Rene'88DT, with a significantly improved balance of creep, damage tolerance and tensile properties. The work done to evaluate experimental alloy compositions in laboratory screening trials and the down-selection to a preferred alloy through scale-up and full scale processing trials is described. In addition, the technology developed to overcome major challenges in grain size control, heat treatability, and achieving the proper balance between property levels and producibility is discussed.

Introduction

Current aircraft gas turbine engines are constrained by disk material capability to operate at compressor discharge temperatures of about 1175°F. To further improve efficiency, this temperature must be raised and disk materials with improved creep capability must be developed. In addition, the introduction of the ENSIP requirements by the USAF in 1983 [1] provided a growing demand for improved fatigue resistance to reduce life cycle costs and increase engine durability. It has thus become desirable to develop disk materials which exhibit greater resistance to crack growth and fatigue failure from inherent or accidental flaws (ie. damage tolerance), and exhibit higher creep capability, while still maintaining adequate tensile strength. It was deemed that powder metallurgy processing, patterned after that for Rene'95 [2], was the only currently viable production method for manufacturing an improved disk alloy. For reasons described later, it was also decided that the forgings would have to be heat treated above the γ' solvus.

The program goals were based on the capability of Rene'95 [2] and prescribed as: 50 percent reduction in cyclic fatigue crack growth rate (RT-1200°F); 25°F minimum improvement in creep and stress-rupture strength; 10 percent maximum reduction in tensile strength. In addition, a minimum effect of peak load hold time on elevated temperature crack growth rates was desired.

The four year program involved four phases. 1) Laboratory scale alloy development based on 1" diameter extrusions to screen experimental alloy compositions. 2) Sub-scale screening trials based on 3" diameter extrusions to evaluate the most promising compositions and develop process parameters. 3) Full scale trials based on 9" diameter extrusions to select the preferred alloy W2 (Rene'88DT). 4) Production introduction of Rene'88DT disk hardware of three different configurations for engine test and certification.

Laboratory Scale Alloy Development Trials

A total of 23 experimental compositions grouped into two series were selected, processed, and evaluated to identify candidate alloys for scale-up trials. The evaluation included tensile, creep, rupture, and fatigue crack growth mechanical property tests. Composition design and heat treatment were based on work done to study the effects of precipitate size and grain size on the fatigue crack growth behavior of γ' strengthened alloys [3-6]. It was generally found that reducing the γ' size promoted dislocation cutting of the precipitates and resulted in planar slip (heterogeneous deformation), while increasing the γ' size promoted dislocation looping and caused wavy slip (homogeneous deformation). For the case of fine γ' and planar slip, fatigue crack growth rates (FCGR) were reduced. This has been attributed to easy dislocation reversal and thus better accommodation of fatigue damage [7]. An increase in grain size has also been found to reduce FCGR. It was postulated that longer slip bands for the coarse grain microstructure could accommodate more dislocation reversals which would reduce the rate of damage accumulation and crack propagation [8]. As a result, most of the experimental alloys selected for study were designed with a γ' content of 35 to 50 percent, and all were solution treated above the γ' solvus prior to aging and evaluation. The coarser grain size developed as a result of supersolvus solution treatment was also expected to benefit creep properties. The focus on low to moderate γ' fraction was expected to minimize solution treatment quench crack risks.

Experimental Alloys

Table 1 lists the baseline and experimental alloy compositions selected for study. The emphasis in Series 1 was on γ' composition effects. In Alloy R1, a small addition of Hf has been added to Rene'95. At the 0.2 w/o level, this addition could provide a strength improvement without causing HfO₂ inclusion problems. Alloys R2 and R3 also contain the Hf addition and form a series with Alloy R1 where the γ' fraction is reduced from approximately 50 to 45 to 40 a/o, respectively. The Al, Ti and Nb levels of Alloys R2 and R3 were selected to maintain the same γ' composition as Alloy R1.

