RENÉ N4: A FIRST GENERATION SINGLE CRYSTAL TURBINE AIRFOIL ALLOY WITH IMPROVED OXIDATION RESISTANCE, LOW ANGLE BOUNDARY STRENGTH AND SUPERIOR LONG TIME RUPTURE STRENGTH

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Introduction

GE Aircraft Engine's first generation single crystal (SX) turbine airfoil alloy, René $N_{n}^{(1)}$ was extensively tested and utilized in turbine blades of development engines. Factory engine testing then revealed that tips of the René N blades suffered from excessive oxidation. It was obvious that a more oxidation resistant single crystal alloy would be required. Two SX development programs were conducted; C. Wukusick and W. King to develop a more oxidation resistant SX alloy, and E. Ross to increase the low angle boundary (LAB) strength of the René N, which could, hopefully, be applied to the new more oxidation resistant SX alloy. This paper will describe the development of this improved first generation SX alloy, René N4,⁽²⁾ which evolved from these two programs.

Improving the oxidation resistance of René N

René N, GE's first SX alloy was 100°F stronger (longitudinal rupture strength) than GE's workhorse conventionally cast turbine airfoil alloy, René 80.

The chemistry of René N is shown in Table I. It can be seen that René N has higher titanium (4.2%) than aluminum (3.7%). The higher level of titanium vs. aluminum imparted excellent hot corrosion resistance to the René N, but sacrificed oxidation resistance as revealed by the engine testing.

Wukusick and King evaluated the hot corrosion and oxidation resistance of many René N variations. Four elements (Mo, W, Cb, Co) always remained constant. The results of the other elements studied were:

1. **Chromium:** The efficacy of Cr in environmental resistance is well known (as long as good stability is retained). Levels of chromium from the 9.25% of René N to 10.25% were evaluated in hot corrosion. As shown in Figure 1, when tested at 1600°F with 2 ppm salt, there were significant improvements in the hot corrosion resistance from the 9.25% to a level of 9.75% Cr with only slight changes at higher levels. Microstructural stability was retained in all alloys. The new replacement alloy would therefore be aimed at 9.75% Cr. Hot corrosion tests at 1700°F also showed significant improvement at 9.75% and 10.25% Cr.

Table I. Aim compositions of René N and René N4, wt.%

Element	René N	René N4
Cr	9.25	9.75
Мо	1.5	1.5
Та	4.0	4.8
Ti	4.2	3.5
Al	3.7	4.2
w	6.0	6.0
Сь	0.5	0.5
Co	7.5	7.5
В		0.004
Hf	_	0.15
С		0.05

2. **Carbon:** The grain boundary strengthening elements (B, Zr and Hf), as well as C, which were not present in René N, were studied to determine their effect on hot corrosion when added to René N. Of these elements only C significantly affected hot corrosion behavior. As shown in Figure 2, there was about an 8X improvement in 1700°F-5 ppm hot corrosion resistance when the C content was increased to 0.05%. The new alloy would then contain 0.05% C. It was also decided to reduce the titanium content to improve oxidation resistance. As a result, the Ta content was correspondingly increased to maintain the γ 'volume fraction and account for the formation of Ta/Ti carbides.

The previously discussed increases in Cr, C and Ta will all increase the new alloy's hot corrosion resistance.

3. Al/Ti Ratio: René N contains 4.2% Ti and 3.7% Al (approximately a 0.9 ratio). Increasing the Al content and lowering the Ti content improved the oxidation resistance of the alloy system, as shown in Figure 3. The hot corrosion resistance, however, was lowered as also shown in Figure 3, but the previously discussed addition of C

significantly reduced this degradation. The Al/Ti content of the new alloy was selected to be 4.2% Al and 3.5% Ti or a ratio of 1.2; this resulted in a 50% reduction in 100 hour oxidation metal loss in a 2150°F Mach 1 test.



