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MICROSTRUCTURAL VARIABLES CONTROLLING TIME-DEPENDENT CRACK GROWTH IN A P/M SUPERALLOY

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Abstract

A study was performed to determine the variables which influence hold time crack growth resistance of Alloy 10, a new powder metallurgy (P/M) superalloy. In a well controlled study, both the effect of compositional changes and variation in heat treatments were investigated. The results indicate that significant changes in the alloy's niobium, tantalum and cobalt content did not have an appreciable effect on hold time crack growth resistance.

In contrast to the composition study, the heat treatments evaluated produced up to an order of magnitude changes in the crack growth resistance. Quantitative image analysis was performed to analyze the microstructural features produced by each heat treatment. It was found that the cooling γ' precipitate size distribution is closely related to the measured hold time crack growth behavior. The larger the mean size of the cooling precipitates, the better is the resistance to hold time crack growth. It is proposed that the size and distribution of the precipitates might play an important role in determining the extent of crack tip relaxation which occurs through creep-type processes. The differences in the stress relaxation rate influence the crack driving forces and thus crack growth resistance.

Introduction

The increase in the operating temperatures of the new generation of advanced gas turbine engines has focused attention on the higher temperature capabilities of current P/M nickel based superalloys primarily. The P/M alloys are primarily utilized to produce turbine and compressor disk components. At the higher operating temperatures, the time dependent crack growth (TDCG) properties of these alloys become increasingly important and can have large influence on the design life of the disk components.

Considerable amount of research has been performed over the years in order to explain and model the time dependent crack growth behavior in superalloys. It has been shown that the typically observed intergranular failure mode is primarily the result of an oxidation process which preferentially attacks the grain boundaries and lowers the resistance to hold time crack growth (1). Numerous hypotheses, which have been proposed to explain and model the large variability in TDCG behavior, concentrated mostly on the resistance of various grain boundary phases and species to environmental attack. The general assumption has been that if phases that are more susceptible to oxidation at the grain boundaries are identified, compositional adjustments could be made which would improve resistance to oxidation. Over the years number of grain boundary candidates susceptible to environmental attack have been proposed, however none of them withstood the test of time.

Recently a new hypothesis explaining TDCG has received considerable attention. The researchers identified grain boundary niobium carbides as the phase most susceptible to environmental attack (2). They proposed that NbC decompose to produce niobium oxides which embrittle the grain boundaries. It was suggested that the removal of Nb would substantially improve TDCG resistance. In addition to removal of niobium, increase in the tantalum content has also been proposed to improve TDCG resistance (4). Another group of researchers (3) have shown that slow cooling rates from a supersolvus solutioning treatment resulted in a significant improvement in TDCG resistance. The slow cooling rates lead to preferential growth of the γ' precipitates at grain boundaries. The growth of large precipitates causes the grain boundaries to become serrated which the researchers associated with improved TDCG resistance. They attributed this improvement to a reduction of boundary surfaces perpendicular to the stress axis. Again, the emphasis was placed on the role of the grain boundaries.

Others have noted that in addition to the importance of grain boundaries, other mechanisms can also influence hold time crack growth resistance. In particular, the response of the highly stressed crack tip region to prolong holds may be influenced by creep stress relaxation (4,5). The crack driving force may be reduced by local stress relaxation during hold time and thus have an effect on the TDCG resistance.

The research effort described herein was undertaken to systematically evaluate both chemistry modifications and the influence of heat treatment on TDCG properties of P/M superalloys. In a well controlled study of P/M Alloy 10, four different variations of niobium to tantalum ratios were investigated at three different cobalt levels for a total of seven compositions. For each chemistry modification, the alloy was heat treated to obtain both subsolvus as well as supersolvus microstructures.

In order to evaluate the influence of cooling rate and subsequent heat treat steps on TDCG behavior, a heat treat study was performed. Six different microstructures produced by the heat treatments were quantified and their relationship to TDCG behavior established.

Material and Procedure

Alloy 10 is a high strength nickel-based superalloy with an approximate γ' content of 55%. The alloy was developed by Honeywell for application in smaller gas turbine engines.

Composition Study

The Alloy 10 composition modifications investigated in the program are listed in Table I. The seven compositions were designed to study the effect of niobium, tantalum and cobalt on TDCG behavior. Compared to the baseline chemistry, tantalum levels were increased while niobium levels were decreased. The rationale for these choices were tied to the opinion that niobium accelerates TDCG while tantalum retards crack growth (2, 6). In addition, the effect of cobalt on crack growth resistance was evaluated since it is known that cobalt content has a strong influence on the solvus temperature.

Table I. Chemistry modification study

Element	D	C	E	N	A	B	Alloy 10
Cr	11.2	11.2	11.3	11.0	11.0	11.2	11.2
Co	15.3	15.2	17.1	15.2	17.0	19.2	15.7
Mo	2.6	2.6	2.6	2.6	2.5	2.6	2.6
Ti	3.8	3.8	3.8	3.8	3.8	3.8	3.8
Al	3.9	3.9	4.0	3.9	3.8	3.9	3.9
W	5.7	5.9	5.8	5.6	5.60	5.8	5.8
Nb	0	0.8	0.9	0.8	0.9	0.9	1.7
Ta	1.8	1.8	1.9	1.0	1.0	1.0	1.0
C	.03	0.042	0.04	.036	0.04	.045	0.045
B	.03	0.03	0.03	.03	0.03	0.03	0.03
Zr	0.1	0.1	0.1	0.1	0.1	0.1	0.1
Ni	55.5	54.6	52.4	55.9	54.2	51.4	54.1
Nb/Ta	0	0.44	0.47	0.8	0.9	0.9	1.7

Baseline Alloy 10 powder as well as the six composition variations were produced by argon atomization. The powder heats were compacted and extruded to produce 75 mm diameter billets. EDM was used to extract specimen blanks from the extrusions. The blanks were then solutioned at 1149°C to obtain a fine grain, subsolvus microstructure or at 1199°C to obtain a coarse grain supersolvus microstructure. All the blanks were air cooled and subsequently aged at 760°C for 16 hours. The resulting grain size was approximately ASTM 11.5 for the subsolvus heat treated bars and ASTM 7.5 for the blanks subjected to the supersolvus heat treatment.

