

DEVELOPING DAMAGE TOLERANCE AND CREEP RESISTANCE IN A HIGH STRENGTH NICKEL ALLOY FOR DISC APPLICATIONS

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Abstract

The development of an affordable nickel disc alloy that shows the required combination of strength, damage tolerance and creep resistance continues to challenge materials engineers. Rolls-Royce has developed the alloy known as RR1000, which when processed to produce a fine grain microstructure, shows at least a 25°C increase in temperature capability over the current disc alloy, 720Li, and has an equivalent crack growth resistance to coarse grain Waspaloy. This paper discusses the steps that were taken to attain these mechanical properties, and also examines the potential to increase temperature capability by a further 30°C through the development of a coarse microstructure in selected areas of disc forgings.

Introduction

To meet the demand for reduced acquisition and life cycle costs in civil aviation, modern gas turbine engines are expected to achieve increasingly higher levels of fuel economy, reduced NOx emissions and noise, and with reduced module weight. These challenges inevitably necessitate higher overall pressure ratios, and higher compressor discharge and turbine entry temperatures. As such, disc rotors in the High Pressure compressor and turbine must accommodate higher temperatures and stresses. This places significant demands on the high strength nickel alloys that are used for these critical components, particularly as loss of integrity can threaten the safety of the aircraft and passengers. If acceptable component lives cannot be achieved using existing alloys, then a new disc alloy will need to be introduced. Rolls-Royce has developed the alloy known as RR1000 to replace alloy 720Li, which is the best of the current list of nickel disc alloys that is available to the Company.

Alloy Composition

Traditionally, alloys have been developed by empirical methods that have involved much iteration and have relied on many years of experience and expertise. These methods have been based on binary or tertiary equilibrium phase diagrams. Although this approach has received some notable successes, it is very time consuming and costly, and may not produce an optimised alloy chemistry. In recent years, Rolls-Royce has chosen to use analytical methods such as phase diagram modelling to assist in the process of alloy chemistry definition.

Advances in thermodynamic modelling have allowed the accurate prediction of complex phase equilibrium in real engineering alloys¹. Rolls-Royce has developed this methodology for use with nickel based superalloys² and is able to predict accurately, the

existence of stable phases at fabrication and operating temperatures (Figure 1)³. The use of thermodynamic modelling also enables the alloy designer to predict the outcome of a set of conditions that deviate from equilibrium, as well as fundamental alloy responses, such as anti-phase boundary energies (APB)⁴, that reflect the potential creep behaviour and strength of alloys.

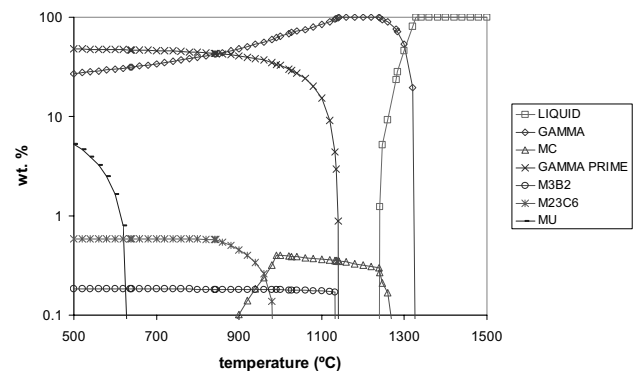


Figure 1. Phase equilibria for RR1000³ generated using JMat Pro.

The database established for the calculation of phase equilibria has been used to predict the formation of undesirable topologically closed packed (TCP) phases during prolonged exposure at elevated temperature². In the past, the phase computation (PHACOMP) method⁵ has been used as a guide to avoid the formation of the deleterious σ phase. The approach relies on calculating the average electron valency number, N_v , for the alloy and defining an alloy specific number, below which, the alloy is deemed to be 'σ safe'. Although this approach has been used extensively, it does not allow for the true complexity of TCP formation to be taken into consideration and ignores other deleterious TCP phases such as μ , χ , P and R. Furthermore, PHACOMP, gives no information on temperature ranges of stability, phase boundaries, segregation effects or potential operating windows. The determination of metastable equilibrium conditions from thermodynamic modelling overcomes most of these limitations. It has thus been used as a starting point for experimental work to characterise the morphology of σ in exposed material and to determine the precipitation kinetics of σ in development alloys⁶. Whilst the precipitation of σ , in a particular morphology and in a sufficient concentration, will degrade mechanical properties, coarsening of tertiary γ' during exposure at elevated temperature may have an equally damaging effect³.