Table I. Aim Compositions of Experimental Alloys and Base Alloy Rene'95 (weight percent)

Alloy	Co	Cr	Mo	W	Al	Ti	Ta	Nb	V	Hf	B	C	Zr	Ni	Predicted γ Fraction (a/o)	Predicted Density (lb./in. ³)
<u>Series 1</u>																
Rene'95	8	13	3.5	3.5	3.5	2.5	--	3.5	--	--	0.015	0.03	0.03	bal	50	0.298
R1	8	13	3.5	3.5	3.5	2.5	--	3.5	--	0.2	"	"	"	"	50	0.299
R2	8	13	3.5	3.5	3.1	2.2	--	3.1	--	0.2	"	"	"	"	45	0.301
R3	8	13	3.5	3.5	2.7	1.9	--	2.7	--	0.2	"	"	"	"	40	0.304
R4	8	13	3.5	2.0	3.1	2.2	1.4	3.1	--	0.2	"	"	"	"	46	0.300
R5	8	13	3.5	3.5	4.5	3.2	--	--	--	--	"	"	"	"	50	0.293
ET1	10	15	2	4	2	4.5	--	--	--	--	"	"	"	"	40	0.300
ET2	10	15	2	4	1.75	5	--	--	--	--	"	"	"	"	40	0.300
ET3	10	15	2	4	1.5	5.5	--	--	--	--	"	"	"	"	41	0.300
W1	13	16	--	7.5	1.7	3.4	--	0.7	--	--	"	"	"	"	37	0.307
W2	13	16	4	4	2.1	3.7	--	0.7	--	--	"	"	"	"	40	0.302
W3	13	16	4	4	1.7	3.4	1.4	--	--	--	"	"	"	"	37	0.307
<u>Series 2</u>																
R6	12	13	3.5	3.5	3.5	2.5	--	3.5	--	--	0.015	0.03	0.03	bal.	49	0.298
R7	16	13	3.5	3.5	3.5	2.5	--	3.5	--	--	"	"	"	"	51	0.298
R8	12	15	3.5	3.5	3.5	2.5	--	3.5	--	--	"	"	"	"	51	0.297
R9	12	13	3.5	3.5	3.1	2.2	--	3.1	--	--	"	"	"	"	45	0.301
R10	16	13	3.5	3.5	3.1	2.2	--	3.1	--	--	"	"	"	"	46	0.301
R11	8	15	3.5	3.5	3.1	2.2	--	3.1	--	--	"	"	"	"	45	0.299
R12	12	17	3.5	3.5	3.1	2.2	--	3.1	--	--	"	"	"	"	46	0.298
R13	16	15	3.5	3.5	3.1	2.2	--	3.1	--	--	"	"	"	"	47	0.299
R14	16	17	3.5	3.5	2.7	1.9	--	2.7	--	--	"	"	"	"	42	0.301
I100	15	10	3	--	5.5	4.7	--	--	1	--	"	"	"	"	63	0.280
R100	15	10	3	--	5.5	4.2	--	--	1	--	"	"	"	"	61	0.281
R97	11.5	11.5	3.25	1.75	4.5	3.4	--	1.75	0.5	0.2	"	"	"	"	56	0.290

Alloy R4 is a derivative of Alloy R2 designed to study a Ta for W substitution. In Alloy R5, the Nb in the γ' has been replaced by Al and Ti. The levels of Al and Ti were adjusted to produce the γ' fraction and Al/Ti ratio present in Rene'95.

Alloys ET2 and ET3 were derived from an experimental base alloy, ET1. In Alloys ET2 and ET3, the Al/Ti ratio of Alloy ET1 (0.8) has been reduced to 0.6 (Alloy ET2) and 0.5 (Alloy ET3) causing the amount of Ti in the γ' to increase from 48 to 53 to 58 a/o, respectively. The amount of Ti in the γ' of Alloy ET3 is very near the expected solubility limit and it was anticipated that this alloy would be prone to Ni₃Ti (eta) formation.

The remaining three Series 1 compositions were based on alloy W1, an alloy similar to that developed in a DARPA sponsored contract [9]. Alloy W2 investigates a Mo for W substitution and an adjustment in the Ti/Al ratio; in Alloy W3, Mo for W, and Ta for Nb, substitutions are investigated.

Most of the Series 2 alloys were designed to study γ composition effects. Alloys R6 through R14 form a matrix of Co and/or Cr substitutions for nickel using Rene'95, a 45 a/o γ' version of Rene'95 (Alloy R2 without Hf), and a 40 a/o γ' version of Rene'95 (Alloy R3 without Hf) as base compositions. Series 2 also included three alloys to evaluate the effects of high γ' fraction; IN100, designated as I100 (66 a/o γ'), R100, a reduced Ti version of I100 (62 a/o γ'), and R97, an experimental alloy (54 a/o γ').