Figure 1. Effect of Increased Cr on Hot Corrosion of René N.



Figure 2. Effect of Carbon on Hot Corrosion of René N.



Figure 3. Effect of Al/Ti Ratio on Oxidation and Hot Corrosion. Also Showing Effect of Carbon on Hot Corrsion Resistance of René N Type Alloys.

4. Hafnium: Wukusick and King also found that the addition of a small amount of Hf, (0.10 to 0.20%) "caused a further reduction in the oxidation rate during the initial stages of oxidation. Hf tends to promote oxide adherence, providing an incubation period when virtually no metal loss occurs". The new alloy would therefore contain 0.15% Hf.

Summary: the New Alloy's Base Chemistry

Wukusick and King had developed a new alloy which not only maintained the strength of René N but also significantly improved the oxidation resistance, as shown in Figure 4. As shown in Table I, the new alloy, René N4, has 0.5% more Cr, 0.8% more Ta, 0.5% more Al, 0.7% less Ti plus 0.15% Hf and 0.05%C, than René N. This chemistry would be combined with the low angle boundary (LAB) work of Ross to realize the final René N4 composition.



Figure 4. High Velocity Oxidation of René N vs René N4.

Low Angle Boundary (LAB) Strengthening

Conventionally cast/equiaxed turbine airfoil alloys such as René 80 contain approximately 150 ppm of B, 0.10% to 0.18% C, and a small amount of Zr. These elements, B particularly, are vital to high angle grain boundary (HAB) strengthening. Without these elements, the rupture strength of these alloys is dramatically reduced. Hafnium, approximately 1.5%, was added in the 1970's by Lund, et al to newer and stronger conventionally cast turbine airfoil alloys to improve castability and rupture strength, particularly lower temperature strength. Similar additions of B, Hf, C and Zr continued with directionally solidified (DS) alloys, such as MM200H and René 142, since these alloys were DS cast with HAB's parallel to the airfoil leading and trailing edges. These grain boundary strengtheners were required to provide transverse rupture strength and ductility.

When directionally solidified single crystal (SX) alloys were invented, B, Hf, C and Zr were not added, since these elements lowered the incipient melting temperature and the SX alloys were to be cast without grain boundaries. Therefore, GE's first SX alloy, René N, did not contain any purposeful addition of B, C, Zr or Hf (Table I).

In evaluating the first SX material cast with René N, it was found that often, the casting process resulted in material that had longitudinal grain boundaries. These boundaries were caused by perturbations in the casting process which occurs with all SX alloys.

Strengthening LAB's

The casting vendor detects these boundaries by carefully etching the parts after surface cleaning. The amount of misorientation of the boundary is determined by x-ray diffraction (small amounts of misorientation can be determined visually by the degree of reflectivity of the adjacent grains). Since the boundaries of René N contained no strengthening additives, it was necessary to determine what degree of the misorientation would be acceptable. GE, therefore, conducted a program on René N slabs which had formed boundaries with various misorientations. This program determined that the transverse rupture strength would not be reduced if the misorientation between the adjacent grains was 6° or less. (These 6° or less boundaries were called low angle boundaries or LAB's).

As a result of the 6° LAB misorientation requirement, a number of René N turbine airfoils were scrapped. A program was therefore initiated at GE to determine if small additions of grain boundary strengtheners could be added to René N (and later to René N4) which would increase the LAB grain boundary acceptance level to $>6^\circ$ without reducing the transverse rupture strength or heat treatability of the airfoils.

Round 1: Procuring slabs which contain LAB's was expensive, difficult, and had to be done at a casting vendor. Round 1 slabs were therefore cast at GE by DS casting (with high angle boundaries) 10 heats of René N to which varying amounts of C, B, Zr and Hf were added. 1800°F transverse rupture tests were then conducted across the HAB's. As shown in Table II, the best rupture lives were with 0.05%C, 0.01%B and 0.004% Zr additions (Heat 10). The 1800°F rupture life was 25 to 100X that of DS René N which had been previously tested with no grain boundary additions.