The composition of RR1000 (Table I) was defined through the use of thermodynamic modelling and an understanding of the role and

benefits of elemental additions and interactions on the mechanical properties of superalloys⁷. The influence of the major alloy

constituents on behaviour is discussed in the following text.

TABLE I. Composition of 3 Variants of RR1000 in Weight Percent⁷.

Alloy	Co	Cr	Mo	Ti	Al	Ta	Hf	C	B	Zr	Ni
#1	18.5	15	5	3.6	3	2	0.75	0.03	0.02	0.06	Bal.
#2	18.5	15	5	3.6	3	2	0.5	0.03	0.02	0.06	Bal.
#3	18.5	15	5	3.6	3	2	0	0.03	0.02	0.06	Bal.

It has been established that strength is derived from solid solution strengthening and γ' precipitates^{8, 9}, the degree of strengthening being a function of the size and concentration of solute atoms and the volume fraction and size of γ' precipitates. The contribution to strength of lattice misfit strain from solute atoms, and the perceived patent restrictions in using tungsten have resulted in high levels of aluminium, molybdenum, chromium and cobalt in RR1000. Principles of alloy design⁸ indicate that a large misfit between coherent γ' precipitates and γ matrix is beneficial for low temperature strength. Conversely, a low misfit is generally preferred for high temperature creep resistance. As disc applications require low temperature strength that is maintained at temperatures in excess of 650°C, these alloy design principles appear to conflict. In designing RR1000, the objective was to develop a misfit that contributes to strength, but to ensure that γ' remained coherent.

Particular attention was paid to the influence of chemistry on crack growth behaviour as previous nickel disc alloys have struggled to exhibit the required balance between strength and fatigue crack growth resistance. Alloys such as IN100 and Rene 95, for example, that have volume fractions of γ' as high as 50-60 % show high tensile strength but suffer from very fast rates of crack growth. Consequently, in developing RR1000, the aim was to achieve the desired levels of strength at a reduced volume fraction of γ' . This involved manipulating the solute atoms that partition to the γ' , specifically aluminium, titanium and tantalum. In practice, the titanium content was adjusted to reflect the necessary levels of tantalum and aluminium. Whilst tantalum increases strength, it was found to increase crack growth rate and as a result was limited to 2.15 wt.%. Similarly, the aluminium content was restricted to 3.15 wt.%, since additional aluminium would increase the volume fraction of γ' and lead to chromium enrichment of the γ matrix and thereby increase the propensity for TCP phase formation. The titanium content was therefore increased to give a 45 % weight fraction of γ' at 725°C.

Encouraging planar slip was also considered to be important if low crack growth rates were to be achieved. In this heterogeneous form of deformation, it is understood that dislocations cut precipitate particles rather than loop around them, as in more homogeneous slip. This is advantageous as reversal of slip during the unloading component of a fatigue cycle is relatively easy compared to slip reversal on more homogeneous, wavy slip paths¹⁰. As such, there is less damage accumulation. The ability to inhibit cross slip also has a beneficial effect on creep behaviour. Since a low stacking fault energy (SFE) is known to promote planar slip, the cobalt content in the alloy was optimised to produce a minimum value of SFE. In addition, cobalt is considered to reduce the solubility of aluminium and titanium in γ ,

which maximises the partition of these elements to the γ' , and to reduce the γ' solvus temperature.