Alloy Processing

Vacuum induction melted ingots weighing ~25 lbs. were prepared and argon atomized for each alloy. Powders from the -140 mesh yield were loaded into stainless steel cans and consolidated by compaction and extrusion at a ratio of 7:1. Compaction and extrusion

Figure 1: Rene'95 grain structure for (a) supersolvus and (b) subsolvus solution treatment; gamma prime distribution for (c) low fraction alloy W2 and (d) high fraction alloy I100.

temperatures were based on an estimated γ' solvus temperature (T_s) for each alloy. Compaction was conducted at $T_s - 150^\circ\text{F}$ and extrusion was conducted at $T_s - 100^\circ\text{F}$. To minimize processing and microstructural variables among the compositions, the solution temperature for each alloy was typically $T_s + 25^\circ\text{F}$. In this case, metallographic studies were performed on each extrusion to determine the solvus temperature. Each 1" diameter alloy extrusion was solution treated for 1 hour. The average cooling rate was controlled at approximately $250^\circ\text{F}/\text{min.}$, which is representative of that achievable in thick disk sections. Two Rene'95 extrusions were processed along with the Series 1 and 2 compositions as baselines. One of the Rene'95 extrusions was subsolvus solution treated at 2025°F and the other supersolvus solution treated at 2150°F . The age treatment for all alloys was $1400^\circ\text{F}/8\text{hr}$, apart from the ET-X alloys, which were aged at $1300^\circ\text{F}/24\text{ hr}$ to reduce the likelihood of eta formation.

Microstructure

The range in grain size for the alloys after supersolvus solution treatment was approximately 15 to $35\mu\text{m}$. In most cases the intragranular γ' , formed during cooling, was uniform in size and ranged from 0.2 to $0.3\mu\text{m}$. Exceptions included the low γ' fraction alloys, which typically exhibited smaller cooling γ' , 0.05 to $0.15\mu\text{m}$, and the high γ' fraction alloys, which typically exhibited larger 0.5 to $0.7\mu\text{m}$ cooling γ' . TEM examination showed that each alloy contained very fine γ' , approximately $100\text{\AA}$ in size, stabilized during aging, located between the predominant intragranular particles. Alloy ET3 also contained a lath-like phase, determined to be eta, and Alloy R12, with 17 w/o Cr, showed some M_{23}C_6 and σ particles at grain boundaries. Most of the alloys also had occasional larger γ' particles at grain boundaries which also formed during cooling. The subsolvus treated Rene'95 alloy consisted of fine grains $3\text{--}5\mu\text{m}$ in size and occasional partially recrystallized, worked regions aligned with the extrusion direction. The γ' distribution consisted of small ($0.1\mu\text{m}$) cooling γ' , stabilized $100\text{\AA}$ aging γ' , and large ($1\text{--}3\mu\text{m}$) primary γ' at grain boundaries. Figure 1 shows the contrast in grain size and γ' distribution for the alloy microstructures in this study.

Mechanical Properties

Tensile properties of each alloy were determined at RT, 750, 1200, and 1400°F. Results showed that the coarse grain Rene'95 base alloy and experimental alloys R1, R2, R6, R7, R13, W2 and R97 on a density normalized basis had UTS and 0.2% YS values $\geq 85\%$ of the fine grain Rene'95 base alloy for RT-1400°F tests. An analysis of composition effects showed that an increase in γ' fraction typically resulted in an increase in yield strength (Alloys R1, R2, and R3), Figure 2, and a 0.2 w/o Hf addition raised strength and ductility at high temperature. The Co and Cr modifications had no consistent effects on tensile strength.

Creep (0.2%) and stress rupture properties of each alloy were determined at 1200°F (125 and 150ksi), 1300°F (125ksi), and 1400°F (60ksi) using constant load test techniques. A review of individual test data showed that only Alloy R7 exceeded the capability of the fine grain Rene'95 base alloy by 25°F for all creep-rupture conditions, while the coarse grain Rene'95 base alloy, Alloys R1-R4, W2-3, ET3, R6-7, R10-11 and the high γ' fraction alloys showed a greater than 75°F improvement for 1400°F 0.2% creep and stress rupture properties. Correlations between composition, microstructure, and creep-rupture performance were generally difficult due to changes in the ranking of alloys as a function of test condition. This is probably due to changes in the deformation mechanisms controlling creep-rupture over this regime of test conditions, particularly when the stress approaches the yield strength. Some consistent behavior, however, was observed for the ET-X series of alloys and the alloys having Co and Cr substitutions. A comparison of Alloys ET1, ET2 and ET3, showed that a higher amount of Ti in the γ' typically had a beneficial influence on creep and rupture properties. For the Co and Cr modified alloys, Figure 3 shows that the addition of 4 or 8 w/o Co was beneficial, while the Cr additions reduced creep-rupture performance.