The blanks were used to machine surface flaw geometry crack growth specimens, termed K_b bar. This type of specimen was selected since it best depicts the crack front geometry, crack constraint and the remote stress found in actual disk engine components. The crack growth testing was conducted at 704°C with a 90 second dwell at the peak load. The load cycle had an R-ratio of 0.1. The crack growth measurements were obtained throughout the test using an electrical potential drop system. In addition to the crack growth tests, tensile properties at 704°C were also measured.

Heat Treat Study

The baseline Alloy 10 extrusion used in the composition study was also utilized to perform a heat treat study in order to evaluate the effect of various microstructures on the TDCG behavior. For the heat treat study only the supersolvus condition was investigated. The main variable evaluated was the effect of the cooling rate from the supersolvus solutioning temperature of 1199°C on the crack growth properties. In addition to varying the cooling rate, the effect of subsequent subsolvus resolutioning steps was also examined.

Shown in Table II are the details of the heat treatments. All the blanks were supersolvus solution treated for 1 hour at 1199°C. Half of the blanks were removed from the furnace and allowed to air cool with the cooling rate being monitored by an embedded thermocouple. The cooling rate for the air cooled blanks was approximately 183°C/min from the solutioning temperature to 649°C. The remaining blanks were furnaced cooled from the solutioning temperature to 982°C at which point the blanks were removed from the furnace and allowed to air cool. The average cooling rate for the furnace cooled blanks was 27°C/min from 1199°C to 982°C.

Two different subsolvus resolutiong heat treatments were used for selected blanks while the remaining blanks were not given any further solutioning treatment. The subsolvus resolutioning treatments consisted of exposure for 1 hr at temperature of either 1149°C or 1171°C followed by air cool. This type of subsolvus resolutioning treatment is

sometimes used for a supersolvus microstructure to increase the tensile properties of the material.

Table II. Details of the Heat Treat Study

Specimen	Supersolvus Solution	Type of cool	Cool Rate °C/m	Subsolvus Resolution Treatment
F-N	1199°C	Furn.	27	None
F-0	1199°C	Furn.	27	1149°C
F-4	1199°C	Furn.	27	1171°C
A-N	1199°C	Air	183	None
A-0	1199°C	Air	183	1149°C
A-4	1199°C	Air	183	1171°C
	1 Hr at temp			1 Hr at temp.

All the blanks were subsequently aged at 760°C for 16 hours.

Microstructural Image Analysis

Multiple high resolution scanning electron images of the six heat treat microstructures were quantitatively analyzed in order to obtain the γ' precipitate size distribution and volume fraction. The quantitative analysis was done using MATLAB software with built in image processing toolbox and user developed routines to enhance and threshold the images, apply masks to separate the particles of interest and perform statistical analysis.

Results

Composition Study

The hold time crack growth data for the Alloy 10 chemistry modification study are shown for both subsolvus and supersolvus conditions in Fig 1. As shown, the TDCG rates are not very sensitive to the modifications of the niobium to tantalum ratio or to the cobalt content within the range of chemical compositions investigated.

The subsolvus TDCG resistance curves fall within a tight range indicating no discernable trends as a function of the chemistry modifications evaluated. The supersolvus TDCG resistance of the seven chemical compositions was almost an order of magnitude better than that of their subsolvus counterparts. However, as was the case for the subsolvus compositions, the crack growth resistance of all the supersolvus compositions did not vary significantly. While the scatter is slightly greater than that of the subsolvus compositions, the effect of changing the niobium and tantalum content on TDCG resistance does not appear to play a major role in influencing hold time crack growth behavior. As shown in Fig 1, the composition for which all the niobium was removed and substituted for by tantalum falls in the mid-range of the measured crack growth resistance curves.

The data suggests that the change in grain size from ASTM 11.5 to ASTM 7.5 is a far more potent driver in its influence on TDCG behavior than the changes in composition. The main effect of decreasing the niobium content was to decrease the yield and ultimate tensile stresses for both the subsolvus and supersolvus conditions (Table III).

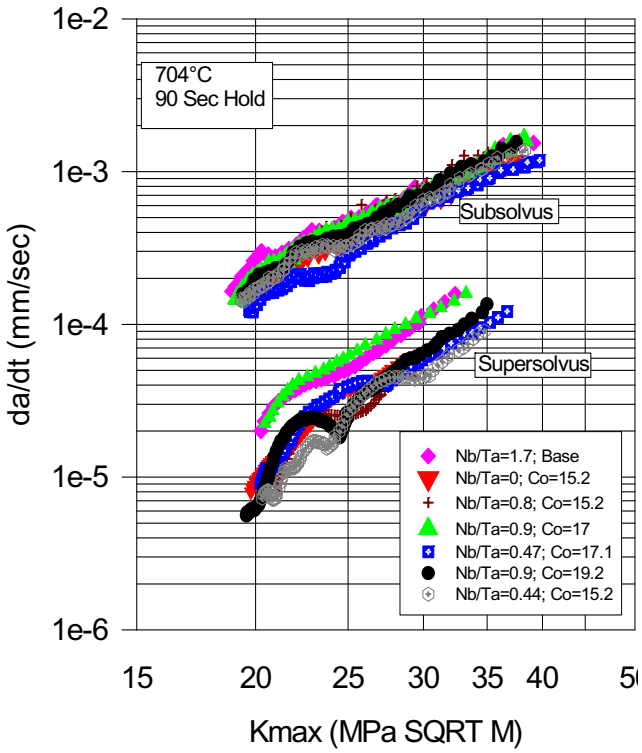


Fig 1. Hold crack growth resistance for the seven subsolvus and supersolvus compositions tested for the chemistry modification study.