Round 2: In this round, René N slabs were again DS cast by GE with heat 10 as a base (now called heat 11) and with 1600° F rupture tests added. As can be seen in Table III, there were a few surprisingly low results in heats 12, 14 and 19, but heat 11 had good lives at both 1600° F and 1800° F. Heat 21 was conducted to note if B alone was as good as B C and Zr. It can also be seen in Table III that the B alone was very beneficial at 1600° F and 1800° F but not as efficacious as when the three elements were added. (It should be noted that in the first round C or C plus Hf, was not beneficial to strengthening).

		Addition	s, wt.%		1800°F/30 ksi		1800°F/20 ksi	
Heat	с	В	Zr	Hf	Life, hrs	R.A., %	Life, hrs	R.A., %
1	0.02	0.003	0.004		2.7	0.6	62.0	0.0
2	0.035	0.0035	0.007		1.8	0.0	30.4	1.2
3	0.05	0.005	0.01		4.6	0.6	63.9	0.6
4	0.035	0.0035	0.007	0.05	11.3	0.0	105.6	0.0
5	0.05			0.05	0.1	0.0		-
6	0.05			0.1	0.2	0.0		
7	0.08				1.3	0.0		
8	0.08	0.002	0.004		1.8	0.6		
9	0.08			0.05	1.8	1.2		
10	0.05	0.01	0.004		53.4	3.8	175.1	0.0
DS René N					< 0.5		~7	

Table II. Round 1 - Transverse Rupture Strength of DS René N with Elemental Additions.

	Additions, wt.%		1800°F	1800°F/30 ksi		1800°F/20 ksi		1600°F/50 ksi		
Heat	с	В	Zr	Hf	Life, hrs	R.A.,%	Life, hrs	R.A.,%	Life, hrs	R.A.,%
11	0.05	0.01	0.004		42.6	1.2	223.3	2.2	498.6	2.0
12	0.05	0.01		0.05	23.3	1.2	211.7	2.2	F.O.L.	2.6
- 13	0.05	0.01		0.1	27.3	0.0	174.4	1.7	295.8	1.5
14	0.05	0.01		0.2	F.O.L.	0.6	164.9	1.4	52.8	1.2
15	0.05	0.015	0.004		18.3	0.6	332.4	1.5	601.4	3.4
16	0.07	0.015	0.004		22.4	1.2	328.3	2.0	392.7	3.2
17	0.07	0.015		0.05	22.3	1.3	162.3	1.0	213.7	2.0
18	0.07	0.015		0.1	34.8	2.0	300.5	1.8	8.5	2.5
19	0.07	0.015		0.2	F.O.L.	0.0	F.O.L.	1.5		
20	0.07	0.02			41-46.6	0.0	74.6	1.7	F.O.L.	-
DS René N					< 0.5		~ 7		< 4	
	Additions, wt.%		1800°F	1800°F/23 ksi		1800°F/20 ksi		/55 ksi		
21		0.01			78.2	0.0	103.4	12	150.2	3.2
DS René N					< 3		~ 7		< 3	

Table III. Round 2 - Transverse Rupture Strength of DS René N with Elemental Additions Made to a Production Heat.