Fatigue crack growth rates of each alloy were determined in air at 750 and 1200°F using a frequency of 20 cpm (0.33 Hz) and R-ratio of 0.05. FCGR were also determined using a 1.5 second ramp up to maximum load, 90 second hold at maximum load, and 1.5 second ramp down to minimum load hold time fatigue cycle at 1200°F. Although certain alloys showed better behavior under specific conditions, only Alloys R2 and W2 exhibited an acceptable improvement in fatigue crack growth resistance for all three test conditions.

Several correlations between composition, microstructure and FCGR were evident when specific groups of alloys and hold time behavior were considered. FCGR under 1200°F, 1.5-90-1.5 cycling decreased with 1) an increase in γ' fraction and size (Alloys R1 vs. R2 vs. R3), 2) an increase in Ti/Al ratio (Alloys ET 1 vs. ET 2 vs. ET 3), a small Hf addition (Alloys R1 vs. coarse grain Rene'95) and a Ta substitution for W (Alloys R1 vs. R4). The high γ' fraction Series 2 alloys (R97, R100, and I100) generally exhibited very good hold time fatigue resistance, but poorer behavior at low temperature. Figure 4 illustrates the differing effects of γ' fraction for low and high temperature tests. For other groups of alloys, such as those selected to evaluate the effects of Co and Cr substitutions, no consistent relationships between hold time FCG resistance, composition or microstructure were evident.

Figure 2: Effects of gamma prime fraction on 0.2% yield strength.

Figure 3: Effects of Co and Cr modifications on creep behavior at 1400°F/60ksi.

Figure 4: Effects of gamma prime fraction on fatigue crack growth rates.

The fatigue fractures of selected alloys were studied using scanning electron microscopy to determine whether characteristics of the morphology could be associated with Region II crack growth performance. For 750°F, 20 cpm cycling, Alloys W3 and R2 exhibited FCGR that were about 2X slower than those of Alloys R1 and R4. Although all four alloys showed a transgranular fracture mode, the alloys having the slower growth rates, W3 and R2, exhibited a more tortuous transgranular morphology. The general appearance of the fracture surface for these alloys could be classified as somewhat heterogeneous, while the fracture surfaces of the other two alloys were more homogeneous. At low to intermediate temperatures, faster FCGR have been associated with a homogeneous morphology. The fatigue morphology observed in studies of this type can be explained on the basis of slip band character and slip band strain localization [10].

For 1200°F, 20 cpm cycling, the crack growth rates of Alloys W2 and R2 were about 2X slower than Alloy R1. Alloys W2 and R2 again exhibited a transgranular mode, but the morphology was more homogeneous than that observed at 750°F. This correlates with the faster FCGR at 1200°F. Although Alloy R1 also exhibited transgranular features, some evidence of intergranular fracture was observed at higher values of ΔK . The faster FCGR of Alloy R1 can be associated with the tendency towards intergranular failure.

Significant differences in fatigue morphology were observed for 1200°F, 1.5-90-1.5 cycling. Comparison of a number of specimens showed that crack growth rates were fast for alloys exhibiting intergranular fatigue fracture, intermediate for alloys exhibiting a mixed mode of predominantly intergranular with some transgranular fracture, and slow for alloys exhibiting transgranular fatigue fracture.

Summary of Laboratory Scale Alloy Development Trials

Results of the alloy development study indicated that supersolvus heat treated alloys with low to medium γ' fractions had the best chance of meeting the property goals. Alloys R2 and W2 showed the best balance of critical properties and were thus selected from the laboratory development trials for scale-up. In addition, parallel efforts at GE Corporate Research and Development [11] showed that a supersolvus heat treated, high γ' fraction alloy, HK44, appeared to meet the program goals. Furthermore, contemporary studies in Europe showed that fine grain Astroloy and U720 had attractive properties. Consequently, HK44 (15Co, 10Cr, 3Mo, 4.9Al, 2.0Ti, 4.7Ta, 2.3Nb, 1.0V, 0.03B, 0.05C, 0.06Zr, bal. Ni), a derivative of IN100 having a γ' fraction of 61 a/o, and Astroloy (17Co, 15Cr, 5Mo, 4Al, 3.5Ti, 0.03B, 0.03C, 0.06Zr, bal. Ni), with a γ' fraction of 47 a/o, were also included in the program for further evaluation.