Table III. Tensile properties at 704°C

CHEM MOD	SUBSOLVUS		SUPERSOLVUS	
	Y.S. MPA	T.S. MPA	Y.S. MPA	T.S. MPA
Nb/Ta=0.0;Co=15.3	1130	1351	1074	1379
Nb/Ta=0.4; Co=15.2	1167	1385	1116	1425
Nb/Ta=0.47;Co=17.1	1169	1402	1134	1445
Nb/Ta=0.8; Co=15.2	1154	1374	1104	1406
Nb/Ta=0.9; Co=17.0	1159	1422	1093	1405
Nb/Ta=0.9; Co=19.2	1132	1382	1092	1413
Nb/Ta=1.7; Co=15.7	1186	1398	1151	1462

Heat Treat Study

The grain sizes for all six heat treatments were measured and are shown in Table IV. Since the grain size is controlled by the original supersolvus solution step, there was very little difference between the individual heat

treatments. As shown in the table the measured grain sizes varied from ASTM 7.1 to ASTM 7.4.

Table IV. ASTM grain size – Heat treat study

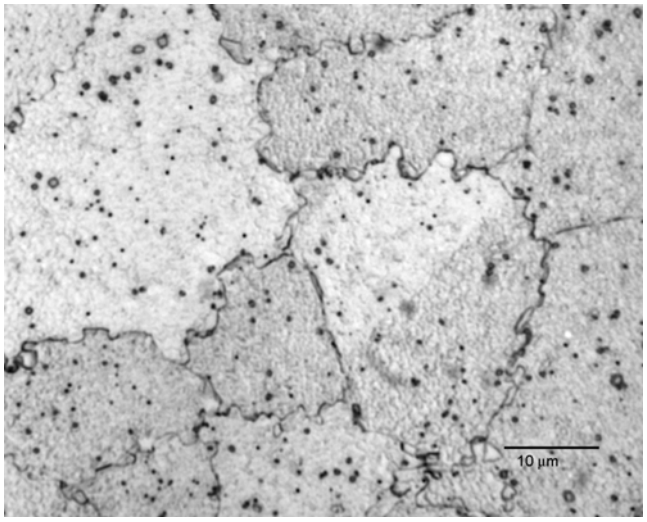
HEAT TREATMENT	ASTM GR. SIZE
Furnace Cool- No Resolution (**F-N**)	7.44
Furnace .Cool + 1149°C Resolution (**F-0**)	7.37
Furnace Cool +1171°C Resolution (**F-4**)	7.34
Air Cool - No Resolution (**A-N**)	7.38
Air Cool + 1149°C Resolution (**A-0**)	7.11
Air Cool – 1171°C Resolution. (**A-4**)	7.12

The tensile properties for the six heat treatments were also measured at a temperature of 704°C and are shown in Table V.

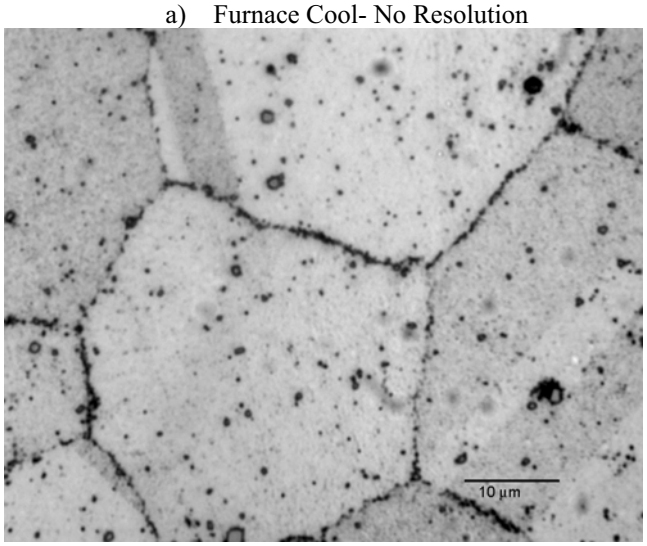
Table V. Tensile results - Heat treat study

HEAT TREATMENT	YIELD ST. (MPA)	TENSILE ST. (MPA)
Furnace Cool- No Resolution	1036	1329
Furnace Cool + 1149°C Resolution	1095	1350
Furnace Cool + 1171°C Resolution	1112	1369
Air Cool - No Resolution	1133	1413
Air Cool +1149°C Resolution	1099	1364
Air Cool +1171°C Resolution	1103	1380

Metallography was done in attempt to qualitatively assess the effect of varying the cooling rate on the formation of grain boundary serrations. The grain boundaries of the furnace cooled blanks were significantly more serrated than the air cooled microstructures as shown in Fig 2. The serrations are caused by the preferential growth of the large γ' at the grain boundaries. The subsequent resolutioning heat treatments may have decreased the level of serrations even though significant amount of serrations persisted.