As well as oxidation resistance, chromium was found to reduce the rate of crack propagation at elevated temperature. However, experience has shown that a level of 18 wt.% in alloy 720 causes excessive precipitation of σ and μ phases after relatively short exposure times above 700°C^{2, 11}. Since these phases are known to degrade the creep properties of nickel disc alloys^{12, 13}, a balance between improved fatigue crack growth resistance and reduced TCP phase formation was struck at around 15 wt.% chromium. Consequently, chromium is critical in achieving a balance between strength and damage tolerance. This was made possible by imposing a limit on the molybdenum content at the expense of potential gains in high temperature strength and creep resistance. The mechanism for the observed improvement in crack growth rates at elevated temperature is not fully understood. One explanation may lie in the ability to decorate grain boundaries with $M_{23}C_6$ ($M = Cr, Mo$) and M_6C ($M = Mo$) carbides in order to retain transgranular crack growth. Alloy variants with reduced levels of chromium, below 15 wt.%, were very stable but showed mixed transgranular and intergranular cracking and higher rates of crack growth. It was discovered that these only had MC ($M = Ta, Ti, Hf$ if added) carbides at the grain boundaries.

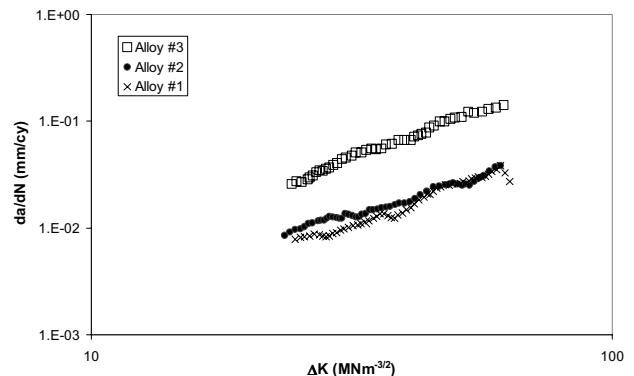


Figure 2. Crack growth rate data at 725°C for hafnium bearing and hafnium free RR1000, generated from fatigue cycles that have a 120 s dwell at maximum load and a stress ratio of 0.1¹⁴. The material showed an average grain size of ASTM 10-9 (11-16 μm).

The minimum levels of boron, carbon and zirconium that were required to produce enhanced high temperature properties, notably creep, were added. Boron was also considered to be of benefit to the fatigue performance of the alloy. It was found that the addition of hafnium enabled significant improvements in the crack growth¹⁴ (Figure 2) and creep resistance of RR1000, and produced moderate increases in tensile strength, since it partitions to γ' .

Hafnium also forms a primary carbide that remains stable at low temperatures, which is considered to prevent excessive $M_{23}C_6$ and M_6C formation at grain boundaries. However, these benefits need to be assessed against the challenges in processing hafnium-bearing alloys and in managing the risk of micron-sized hafnia particles, either as small agglomerations or attached to the periphery of other melt anomalies.

Having determined the chemistry of this new alloy, decisions then had to be made concerning the processing route and heat treatment schedule. Both of these steps have a profound effect on the mechanical properties and the material cleanliness of the alloy.

Production of Billet for Large Diameter Disc Forgings

Although billet for large-scale forgings of alloy 720Li are being produced via the cast and wrought route¹⁵, powder metallurgy is being considered for increasingly complex nickel disc alloys, such as RR1000, to avoid the risk of elemental segregation and unacceptable levels of carbonitride stringers. It is recognised that the use of powder processing will increase the price of billet material above the level for cast and wrought 720Li, and requires specific methods for probabilistic design/lifting and for ensuring that a consistent level of cleanliness is maintained in production material. Whilst the powder route is currently preferred, it is considered that large diameter billet can be produced from cast and wrought processing but is unlikely to offer a cost saving over the powder product.

It is well documented^{16, 17} that the presence of inclusions, which are picked up during powder processing, can severely limit the fatigue performance of the alloy. Over the last decade, the cleanliness of argon gas atomised powder has improved considerably through the elimination of organic seals and the use of progressively finer sieve sizes to screen out non-metallic inclusions. This has been made affordable by the development of nozzles for high yield powder production¹⁸. RR1000 powder is now being produced using these nozzles and argon gas atomisation, and is screened to a final sieve size of -270 mesh (53 μm).

Following consolidation of canned powder by hot isostatic pressing or hot compaction, it is the industry practice to extrude the cans to a reduction ratio greater than 4.5:1. Nominally 12" diameter RR1000 billet, which is preferred for large turbine discs, is extruded to this specification. The purpose of the extrusion stage is to break down any prior particle boundaries and to produce a fine grain size that is typically smaller than ASTM 12 (6 μm). This structure is ideal for inspection of the billet and subsequent superplastic deformation on isothermal forging. Given the limited number of sources for extrusion of large diameter billet, smaller reduction ratios¹⁹ and alternative consolidation methods are being evaluated^{19, 20}. These are particularly attractive for small powder campaigns as they offer cost savings.