Round 3: A 300 lb. heat of the new (René N4) Wukusick/King chemistry which had the modified levels of Cr, Al, Ti, and Ta plus 0.05%C (but did not contain Hf) was procured, and then DS cast at GE with no additions of B, Zr, or Hf in one slab and with varying amounts of these additions in 5 heats. Transverse rupture tests were then conducted at 1600°F, 1800°F and 2000°F. As shown in Table IV, the base heat (with C the only addition) had 6 to 10X the transverse rupture life of DS René N, but the B and Hf additions increased the lives another 10 to 20X. The 0.0075% B plus 0.2% Hf (with the base 0.05%C) was particularly excellent at 1600°F and 1800°F. The results at 2000°F showed no significant improvement. **Round 4**: Five 300 lb. heats of the modified chemistry (René N4) were then procured; all had 0.05%C and 0.15 or 0.2% Hf. These 5 heats had 0 (0 ppm), 0.002% (20 ppm), 0.003% (30 ppm), 0.004% (40 ppm) or 0.0075% (75 ppm) B. They were also <u>DS</u> cast into slabs at GE and transverse rupture tested at 1400°F, 1600°F, 1800°F and 2000°F. As shown in Table V, 0 ppm B and 20 ppm B were not beneficial to transverse rupture strength at 1400°F to 2000°F, but 30 ppm, 40 ppm and 75 ppm all had excellent transverse rupture strength at 1400°F to 1800°F, but again was not beneficial at 2000°F.

Table IV.	Transverse	Rupture Strength	of DS René N	4 with Elemental	Additions Made to	a Production Heat.	All Contain	0.05 wt.% C
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	Ad	Additions, wt.%			1600°F/50 ksi		1800°F/23 ksi		2000°F/14 ksi	
Heat	В	Zr	Hf	Life, hrs	R.A., %	Life, hrs	R.A., %	Life, hrs	R.A., %	
Cl				43.6	1.0	18.5	1.1	3.1	1.2	
C2	0.0075			320.7	4.7	325.2	0.6	11.4	0.8	
C3	0.0125			397.0	18.2	73.4	0.0			
						147.0	0.0			
C4	0.0125	0.006		379.0	8.7	343.1	1.2			
C5	0.0075		0.2	520.3	11.3	536.7	1.0	4.0	0.0	
<u>C6</u>	0.0125		0.2	455.4	7.9	409.0	0.6			
DS René N4 (no C)				< 4		< 3		< 2 (coated)		

	Str	ess Ruptu	Compararison to René N		
B, ppm	Temp, F	Stress, ksi	Life, hrs	R.A., %	Life, hrs
0	1400	90	4.0	0.0	N.A.
	1600	55	1.9	0.0	< 3
	1800	26	2.3	2.7	< 1
	2000	12	3.1	0.0	< 4
20	1400	90	3.3	0.0	N.A.
	1600	55	15.6	0.8	< 3
	1800	26	9.2	0.0	< 1
	2000	12	4.5	0.0	< 4
30	1400	90	184.4 ⁽¹⁾	3.8	N.A.
	1600	55	69.2	0.0	< 3
	1800	26	65.6	0.0	< 1
	2000	12	9.1	1.3	< 4
40	1400	90	92.5 ⁽²⁾	6.2	N.A.
	1600	55	133.8	2.5	< 3
	1800	26	50.0	0.0	< 1
	2000	12	1.8	0.0	< 4
75	1400	90	92.4 ⁽³⁾	32.0	N.A.
	1600	55	54.1	0.0	< 3
	1800	26	98.1	0.6	< 1
	2000	12	4.1	0.6	< 4

Table V. Transverse Rupture Strength of DS René N4.*

* 300 lb. heats contain 0.05 wt.% C and 0.15-0.20 wt.% Hf

(1) Step loaded to 120 ksi (2) Step loaded to 150 ksi (3) Step loaded to 140 ksi

Round 4A: The 300 lb. heats of René N4 from round 4 with varying B levels were <u>SX</u> cast into slabs at a casting vendor using a seeding method to produce slabs with a 9°-15° LAB down the center of the slab. The resultant LAB's were tested in transverse rupture at 1500°F to 2000°F. The 1500°F to 1800°F results (Table VI) show that 12° to 15° LAB's with 30-75 ppm of B had >50X improvement in transverse rupture over previously tested René N with 12° LAB's. The results at 1800°F to 1900°F (Table VII) were also excellent, while tests at 2000°F again showed no significant improvement.