a) Furnace Cool- No Resolution

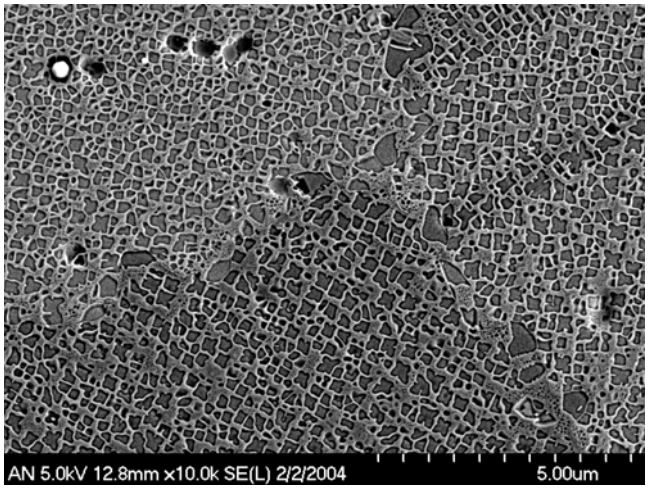


b) Air Cool – No Resolution

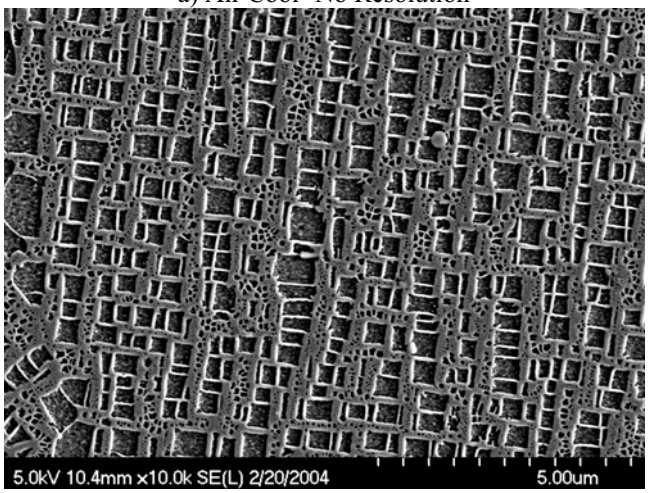
Fig 2. Magnitude of grain boundary serrations is much greater for the furnace cooled condition than air cooled.

The six individual heat treatments had a strong effect on the resulting cooling γ' morphology as shown in high resolution SEM images in Fig 3. Both the 1149°C and 1171°C subsolvus resolutioning treatments resulted in a microstructure consisting of a bi-modal distribution of cooling γ' precipitates for both furnace and air cooled specimens as shown in Fig 3b, 3c, 3e and 3f. In contrast, for both the furnace and air cooled samples without subsequent resolution treatment, only one type of the cooling γ' precipitates is present (Fig 3a and 3d).

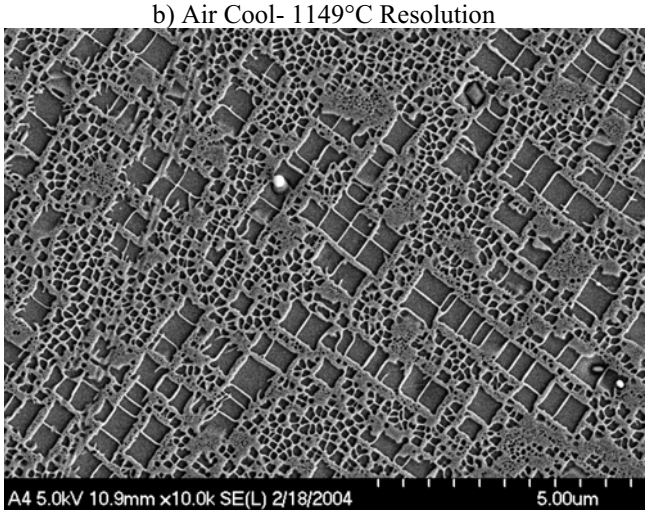
The results of the quantitative image analysis performed are shown in Fig 4. Size distribution of the cooling γ' for all the heat treatments is displayed in a cumulative probability plot.