Forging and Heat Treatment

Whilst the hot die route can be used for some forging operations on RR1000, the majority of the necessary hot work required to produce the desired finished shape and grain structure is

accomplished by isothermal forging. As lower strain rates are used during deformation, isothermal forging currently offers greater control of structure and the ability to forge quite complex, 'near-net' shapes that enable significant reductions in material input weight and machining time²¹.

As grain size and the size of secondary and tertiary γ' precipitates determine mechanical properties, forging and solution treatment temperatures, quench rates from solution treatment, and ageing temperature and time need to be considered carefully so that heat treated forgings meet expectation. Strength and resistance to low cycle fatigue have been optimised for disc applications by developing a fine grain size through sub-solvus heat treatment, and by quenching in oil to maximise cooling rates, and thereby, minimise the size of secondary (and tertiary) γ' . This approach tends to produce less appealing crack growth resistance. There is some evidence to suggest that fast quenching leaves grain boundary regions free of secondary and tertiary γ' , which increases the propensity for intergranular cracking and very fast rates of crack growth under creep-fatigue conditions²². A more homogeneous distribution of these γ' particles and consequently, a significantly superior resistance to elevated crack growth, is produced from slower rates of cooling, such as from air-cooling.

There are other concerns for alloys that have a high volume fraction of γ' . Rapid quenching in oil from solution treatment can cause quench cracking in thin sections or at corners of complex forging shapes. Significant internal residual stresses also develop from cooling rate gradients that can occur in forgings due to differences in section size. These have been found to produce distortion during machining, which increases the time and cost for component manufacture. Residual stresses can be relieved by heat treatment but inevitably at temperatures that are higher than the ageing temperature, which coarsens the tertiary γ' and leads to a degradation of mechanical properties.

There is added incentive to minimise quenching stresses in powder-processed alloys due to the occurrence of ceramic inclusions. Although the majority of these inclusions are small, they must remain uncracked if the finished component is to realise the full fatigue life potential of the alloy. Evidence from cyclic rig tests on subscale powder discs²³ has shown that having considered quenching residual stresses after ageing, discs produced from fan air-cooled forgings exhibit a greater fatigue life than those machined from oil quenched forgings. For this and the other reasons given above, RR1000 forgings have been fan air-cooled after solution treatment. However, as the strength of the alloy is not fully optimised, the design envelope for discs may increase to provide the required tolerance to counter high stresses that may arise from unforeseen shaft over-speed events.

The decision not to oil quench has limited the solution heat treatment of RR1000 to temperatures below the γ' solvus. This is apparent from Figure 3. It shows that the proof and tensile strength of material that has been taken from a large section of a forging, such as the bore/cob, are considerably reduced if the material has a coarse microstructure. Consequently, work has been directed towards developing a process window for a near-solvus solution treatment, to produce an average grain size of ASTM 12 to 9 (6-16 μm), which would enable an acceptable balance to be achieved between tensile strength and crack growth resistance. Even within this relatively small range of target grain sizes, it has been found that a significant variation in crack growth

rate can occur at temperatures and conditions in which oxidation and time dependent deformation result in a mixed transgranular and intergranular crack growth mechanism²⁴. Figure 4 illustrates this observation. It shows that a reduction in grain size from ASTM 10-9 (11-16 μm) to ASTM 12-11 (6-8 μm) has increased the rate of crack growth at 650°C beyond expected levels of scatter.

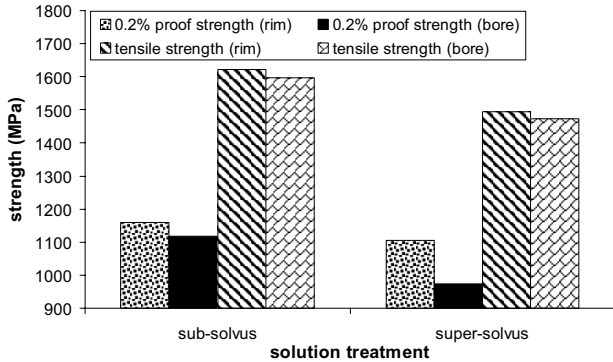


Figure 3. The influence of solution treatment temperature on the room temperature proof and ultimate tensile strength of Alloy #1 (in the fully heat treated condition). Fan air-cooling was used to quench large-scale forgings after solution treatment.