Scale-Up Of René N4: It was determined from the previously discussed results that a B level of 30 ppm minimum with 0.05%C and 0.15%Hf would be necessary in René N4 to achieve a significant improvement in René N4's LAB capability from the 6° maximum of René N to possibly 12°. To keep the B level as low as possible to prevent incipient melting, a 3000 lb heat of René N4 was procured with 40 ppm of B (for an eventual range of 30-50 ppm), 0.05%C and 0.15%Hf.

Slabs were then <u>SX</u> cast at the casting vendor with 11° to 16° LAB's which were tested at 1500°F to 2000°F. As shown in Table VIII, the results (even at 2000°F) were excellent.

Table VI. 1500-1700°F Transverse Rupture Strength Across LAB's in René N4.*

	René N					
B, ppm	LAB °	Temp,	Stress,	Life, hrs	R.A., %	Life, hr;
		۴	ksi			(LAB°)
40	14	1500	75	185.0 ⁽¹⁾	2.5	N.A.
0	13	1600	58	24.6	0.0	2 (12°)
0	12	1600	58	10	1.2	2 (12°)
20	9	1600	58	146.0	0.0	15 (10°)
20	12	1600	58	77.7	0.0	2 (12°)
40	14	1600	58	304.0 ⁽²⁾	2.5	N.A.
40	15	1600	58	109.8	1.2	2 (12°)
30	12	1600	~55	175.1	1.8	2 (12°)
75	14	1600	58	347.9	1.8	2 (12°)
40	14	1700	45	92.2	0.7	N.A.

* Heats contain 0.05 wt.% C and 0.15-0.2 wt.% Hf

(1) Step loaded to 135 ksi to failure

(2) Step loaded to 78 ksi to failure

Table VII. 1800-1900°F Transverse Rupture Strength Across LAB's in Renć N4.*

	René N					
B, ppm	LAB °	Temp,	Stress,	Life, hrs	R.A.,	Life, hr;
		ዋ	ksi		%	(LAB °)
0	12	1800	27.5	84.6	0.6	N.A.
0	13	1800	30	10.7	1.3	36 (11°)
0	12	1800	30	55.7	1.2	36 (11°)
20	12	1800	30	31.0	0.6	36 (11°)
20	9	1800	30	68.3	1.9	36 (11°)
20	12	1800	30	52.0	0.0	36 (11°)
30	12 Radius	1800	~28	234.0 ⁽¹⁾	33.6	N.A.
	4 Gauge		30	234.0 ⁽¹⁾		134 (°4)
40	15	1800	24	124.7	0.6	N.A.
40	14	1800	30	108.7	1.3	N.A.
40	15	1800	30	33.3	0.0	N.A.
75	14	1800	30	73.0	0.6	N.A.
20	12	1900	14	140.8(2)	0.0	N.A.
40	15	1900	14	115.7	0.7	N.A.
75	14	1900	14	129.4	0.9	N.A.

* Heats contain 0.05 wt.% C and 0.15-0.2 wt.% Hf

(1) Step loaded to 50 ksi to failure

(2) Step loaded to 17 ksi to failure

The 1600°F and 1800°F transverse rupture life across LAB's of SX René N and SX René N4 are plotted in Figures 5 and 6. It shows the LAB lives of René N (no grain boundary additions) which resulted in the 6° maximum René N LAB specification vs. the René N4 lives which resulted in the 12° LAB maximum René N4 specification. This LAB improvement was due primarily to the B and C added to René N4.

	DS René N				
LAB °	Temp, °F	Stress, ksi	Life, hrs	R.A., %	Life, hrs
11	1600	58	380.1	24.9	< 3
	1800	30	118.8	0.6	< 0.4
	1900	14	175.8	0.0	5
	2000	12	89.3	0.0	.4
14	1500	75	210.8 ⁽¹⁾	7.4	10
	1600	58	171.4	2.5	< 3
	1800	30	51.0	2.5	< 0.4
	1800	24	296.1	0.0	2
	1900	14	148.7	0.0	5
	2000	12	47.9	1.9	4
16	1600	58	168.0	3.7	< 3
	1800	30	73.1	0.0	< 0.4
	1900	14	192.8	0.0	5
	2000	12	49.5	0.0	4

Table VIII. Transverse Rupture Strength Across LAB's in René N4.

