ADVANCED HIGH TEMPERATURE MATERIALS: AEROENGINE FATIGUE

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ABSTRACT: The aeroengine business is intensely competitive and commercial success relies on continuous improvements in engine efficiency, with reduced environmental impact and lower operating costs. Innovation through advanced materials is a powerful tool in the quest for market share. This paper discusses the recent advances in materials technologies for aeroengine components with particular reference to enhancing the fatigue life of turbine disc components. Computational modelling of materials processing is contributing strongly to the improved design of the microstructure and the delivery of enhanced high temperature fatigue properties. Future trends are towards new manufacturing approaches in which the microstructure/composition is varied within the component in response to the service temperature and stress distribution. Dual microstructure and dual alloy concepts are described. Implementation of these approaches requires detailed knowledge of microstructure/property relationships and process models to simulate the structure generated during the complex thermomechanical processing.

Keywords: Nickel alloys, Fatigue, Crack growth, Thermomechanical processing

RESUMO: A indústria de turbinas de aviões é altamente competitiva e nela o sucesso depende de melhorias contínuas na eficiência dos motores, na redução dos impactos ambientais e no abaixamento dos custos operacionais. A inovação através de materiais avançados é um poderoso instrumento na procura de cota de mercado. Este artigo discute os avanços recentes em tecnologias de materiais para componentes de turbinas de avião, com particular referência à melhoria da vida à fadiga de discos de turbinas e seus componentes. A modelação computacional do processamento dos materiais contribui fortemente para a melhoria do projecto de microestrutura e para se conseguirem melhores propriedades de fadiga a alta temperatura. As tendências futuras apontam para novas abordagens de fabrico nas quais a microestrutura e composição são variadas no componente em resposta à temperatura de serviço e distribuição de tensões. Microestruturas duais e conceitos *dual alloy* são discutidos. A implementação destas abordagens requer conhecimento detalhado das relações microestrutura/propriedades e modelações do processo para simular a estrutura obtida durante processos termomecânicos complexos.

Palavras chave: ligas de níquel, fadiga, crescimento de fissura, processamento termomecânico

1. INTRODUCTION

One of the greatest engineering achievements of the second half of the 20^{th} century was the development of mass air transport. It is almost 50 years since a de Havilland Comet IV made the first commercial crossing of the Atlantic by a jet powered aircraft. The Comet of 1958 carried 76 passengers and had a range of 2870miles. The latest long range airliner to enter service is the Airbus A380 which can carry 555 passengers up to 9200miles. These advances have come about through continuous and intensive development of gas turbine propulsion systems. Table 1 compares some aircraft and engines over the past 5 decades. The range and payload have increased by approximately a factor of 4, whereas the engine thrust has increased by nearly an order of magnitude.

Perhaps even more remarkable has been the increase in fuel efficiency, Figure 1. Fuel consumption has dropped by 25% since the early 1970's and this improvement is predicted to continue. Major reductions in noise and emissions are minimising the environmental impact and exceptional reliability is achieved routinely to ensure punctual services and commercial profitability. It is not unusual for engines to stay on the aircraft for over 20,000 flying hours, with only minor routine maintenance being required for several years. This contrasts sharply with the Comet that required extensive maintenance after only a few hundred hours. All these advances have been underpinned by airworthiness requirements that ensure unprecedented levels of safety.

Ultimately the design of an aeroengine brings together a multitude of disciplines and technology but the entire system is dependent on materials with the performance to meet the design intent. Table 1 includes the turbine entry temperature (TET) of the various engines [1]. A rise of nearly 700° has been made possible by advancing materials technologies for the turbine. Figure 2 shows the temperature capability of nickel superalloys developed for turbine blade applications. The increased temperature capability, nearly 400°, has been achieved by a combination of alloy and process developments. Early alloys were formed by wrought processes, but all turbine blades are now cast, most via the single crystal process to maximise the creep resistance.

Table 1. Comparison of airliners and Rolls-Royce engines over the past 50 years

Aircraft	Date	Passengers	Range, miles	Engines	Thrust, lb	TET °C
De Havilland Comet IV	1958	76	2870	RR Avon RA 29	10500	900
Boeing 707	1960	181	5700	RR Conway	17500	1020
Lockheed Tristar	1972	256	4100	RR RB211-22B	42000	1250
Boeing 747-400	1989	416	7000	RR RB211-524	58,000	1350
Boeing 777-200	1995	301	8800	RR Trent 800	93,000	1550
Airbus A380	2007	555	9200	RR Trent 900	80,000	1580