At this stage in the program it was recognized that coarse grain, lower γ' fraction alloys (R2 and W2) provided superior 750°F FCGR and probably superior quench crack resistance. The latter because of the lower supersolvus heat treatment required. It was also recognized that higher volume fraction γ' alloys typically exhibit superior high temperature crack growth resistance in the presence of a hold time. They might also exhibit better creep and tensile properties than lower γ' fraction alloys, but could be much more quench crack sensitive.

Figure 5: Effects of cooling rate on 0.2% yield strength for subscale alloy forgings.

Finally, subsolvus heat treated alloys (Astroloy) were expected to have the best tensile properties and to be the most producible, but were expected to fall short of the 750°F FCG and creep goals. Because no approach was clearly superior, it was decided to continue with the program with the best representation of these approaches. Available data indicated that U720 had properties similar to Astroloy and did not offer a unique approach, consequently it was not included in the scale-up program.

Subscale Screening Trials

The subscale screening trials were aimed at developing consolidation, forging and heat treatment parameters for the full scale processing of production size extrusions and disks, and further alloy down selection. Five inch diameter by two inch thick pancakes were isothermally forged from three inch diameter, fine grain extruded billet and then heat treated. All alloys were heat treated using supersolvus solution temperatures. This produced an average grain size for the alloys in the range of ASTM 8 to ASTM 5.5 (22μm to 78μm). Astroloy was also heat treated below the γ' solvus, resulting in a finer grain size of ASTM 10 ($\leq 11\mu\text{m}$). Anticipating a strong correlation between mechanical properties and cooling rate from the solution temperature, a wide range of quench media was used. Cooling rates varied from approximately 36°F/min (vacuum furnace cool) to 600°F/min (1000°F salt bath). The average cooling rates from 2000 to 1600°F were measured from embedded thermocouples. All alloys except Astroloy were aged at 1400°F for 8 hours. The coarse and fine grain Astroloy was aged at 1200°F for 24 hours, followed by 1400°F for 8 hours.

An extensive mechanical properties test matrix was employed to evaluate the subscale alloys. Properties included tensile (RT, 750, 1200°F), notched tensile (750°F, $K_t=3.5$), creep/rupture (1230°F/123ksi, 1300°F/125ksi), notched rupture (1300°F/125ksi), and fatigue crack growth (750°F/20cpm, 1200°F/20cpm, 1200°F/90 sec hold). In general, the tensile, creep and stress rupture strengths increased with increasing cooling rate, as shown in Figure 5 for 750°F 0.2% yield strength. Of the monotonic properties, yield strength required the fastest cooling rate (in excess of 300°F/minute) to meet property goals. The yield strength data for R2, W2, HK44 and fine grain Astroloy are similar for equivalent cooling rates, even though they contain a wide range of γ' fraction. Based on the low tensile strength of coarse grain Astroloy, it was eliminated from the program.

Residual life data are shown in Figure 6. The data represent the number of cycles required to grow a semicircular surface flaw from 0.008" to 0.060" in depth, with all data normalized to Rene'95 20 cpm behavior. At 750°F, W2 was the only alloy which consistently exhibited residual lives better than Rene'95. The 750°F data also show that the FCGR of W2 are relatively insensitive to cooling rate. At 1200°F and 20cpm, both W2 and R2 showed residual lives more than 2X Rene'95 for a cooling rate of 130°F/min.

Figure 6: Effects of cooling rate on residual life for subscale alloy forgings.

that HK44 was the most susceptible since it cracked when quenched in 1200°F salt (250°F/min). R2 and coarse grain Astroloy were not as susceptible as HK44, but did crack in the 1000°F salt bath (600°F/min). W2 did not crack in any of the quench media. Fine grain Astroloy was not quenched in the 1000°F salt bath, but it was believed that the alloy would not have cracked. The responses of the alloys suggested that it was highly unlikely that HK44 would meet the strength requirement without cracking, and that R2 and coarse grain Astroloy were marginal.