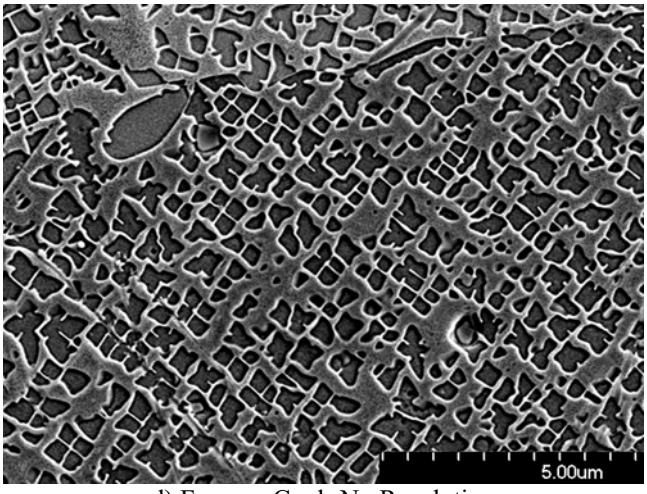
a) Air Cool- No Resolution



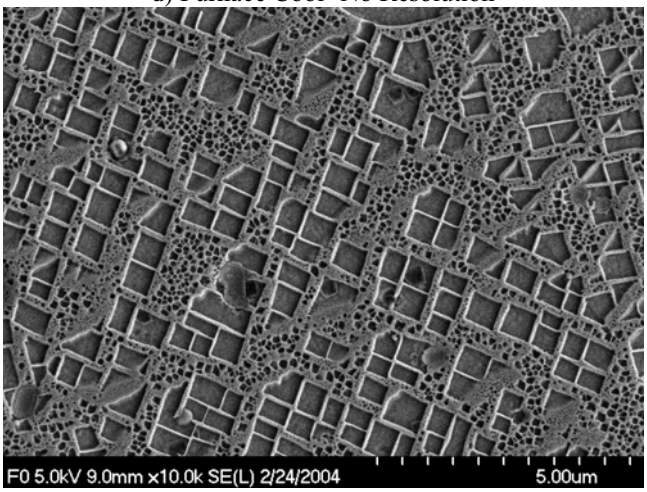
b) Air Cool- 1149°C Resolution



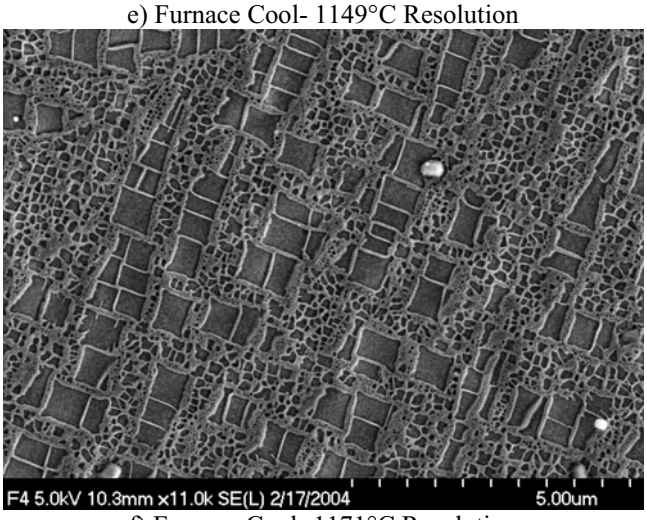
c) Air Cool – 1171°C Resolution



d) Furnace Cool- No Resolution



e) Furnace Cool- 1149°C Resolution



f) Furnace Cool- 1171°C Resolution

Figure 3. Effect of the individual heat treatments on cooling γ' microstructure for both air cooled and furnace cooled heat treatments.

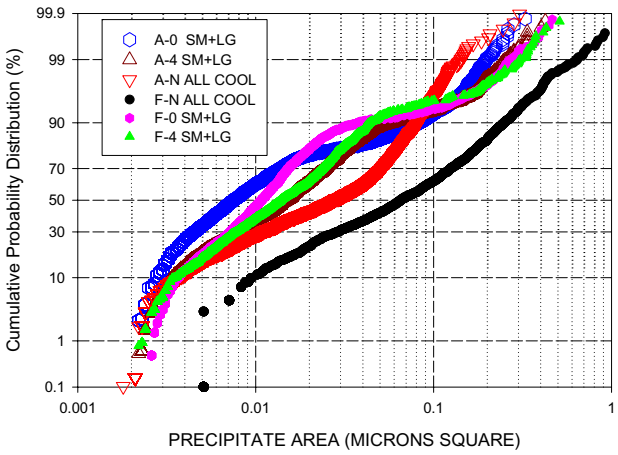


Figure 4. The results of the quantitative image analysis.

The bi-modal γ′ distribution consists of small precipitates created by the resolutioning step and larger, cuboidal precipitates left over from the original supersolvus heat treatment. The image processing was done to separate bi-modal distributions into their components. An example of the results obtained is shown in Fig 5.

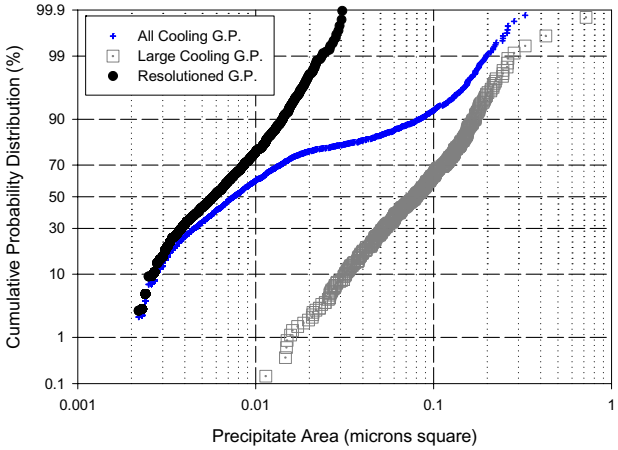


Fig 5. Separation of the bi-modal γ ′ distribution into its components. Air cool followed by the 1149°C resolution.

A closer examination of the cumulative probability distributions shown in Fig. 4 revealed that the size of the resolutioned γ′ to be more of a function of the resolutioning temperature than the initial cooling rate from the supersolvus solution. Thus, the similarity in size distribution for the small cooling γ precipitates for the two 1171°C resolutioned heats (A-4 and F-4). The two specimens given the 1149°C treatment (A-0 and F-0), also had similar distributions of the resolutioning γ ′, regardless of the supersolvus cooling rate.

The summary of the microstructurally characterized data obtained from the quantitative image analysis is shown in Table VI. It should be noted that at this point we have not as yet concluded the analysis of the aging γ ′ content.

Table VI. Results of the Cooling γ′ Image Analysis

Heat Treat	Supersolvus γ ′ mean size (sq. μm) and [% volume]	Resolution γ′ mean area (sq. μm) and [% volume]
F.C. only	0.064 [42%]	Not Applicable
F.C.+1149°C	0.13 [27%]	0.0096 [17%]
F.C.+1171°C	0.18 [25%]	0.012 [24%]
A.C. only	0.028 [53%]	Not Applicable
A.C. +1149°C	0.079 [37%]	.0059 [11%]
A.C.+1171°C	0.125 [25%]	.012 [26%]

Crack growth results

The heat treat study evaluated the role that the cooling rate from the supersolvus solutioning temperature and the subsequent subsolvus resolutioning steps have on the TDCG behavior. As shown in Fig 6, in contrast to the composition modification study, the variation in the heat treatments resulted in order of magnitude differences in the hold time crack growth rates.