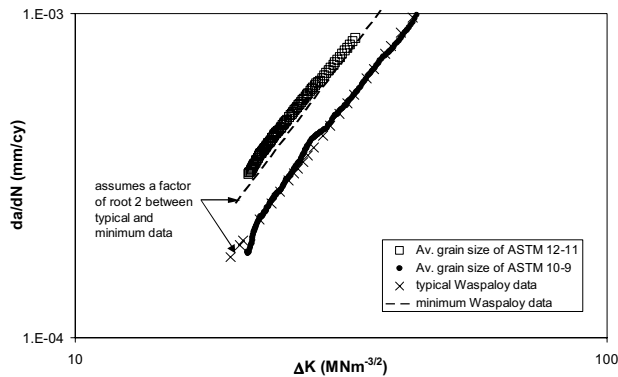


Figure 4. Crack growth rate data at 650°C for Alloy #1 and Waspaloy (grain size of ASTM 8-4) that have been generated at a cyclic frequency of 0.25 Hz and a stress ratio of 0.1. The RR1000 data indicate that a relatively small change in average grain size, under creep-fatigue conditions, can result in significant changes in crack growth rate.

Although the rates of crack growth in RR1000 are equivalent to those of coarse grain Waspaloy for material that exhibits an average grain size of ASTM 10-9 (11-16 μm), there is limited scope for increasing the average grain size beyond ASTM 12-11 (6-8 μm) by applying a near-solvus solution treatment temperature following a sub-solvus forge. This is due to primary γ', and to a lesser extent MC carbides, that pin grain boundaries at solution treatment temperatures below the γ' solvus (Figure 5). The required grain size has been achieved via a near-solvus forge and solution treatment, but the process window is considered to be narrow for some complex, large-scale forgings. In such forgings, an average grain size of ASTM 12-11 (6-8 μm) is produced consistently by applying the former, more conventional route.

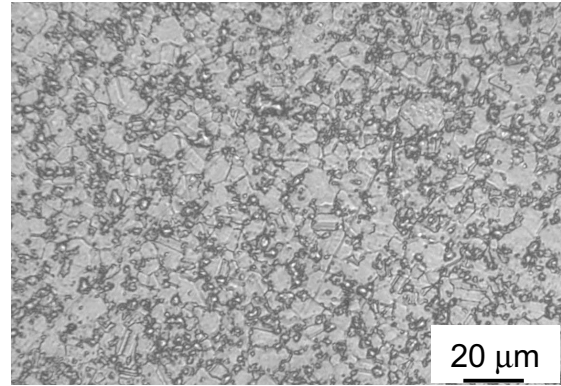


Figure 5(i). Optical micrograph showing the microstructure of Alloy #2 after near-solvus solution treatment at 1125°C.

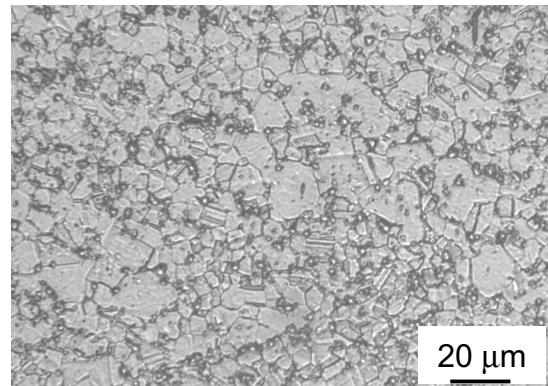


Figure 5(ii) Optical micrograph showing the microstructure of Alloy #2 after near-solvus solution treatment at 1131°C. The increase in solution treatment temperature from (i) has reduced the volume fraction of primary γ' and has increased the As Large As (ALA) grain size but has had a limited effect on the average grain size.