Full Scale Trials

Subscale properties vs. cooling rate data indicated that a cooling rate in the vicinity of 250°F/min would be required to produce the best balance of properties. A full scale high pressure turbine disk (HPTD) was forged from IN901 and instrumented with embedded thermocouples to characterize different quench media. It was determined that the HPTD should be quenched in an oil bath to attain the required cooling rate in the thick bore section and meet tensile strength requirements. HPTD's from alloys R2, W2, and HK44 were supersolvus solutioned, removed from the furnace and given a 120 second delay before being oil quenched. The Astroloy HPTD was given a subsolvus solution heat treatment with a 45 second delay before being quenched in oil. Only two alloys, W2 and Astroloy, met the tensile goals and did not quench crack. Average 0.2% YS and UTS for W2 were 161 and 214 ksi, respectively; for Astroloy they were 162 and 221 ksi. Additional testing of the W2 and Astroloy HPTD's resulted in the following conclusions. Astroloy had creep and rupture properties similar to Rene'95, while W2 met or exceeded the Rene'95+25°F goal (Figure 7). As shown in Figure 8, W2 had significantly better 750°F/20cpm FCG performance (1.3-1.8X improvement over Rene'95) than Astroloy (0.8-1.0X Rene'95). W2 FCGR at 1200°F/20cpm were about 2X lower than Astroloy and Rene'95, and exceeded the goal. Both alloys exhibited similar 1200°F/90 second hold time FCGR, and were 2.5X better than Rene'95. LCF behavior was different between the two alloys.

At faster cooling rates, W2 shows a significant reduction in residual lives. At 1200°F and 90 second hold time, the best alloy was HK44, which supports the correlation between slower hold time crack growth rates and higher volume fraction γ' . The hold time data for W2 also shows faster crack growth rates for increased cooling rate. The 1200°F FCGR data show an opposite trend with cooling rate, compared to the tensile data. A cooling rate for production hardware must be chosen so as to balance tensile requirements with FCG requirements. As will be described for production hardware, property goals were achieved and an attractive balance in properties was realized.

Fractography results for W2 showed that all 750°F/20cpm specimens had a transgranular fracture path, and the 1200°F/90 second hold time specimens had an intergranular fracture path. For the 1200°F/20cpm specimens, the fracture path transitioned from transgranular to intergranular as cooling rate increased.

Equally important to defining the cooling rate required to meet property goals is defining the alloys' susceptibility to quench cracking of (see Figure 5). Evaluations of the surfaces of pancake forgings after heat treatment revealed

Astroloy was similar to Rene'95 in average cycles to failure at 750 and 1000°F. W2 showed lower average LCF lives compared to Astroloy, but the scatter in the data was extremely small. Both alloys would produce design curves with minimums as good or better than conventional high strength nickel-base disk alloys. Because of W2's superior creep/rupture and fatigue crack growth properties, it was chosen for introduction into engine service in 1988. The alloy was given the official name Rene'88DT (damage tolerant).

Figure 7: Comparison of 0.2% creep results for full scale Alloy W2 and Astroloy forgings.

The ability to produce forgings with acceptable, predictable properties is dependent upon achieving controlled, uniform grain growth during supersolvus heat treatment. Abnormal, or critical grain growth was observed in the Rene'88DT forgings, a condition which would produce unacceptably low tensile strength and LCF behavior. A detailed study of the phenomenon revealed that the strain and strain rate deformation history during isothermal forging, and heat up rate to the solution temperature, controlled the resultant solutioned grain size [12].

Hardware Production

Transitioning Rene'88DT to production hardware introduced the complexity of three different disk geometries: the high pressure turbine disk, low pressure turbine disk, and forward outer seal. Knowledge gained from the full scale trials identified isothermal forging parameters and quench cracking as

Figure 8: Fatigue crack growth rates for full scale Alloy W2 and Astroloy forgings for a) 750°F/20cpm, b) 1200°F/20cpm, and c) 1200°F/1.5-90-1.5 test conditions. Baseline data are Rene'95 20cpm behavior.

two areas requiring strict process control. To aid in the identification of forging and heat treatment process parameters, computer aided modeling was employed. ALPID [13] was used to assure acceptable values of strain and strain rate throughout various disk geometries. NIKE-TOPAZ [14] allowed cooling rates, thermal gradients and thermally induced stresses to be calculated. To assure that a part would not quench crack, the NIKE-TOPAZ data were compared to on-cooling tensile data so that heat treatments could be defined which would keep surface stresses below the ultimate tensile strength of the material as measured from the on-cooling tensile test [15]. Accurate modelling of the forging and heat treatment processes allowed the disk and seal hardware to be produced successfully during initial attempts at three forging suppliers, enabling aggressive engine certification deadlines to be met.