The slowest crack growth rates were exhibited by the specimens which were not given any subsolvus resolutioning treatment, with the furnace cooled specimen having somewhat better crack growth resistance than the air cooled specimen. The resolutioning treatments substantially worsened the hold time crack growth resistance of Alloy 10. The 1171°C resolutioned specimens grew at crack growth rates approximately 5 times faster than the specimens which were not resolutioned. Significantly, the crack growth behavior for both air cooled and furnace cooled specimens was virtually the same. Finally, the 1149°C resolutioned specimens exhibited the worst TDCG resistance, again without significant difference between the air cooled and furnace cooled conditions.

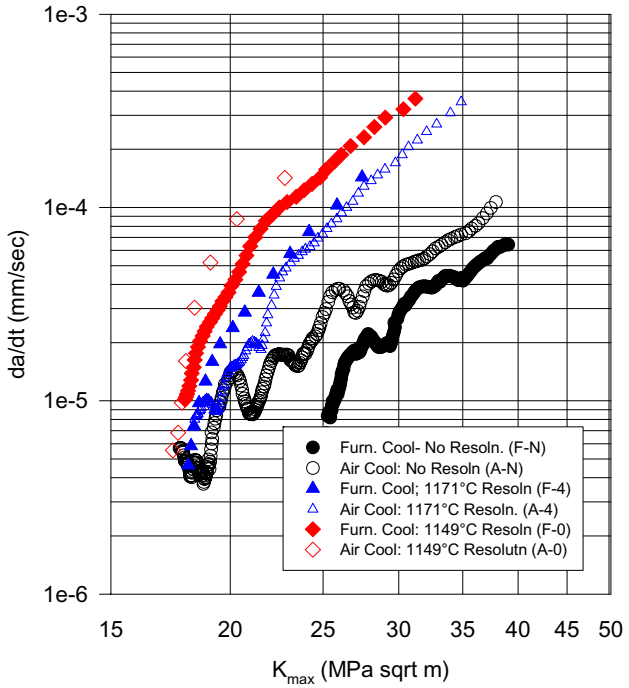
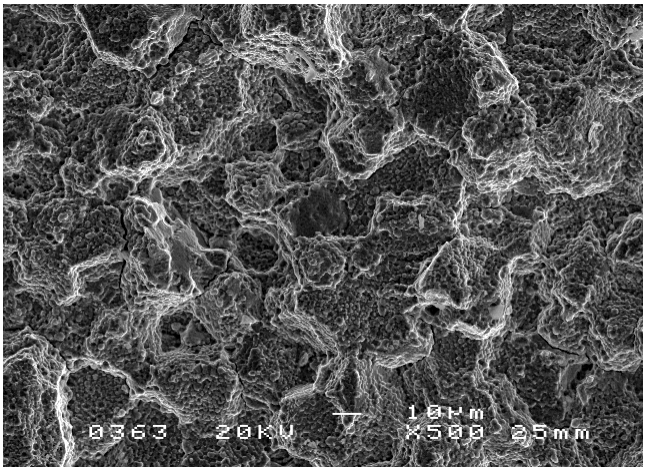


Fig 6. Hold time crack growth results for the heat treat study of Alloy 10.

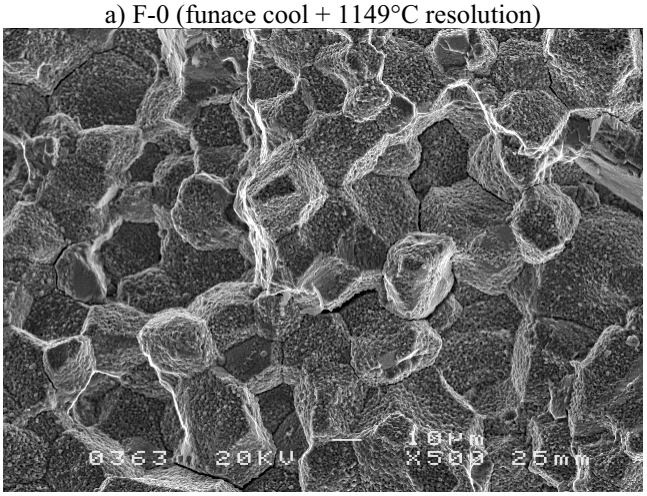
Fractography

All hold time crack growth tests exhibited a predominantly intergranular failure mode. All three furnace cooled specimens exhibited a rougher and more tortuous grain boundary failure appearance than their air cooled counterparts. The roughness of the furnace cooled specimens was due to the presence of large preferentially grown γ ′ precipitates at the grain boundaries during the slow cool from the supersolvus temperature.

Qualitatively, the degree of roughness or tortuosity did not correlate with the measured TDCG resistance. For example, Figure 7 is a comparison of fractographic failure features of specimen F-0 (furnace cool followed by the 1149°C resolutioning treatment) and the A-N specimen (air cool with no subsequent resolutioning treatment). As shown, both specimens failed by a mostly intergranular failure mode. The F-0 specimen (Fig 7a) has a considerably rougher looking failure appearance than the A-N specimen due to the presence of globular-like particles on the fracture surface. These globular particles are the large preferentially grown grain boundary γ ′ precipitates formed by the slow cooling rate from the supersolvus solution. In comparison, the fractograph of the A-N specimen (Fig 7b) reveals the grain boundary γ ′ precipitates to be considerably smaller and the failure mode significantly less tortuous. Yet the crack growth rates of the A-N specimen were at least 5 times slower than the F-0 specimen.



a) F-0 (funace cool + 1149°C resolution)



b) A-N (air cool - no resolution)

Fig 7. Comparison of the intergranular failure modes between furnaced cooled and air cooled specimens.