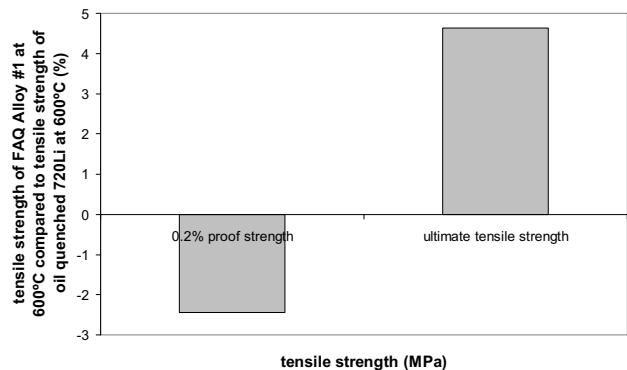


Figure 6. Chart showing the tensile strength of Alloy #1 at 600°C. The proof strength is lower than that for oil quenched, cast and wrought 720Li as fan air quenching (FAQ) was used to cool forged RR1000 material after solution treatment. The RR1000 and 720Li material showed an average grain size of ASTM 10-9 (11-16 μm), ASTM 11-10 (8-11 μm) respectively.

Near-solvus solution treatment, followed by ducted fan air-cooling has also delivered the required levels of tensile strength (Figure 6) and creep resistance (Figure 7). Given that fan air-cooling generates lower rates of cooling than oil quenching, it is not unexpected that the proof strength of fan air-cooled RR1000 is lower than that for oil quenched 720Li, although 720Li has a slightly reduced volume fraction of γ' , being 44 % at 725°C. Further improvements in the strength and creep resistance of RR1000 are now possible, however, as a result of recent developments in processing technology. These are described in a subsequent section.

Unless forgings are for components that are to be joined by inertia welding, RR1000 is aged at 760°C for 16 hours. For welded components, the heat treatments after quenching and welding have been designed to optimise the relief of residual stresses whilst ensuring that the size of tertiary gamma prime²⁵, which is produced from the specified ageing heat treatment, is maintained.

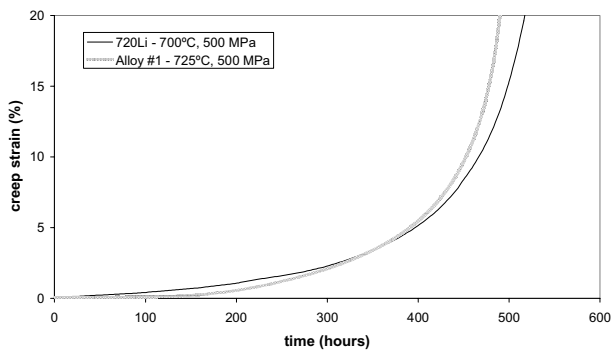


Figure 7. Creep strain data for oil quenched, cast and wrought 720Li and fan air-cooled RR1000 from a stress of 500 MPa. The RR1000 data were generated at 725°C, the data for 720Li at 700°C. The average grain size for the RR1000 material was ASTM 12-11 (6-8 μm), and ASTM 11-10 (8-11 μm) for the 720Li material.

Super Cooler

	A = - 473
	B = - 276
	C = 0
	D = 276
	E = 551

Maximum Residual Stress (Mpa) = 590
 Minimum Residual Stress (Mpa) = - 473

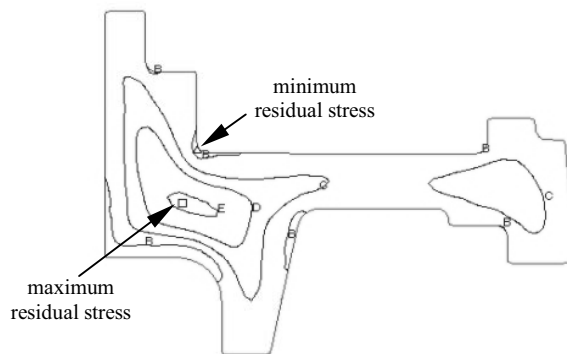


Figure 8(i). Predicted values of residual stress for a generic disc forging that is quenched using the Super Cooler.