* (1) Step loaded to 135 ksi to failure

Conclusions (LAB Program)

- René N4 contains 0.004% (40 ppm) B + 0.05% C for LAB strengthening (plus 0.15% Hf)
- René N4 has 2X the maximum allowable LAB of René N (12° vs 6°)
- Second and third generation GE single crystal alloys have similar additions for LAB strengthening and 12° maximum allowable LAB's



Figure 5. Effect of LAB Misorientation on 1600°F Rupture Life of René N vs. René N4.

René N4 Long Life Rupture Behavior at 1800°F

The "first generation" SX alloys are still preferred in several applications due to advantages of density, cost and surprisingly strength. In the 1800°F temperature range, alloys like René N4 are optimally balanced for prolonged rupture life; as Figure 7 shows, René N4 surpasses the "stronger" next generation alloys in the 7000 - 10000 hour life regime.



Figure 6. Effect of LAB Misorientation on 1800°F Rupture Life of René N vs. René N4.



Figure 7. 1800°F Long Time Rupture Behavior of René N vs. Other Single Crystal Superalloys.

An interesting trend is noted in 1800°F rupture behavior by studying the rupture curves for several commercial and experimental alloys. Figure 8 shows alloy density plotted against the 1800°F isothermal rupture slope taken from regressing the logarithm of rupture life versus the logarithm of stress. Here a larger slope is a figure of merit for prolonged low stress, rupture life. Ironically as alloy density increases, the slope decreases. The alloy developer's traditional strengthening methodology has been to increase the refractory content largely in the matrix using rhenium additions. This approach unfortunately can compromise stable long term rupture behavior in the 1800°F temperature range.

An extensive internal research and development $program^{(3)}$ was undertaken at General Electric Aircraft Engines to study several superalloys with varying 1800°F slope behavior. Interrupted creep testing was performed over a range of temperatures and stresses from 10 to 1000 hours. Some specimens were continued to rupture. Optical and TEM metallographic studies were completed. The microstructural evolution of rafting was noted along with the attendant dislocation reactions, especially the formation of mismatch accommodating nets at the gamma-gamma prime interface. Particular attention was paid to the comparison between René N4 and a 3% Re alloy designated alloy 821, which displayed inferior life to René N4 at low stress owing to its steeply sloped isothermal rupture behavior when viewed with the traditional time axis as the horizontal one. Other second generation alloys were found to duplicate alloy 821's behavior in varying degrees.



Figure 8. Slope of 1800°F Rupture Life vs. Stress as a Function of Density. Supplemental Data from Ref. 4 and 5.

The principal difference in 1800°F rupture behavior between René N4 and the second generation alloy 821 was associated with the rafting behavior. In René N4, well defined primary, secondary and tertiary stages of creep occurred, typical of high temperature creep behavior. Rafting occurred early, usually completed by the onset of the secondary regime. The gamma prime in secondary creep remained dislocation free. The second generation single crystal materials typically exhibited limited primary creep, a short secondary and prolonged tertiary creep response. Rafting mostly occurred during secondary creep; furthermore, as stress was decreased in alloy 821, a change in the stress exponent (n) occurred from ~ 10 to 3.5. This change in exponent with stress was attributed to the fact that rafting occurred earlier in life at lower stresses, so that the weaker rafted structure was present for a larger fraction of the total specimen life. The weakness of the gamma prime rafts was highlighted by pre-rafting experiments. Pre-rafting significantly degraded alloy 821 creep strength while being somewhat neutral to slightly degrading in René N4. This pre-rafting was performed at both 1800°F and 2000°F followed by 1800°F creep testing.