Discussion

Grain boundary species

The results of the present study show that a systematic reduction of the niobium content did not produce the desired effect of improving TDCG resistance as was postulated previously (2). The crack growth resistance was insensitive to the significant changes in the niobium, tantalum or cobalt content in the Alloy 10 composition. It is thus likely that this hypothesis falls in line with number of others which have in the long run been unsuccessful in identifying the grain boundary species which is "responsible" for the increase in the crack growth rates during sustained loading in oxygen rich environment. The other two compositional variables examined in this study, niobium to tantalum ratio and the cobalt content also did not exhibit a significant effect on TDCG behavior.

The continuous emphasis of research studies on the role that grain boundary phases and other species play as the major contributor towards accelerated crack growth under sustained loading has not yet paid any significant

dividends. While it is beyond reasonable doubt that oxygen increases the crack growth rates of high strength nickel based superalloys resulting in a mostly intergranular failure mode, the wide variations in TDCG resistance between different alloys and heat treatments have not been successfully explained by the use of arguments centered on the role of grain boundary species in the crack growth process.

The one variable which did have a very significant effect on TDCG resistance in the compositional study was the change in the grain size from subsolvus to supersolvus. The increase in the grain size resulted in an order of magnitude improvement in hold time crack growth resistance. These results are in agreement with results from many other studies which have shown an increase in grain size typically improves TDCG resistance.

Heat Treat Study

In contrast to the chemistry modification study, the heat treat study of the baseline Alloy 10 resulted in an order of magnitude differences in the measured hold time crack growth rates (Fig 6). Thus the modification of the cooling rate and the subsequent resolutioning treatments had a much greater effect on the TDCG behavior than compositional modifications. The question remains, by what mechanisms did the change in the microstructure due to the applied heat treatments influence hold time crack growth behavior?

In order to identify the mechanisms by which the heat treatments effect TDCG behavior, a quantitative analysis of the resulting microstructure is required. We decided to focus not only on the grain boundary species but to also quantify the size and distribution of γ' throughout the microstructure since the previous focus on grain boundaries has not produced the desired results.

Effect of cooling γ' on TDCG

A close relationship exists between the measured crack growth rates and the type of heat treatment applied (Fig 6). The samples which received the same resolutioning treatment, have very similar crack growth resistance, regardless of the initial cooling rates from the supersolvus solution temperature. This similarity in crack growth resistance as a function of the resolutioning treatment suggests that the microstructure produced by this particular heat treat step has a strong influence on TDCG behavior. For specimens not given the subsolvus resolutioning treatment, the cooling rate from the supersolvus solutioning temperature does have a significant effect on crack growth resistance. The considerably slower cooling rate of the furnace cooled only specimen (F-N) resulted in superior hold time crack growth resistance in comparison to the air cooled only specimen (A-N) which was cooled at a much faster rate.

We will show next that the hold time crack growth behavior correlates very well with the size distribution of the cooling γ' precipitates for all six heat treatments examined in this study. A qualitative relationship between the cooling γ' size and crack growth resistance can be discerned by comparing the cumulative precipitate size distribution to the measured crack growth rates (see Figs 4 and 6). From the evaluation of these figures it can be discerned that the heat treatments which increased the size of the cooling γ' precipitates exhibited an improvement in the crack growth resistance.

Since the resolution heat treat steps produced the largest change in the hold time crack growth rates, we decided to narrow the analysis to the resolution γ' size distribution for the four resolutioned specimens and compare it to the two specimens which were not given the resolution treatments. The qualitative results are shown in Fig 8. The trend is now more easily discerned: the bigger the cooling γ' size, the better is the resistance to hold time crack growth. As shown in Fig 8, the furnace cool with no resolutioning produced the largest mean γ' size and the best crack growth resistance. The second largest mean precipitate size was a result of the air cool with no subsequent resolution heat treatment which resulted in the second best crack growth resistance.

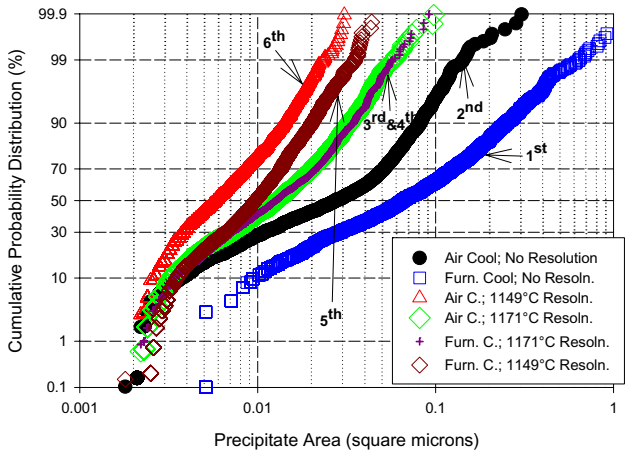


Fig 8. Relationship between the cooling γ' distribution and ranking of hold time crack growth resistance for the six heat treatments evaluated.

In case of the resolution heat treatments which produced a bi-modal distribution of the cooling γ', the results indicate that the smaller cooling γ' precipitates created by the subsolvus resolutioning steps have the most significant influence on hold time crack growth behavior. Thus, the 1171°C resolutioning treatments produced almost identical distributions of the smaller γ' regardless of the initial cooling rate. The result was an almost identical crack growth resistance, Fig 6. This trend also is in place for the specimens subjected to the 1149°C resolutioning treatment.

These heat treat step produced the smallest resolutioning γ′ and the fastest crack growth rates. Again the resolutioning γ′ size distribution and crack growth resistance were fairly similar for both furnace and air cool specimens, Fig 8.