Oil Quench

	A = - 1592
	B = - 861
	C = 0
	D = 861
	E = 1626

Maximum Residual Stress (Mpa) = 1626
 Minimum Residual Stress (Mpa) = - 1592

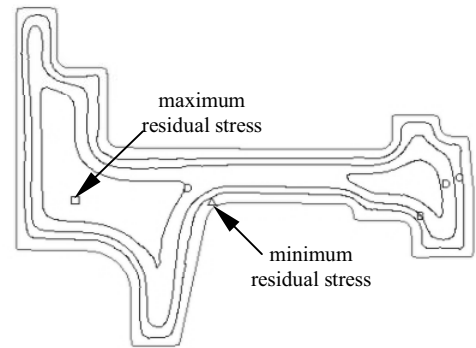


Figure 8(ii). Predicted values of residual stress for a generic disc forging that is quenched by immersion in oil. An identical solution treatment was considered in calculating the predicted values of residual stress that are shown in (i) and (ii).

Recent Developments in Heat Treatment and Quenching of Forgings

The Super Cooler is a compressed air facility that has been developed by Ladish²⁶. Since quenching is achieved by the controlled delivery of compressed air over the surface of the forging, computer-based analytical models can be used to simulate the process. This ability allows the process to be optimised to produce the fastest possible cooling rates throughout the forging and to accommodate differences in section size to minimise cooling rate gradients and hence residual stresses. Results from large development forgings have shown that the use of the Super Cooler has increased the proof strength of RR1000 by 4 % but without the high levels of residual stress that are associated with oil quenching. Figure 8 shows predicted values of residual stress for a generic disc forging that is given an identical solution treatment but which has been quenched using the Super Cooler and by immersion in oil. The residual stresses that result from the Super Cooler are approximately one-third of those that are generated from oil quenching.

Although the data in Figure 3 indicate that solution treatment above the γ' solvus temperature is unlikely to generate the required strength in the bore/cob of discs, the difference in proof strength between a fine and coarse microstructure becomes less significant at the rim of the forging as the smaller section size enables higher rates of cooling to be achieved throughout the section. At temperatures in excess of 650°C, at which the rim of discs are required to operate, the difference in ultimate tensile strength between material that exhibits a fine and coarse microstructure becomes insignificant. Furthermore, the ability to

produce a coarse grain size in the rim can increase the creep resistance of the disc by over 30°C (Figure 9) and further improve crack growth resistance, at temperatures at which oxidation and time dependent deformation accelerate crack growth.

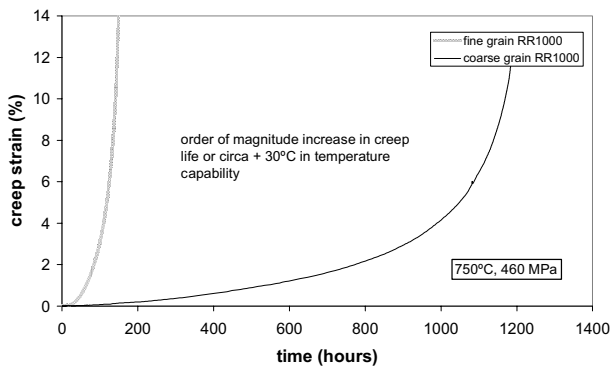


Figure 9. The influence of solution heat treatment temperature on the creep behaviour of Alloy #1 (in fully heated condition). Fan air-cooling was used to quench forged material after solution treatment.

A complete strain, strain rate and temperature map for RR1000 has been established to predict grain growth and to avoid conditions that may give rise to abnormally large grains during a super-solvus heat treatment. Evidence from a large intermediate pressure (IP) turbine disc forging (Figure 10) has indicated that a coarse grain structure (Figure 11-12) can be produced without critical grain growth.