Conclusions

Through the use of state-of-the-art technologies, Rene'88DT was developed with an attractive balance of critical disk properties and producibility. Significant improvements in FCGR and creep strength were achieved while still maintaining ultimate tensile strength no less than 90% Rene'95. Rene'88DT was developed and successfully transitioned to production engine disk hardware in the relatively short period of four years. Rene'88DT continues to expand its applications to both military and commercial engines.

Acknowledgements

The authors would like to thank the following individuals and companies for their invaluable contributions of time and creativity to the success of the Rene'88DT Introduction Program: P. Bhowal, R. Wallace, E. Raymond, M. Henry, J. Groh, R. Van Stone and C. Wukusick; Cameron Forge Company, Cameron Powder Systems, Wyman-Gordon Company, Ladish Company, and Wright Patterson Air Force Base.

References

- 1) T.T. King, J. Hurchalla, and D.H. Nethaway, "United States Air Force Engine Damage Tolerance Requirements," AIAA-85-1209, AIAA/SAE/ASME/ASEE 21st Joint Propulsion Conference, Monterey, Calif., July 8-10, 1985.
- 2) D.R. Chang, D.D. Krueger, and R.A. Sprague: "Superalloy Powder Processing, Properties, and Turbine Disk Applications", *Superalloys 1984*, M.Gell, C. Kortovich, R. Bricknell, W. Kent and J. Radavich, eds., TMS-AIME, Warrendale, PA, 1984, pp. 245-273.
- 3) J.L. Bartos and S.D. Antolovich, *Fracture*, 1977, Vol. 2, pp. 996-1006.
- 4) B. Lerch, "Microstructural Effects on the Room and Elevated Temperature Low Cycle Fatigue Behavior of Waspalloy", M.S. Thesis, Univ. of Cincinnati, Cincinnati, OH, Sept., 1981.
- 5) S.D. Antolovich and N. Jayaramen, "The Effect of Microstructure on the Fatigue Behavior of Ni-Base Superalloys", *Fatigue*, J.J. Burke and V. Weiss eds., Plenum Press, New York, N.Y., 1983, pp. 119-144.
- 6) R.H. Van Stone and M.F. Henry, "Processing Effects on Microstructure and Properties of Nickel-Base Superalloys", NADC-80124-60, NAVAL AIR Systems Command, Washington D.C., Jan., 1982.
- 7) E. Hornbogen and K-HZ Gahr, *Acta Metallurgica*, Vol. 24, 1976, pp. 581-592.
- 8) J. Lindigkeit, G. Terlinde, A. Gysler, and G. Lutjering, *Acta Metallurgica*, Vol., 27, 1979, pp. 1717-1726.
- 9) R.H. VanStone, "Long Life Disks from Rapidly Solidified Materials: Powder Metallurgy Alloy Development", AFWAL-TR-82-4032, Wright-Patterson AFB, Dayton, OH, June, 1982.
- 10) D.D. Krueger, R.H. VanStone and S.D. Antolovich, *Metall. Trans A*, Vol. 18A, 1987, pp. 1431-1449.
- 11) M.F. Henry, Unpublished Research, General Electric Corporate Research and Development, Schenectady, N.Y.
- 12) US Patent No. 4,957,567, D.D. Krueger, R.D. Kissinger, R.G. Menzies, and C.S. Wukusick, General Electric Co.
- 13) S.I. Oh, *International Journal of Mechanical Science*, Vol. 24, 1982, pp. 479-493.
- 14) R.A. Wallis and P.R. Bhowal, "Property Optimization in Superalloys Through the Use of Heat Treat Process Modeling", *Superalloys 1988*, D. Duhl, G. Maurer, S. Antolovich, C. Lund, and S. Reichman, eds., TMS-AIME, Warrendale, PA, 1988, pp. 525-534.
- 15) R.A. Wallis, N.M. Bhathena, P.R. Bhowal and E.L. Raymond, "The Application of Process Modeling to Heat Treatment of Superalloys", AGARD-Structures and Materials Panel, Turkey, 1987.