It was postulated that alloy 821 possessed a strength imbalance between the gamma and gamma prime phase. When the gamma matrix phase is strengthened without an accompanying strength increase in gamma prime, the gamma channel stresses can build up to higher levels (provided other factors such as gamma and gamma prime modulus difference, lattice mismatch and volume fraction are constant). This makes the gamma prime even more vulnerable to dislocation activity. To support this concept, it was shown that alloying the gamma prime in alloy 821 with Ta or Ti additions improved the 1800°F rupture life.⁽³⁾ These alloys are shown in Figure 8 as X1-X6.

Apparently, at 2000°F a more overall favorable strength balance was maintained between the gamma and gamma prime in alloy 821. At this

temperature, a more typical 3X rupture life advantage was noted between first and second generation materials (including alloy 821) without any tendencies for life crossover at low stress. Well defined primary, secondary and tertiary creep regimes were noted.

Interestingly, in third and fourth generation (with even more Re) single crystal alloys, there is a tendency for prolonged creep at 1800°F to induce a change to a gamma prime matrix during rafting.⁽⁶⁾ Here the strength of the gamma prime is critical. Unique strengthening of the gamma prime phase can be induced along with exceptional alloy stability in a new class of superalloys recently discovered⁽⁷⁾ called "Reverse Partitioning Ni-base Superalloys".

Acknowledgments

The authors particularly thank Carl Wukusick, now retired from GE Aircraft Engines, and Warren King who is now with GE Power Generation for the development of René N4's base chemistry. The authors gratefully acknowledge the contribution of Gary McCabe of GE Aircraft Engines in making the small heats of René N and René N4 and then the SX and DS (HAB) slabs. Other GE members of the alloy development and scale-up teams included Dick McDaniel, Peg Jones and Tom Berry. Of particular importance was Bob Allen who procured the funding and led the scale-up efforts for René N4. A key ingredient in the success of the program was the casting of the René N4 LAB slabs by Howmet, Whitehall, Michigan. The authors acknowledge the single crystal superalloy creep deformation studies which were performed at GE Aircraft Engines by Tresa Pollock, Wendy Murphy and Bob Field. We would like to also extend special thanks to Stephanie Boone of GE Aircraft Engines for being instrumental in the organization and execution, and Scott Walston for editing this paper.

References

- Carl S. Wukusick and Leo Buchakjian, Jr., U.S. Patent 5,154,884, "Single Crystal Nickel-Base Superalloy Article And Method For Making", 1992.
- E.W. Ross, C.S. Wukusick and W.T. King, U.S. Patent 5,399,313, "Nickel-Based Superalloys For Producing Single Crystal Articles Having Improved Tolerance To Low Angle Boundaries", 1995.
- W. Murphy, T. Pollock and R. Field, Mechanism Studies in René N5 and Related Single Crystal Superalloys", (GE Aircraft Engines internal report, 1991).
- A.D. Cetel and D.N. Duhl, "Second Generation Nickel-Base Single Crystal Superalloy", Superalloys 1988, ed. D.N. Duhl, et al, TMS, 1988, 235-244.
- K. Harris, et al, "Development of Two Rhenium-Containing Superalloys for Single-Crystal Blade and Directionally Solidified Vane Applications in Advanced Turbine Engines", <u>J. Mat. Eng. and</u> <u>Perf.</u> (1993), 481-495.
- W.S. Walston, K.S. O'Hara, E.W. Ross, T.M. Pollock and W.H. Murphy, "René N6: Third Generation Single Crystal Superalloy", Superalloys 1996, ed. R.D. Kissinger, et al, TMS, 1996.
- 7. K.S. O'Hara, W. S. Walston, E. W. Ross and R. Darolia, US Patent 5,482,789, "Nickel-Base Superalloy And Article", 1996.