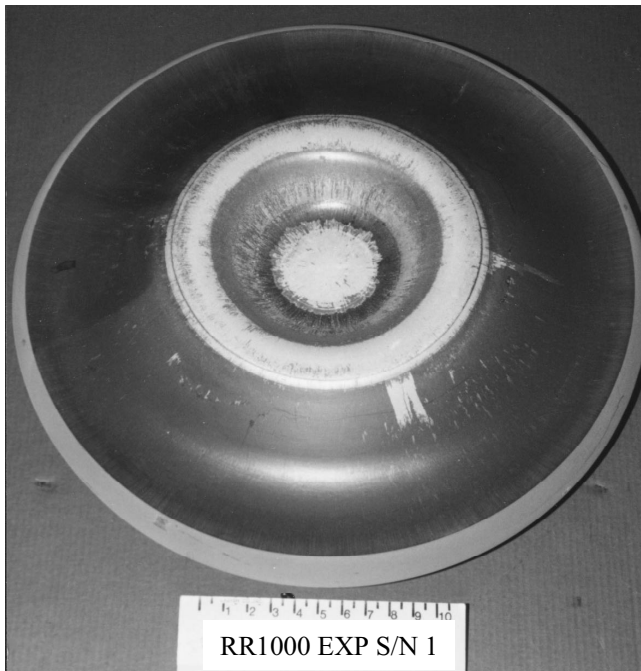


Figure 10. An intermediate pressure (IP) turbine disc forging made from Alloy #1.

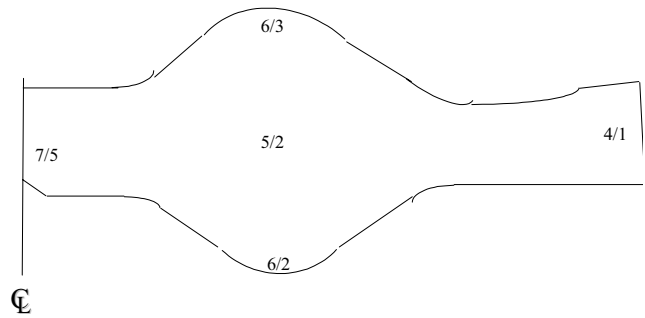


Figure 11. Average and As Large As (ALA) grain sizes obtained from a forged and super-solvus heat treated RR1000 IP turbine disc (Figure 10).

Developments in heat treatment technology²⁷ to produce dual microstructure forgings are now being validated for production use. These will enable such increases in temperature capability to be realised for alloys that can be super-solvus heat treated to produce a uniform coarse grain size.

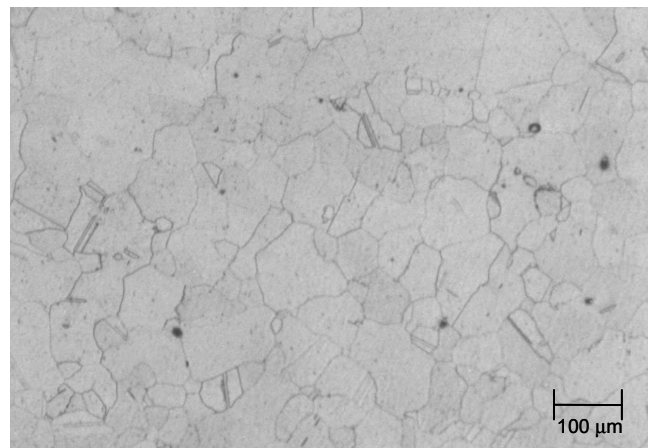


Figure 12. The microstructure of Alloy #1 after super-solvus solution treatment, and age.

Conclusions

The composition of RR1000 has been defined so that the minimum volume fraction of γ' achieve the required levels of strength. This approach has been used to encourage damage tolerance. Specific quantities of chromium, cobalt and hafnium, and the ability to decorate grain boundaries with specific amounts of $M_{23}C_6$, M_6C and MC are considered to be critical in developing good crack growth resistance at elevated temperature, particularly under creep-fatigue conditions. The levels of aluminium, chromium and molybdenum have also been manipulated to avoid excessive TCP formation at the projected time at peak temperature in service. Near-solvus solution heat treatment and fan air-cooling of forged material from consolidated powder billet has established an acceptable balance between tensile strength and crack growth resistance. The resulting fine grain microstructure, coupled with the alloy chemistry, enables a 25°C increase in temperature capability over the current disc alloy, 720Li, and an equivalent crack growth resistance to coarse grain Waspaloy. The

use of an advanced compressed air facility at Ladish has led to a 4% increase in the proof strength of RR1000, and through the development of a coarse microstructure in selected areas of disc forgings, there is an opportunity to increase the temperature capability of the alloy by 30°C.

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