In order to obtain a more quantitative relationship between the precipitate size and the crack growth rate, the mean cooling γ′ size for all the heat treatments was plotted versus a crack growth rate measured at a Kmax level of 25 MPa√m. The results, plotted in Fig 9, show a clear relationship between these two variables. This relationship extends over a wide range of the cooling γ′ size.
Further work is being conducted in an attempt to include the other types of γ′ precipitates (such as aging and grain boundary) and their respective volume fractions in order to obtain a more comprehensive relationship between γ′ and hold time crack growth behavior for this alloy.

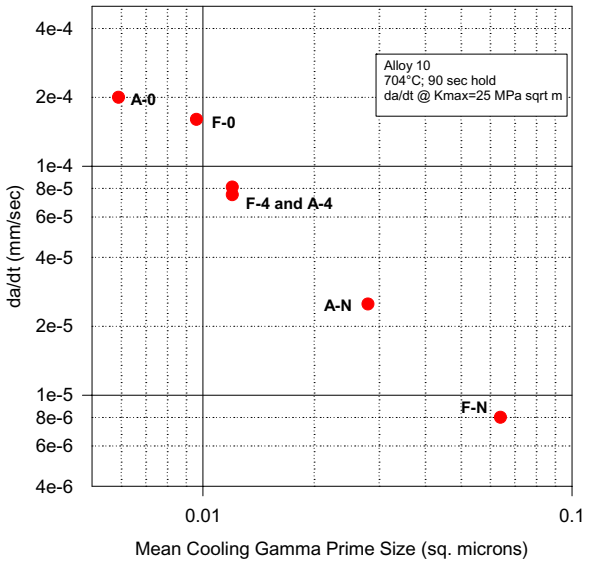


Figure 9. Relationship between the mean cooling γ′ precipitate size and crack growth behavior for the six heat treatments evaluated. Crack growth rates at a Kmax of 25 MPa √m.

What is the mechanism by which the size and volume fraction of the precipitates throughout the microstructure become the controlling variables for hold time crack growth resistance? The strong relationship which exists between the size distribution of the cooling γ′ and hold time crack growth behavior for Alloy 10 points to the creep activated stress redistribution in the crack tip region as a key variable influencing hold time crack growth resistance. It has been suggested previously (7) that at the temperature at which the hold time crack growth tests were conducted, 704°C, most of the stress relaxation takes place through a dislocation creep mechanism which occurs within the grains, with little or no contribution from the classical grain boundary sliding mechanisms. Thus the volume fraction and size distribution of the strengthening precipitates may play a critical role in controlling creep deformation and the associated stress relaxation at the test temperature. If this scenario is true, then the reduction of the crack driving force during hold times should be a function of the strengthening precipitate size distribution which could explain the effect of the varied heat treatments on hold time crack growth resistance.

The possible role of crack tip stress relaxation in hold time crack growth behavior has been postulated by others (5,6,8). The most convincing work was by done by Molins et.al. (5). They showed that for a hold time test conducted in vacuum, the introduction of oxygen for the same increment of time at different stages of the dwell has a very significant effect on crack growth resistance. The introduction of oxygen for the same time increment resulted in much faster crack growth rates if oxygen was introduced early in the hold cycle as opposed to later on. They attributed their results to crack tip stress relaxation which reduces the crack driving forces. Thus, when the embrittling agent is introduced at the latter stages of the hold cycle, considerable crack tip stress relaxation has already taken place and the resulting crack growth rates are lower. Our results are in general agreement with this scenario provided that it can be shown that the precipitate size distribution affects the stress relaxation behavior.

It should be pointed out that our work in quantifying the distribution of γ′ is not yet complete. We are currently working on characterizing the size distribution of the aging precipitates. While it is clear that their volume fraction is relatively small, their effect on the stress relaxation behavior is unknown but may be significant. We are also in the process of conducting stress relaxation tests in order to simulate the proposed crack tip behavior.

Summary

A study was performed to determine the variables which influence hold time crack growth resistance of Alloy 10, a new P/M superalloy. In a well controlled study, both the effect of compositional changes and variation in heat treatments were investigated. The results indicate that significant changes in the alloy's niobium, tantalum and cobalt content did not have an appreciable effect on hold time crack growth resistance as has been postulated by others. This was true for both subsolvus as well as supersolvus microstructure. The change in the grain size had a much stronger influence on crack growth than the compositional variations.

In contrast to the composition study, the heat treatments evaluated produced up to an order of magnitude changes in the crack growth resistance without any significant changes in the grain size. The addition of a subsolvus resolutioning step proved to be highly damaging to the crack growth

resistance regardless of initial cooling rate or grain boundary γ′ morphology. For specimens not given the subsolvus resolutioning treatment, the slower cooling rate from the supersolvus solutioning temperature produced a significant improvement in hold time crack growth resistance.

Quantitative image analysis was performed to analyze the microstructural features produced by each heat treatment. It was found that the cooling γ′ precipitate size distribution is closely related to the measured crack growth behavior. The larger the mean size of the cooling precipitates, the better is the resistance to hold time crack growth.

It was hypothesized that the size and the distribution of the precipitates play an important role in determining the extent of crack tip relaxation which takes place during hold time. The differences in the stress relaxation rate influence the crack driving forces and thus crack growth resistance. Superior ability to relax the crack tip stresses improves hold time crack growth resistance.

The results of this study suggest that the emphasis of numerous other researchers over the years on the characterization of grain boundary species which can control hold time crack growth resistance was misplaced. The current results indicate that the size distribution of the strengthening precipitates and the materials ability to relax the crack driving force is the major factor in controlling hold time crack growth resistance.

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