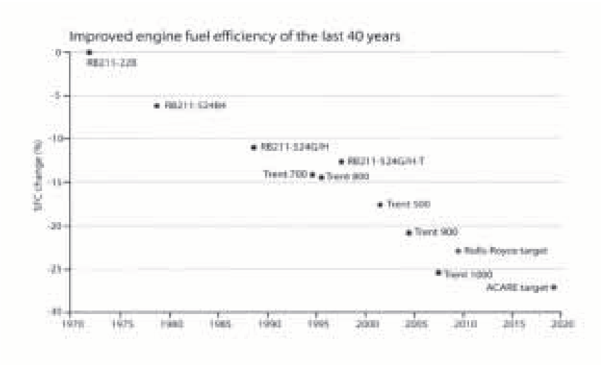

Fig. 1. Specific fuel consumption of Rolls-Royce engines, 1970-2020 [1].

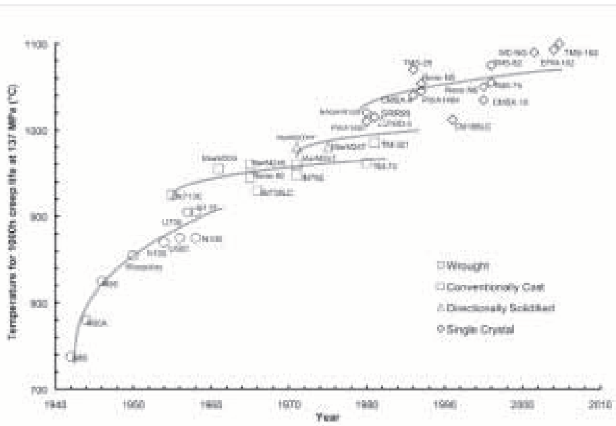

Fig. 2. Temperature capability of turbine blade alloy [2]

Plots of the form shown in Figure 2 are presented frequently to highlight the advances in turbine materials technology but what is equally important, and arguably much more difficult to introduce, is the supporting turbine disc technology to retain the blades at higher rotating speeds/temperatures and transmit the power to the high pressure turbine shaft. Failure of a turbine disc will result in the uncontained release of large pieces of metal with extremely high kinetic energy. This places the safety of the entire aircraft at risk. As a consequence the design of discs must demonstrate an extremely remote probability of failure within the service life. Methodologies for the lifing of discs are described in Reference [3]. The development objective for disc materials has been to achieve the longest possible service life commensurate with the validated lifing methods. In simple terms that means designing for the delay of fracture under ever increasing engine stresses and temperatures.

The remainder of this paper will focus on turbine disc alloys and the advances that have been made in the understanding of fatigue properties, their relationship to microstructure and the development of processes to maximise the disc life as operating temperatures rise.

2. TURBINE DISC ALLOY DEVELOPMENT

The turbine disc is subject to very high centrifugal stresses induced by the high rotational speed, plus additional loads on the rim imposed by the blades and

high thermal stresses induced in the bulk of the disc by the temperature gradients during engine start and shut down. The development of alloys and processes for turbine disc manufacture has concentrated on maximising the fatigue strength and minimising the scatter in properties. There is an additional requirement for disc designs to demonstrate adequate burst strength to protect the system in the event of an overspeed. A disc must be capable of withstanding an excursion to 122% of the maximum engine speed. Burst strength is essentially a direct function of the tensile strength, so a high tensile strength becomes a key design requirement. This is generally also directly related to fatigue strength.

Disc design has been predicated on the assumption that materials properties are not time dependent and hence creep and oxidation effects are not significant. It has therefore been the standard practice to design within the stress/temperature regime where these time dependent effects can be ignored. This assumption may not be valid in the future and will be discussed later in the paper.

For engines developed in the 1950's all the requirements could be met by forged steel components, typically based around a ferritic 12%Cr composition. Steel discs are relatively low cost, easy to manufacture, and provide good fatigue resistance. While turbine gas temperatures

Table 2. Chemical composition of some typical turbine disc alloys, wt%.

Alloy	Ni	Fe	Co	Cr	Mo	W	Ta	Al	Ti	Nb	Zr	Hf	C	B
FV535	1.0 max	Bal	5.0-7.0	9.8-11.2	0.5-1.0	---	---	---	---	---	---	---	0.05-0.12	---
Nim901	40.0-45.0	Bal	1.0 max	11.0-14.0	5.0-6.5	---	---	0.35 max	2.8-3.1	---	---	---	0.1 max	---
IN718	Bal	19	0.4	18	3	---	---	0.5	1.0	5	---		0.03	0.003
Waspaloy	Bal	---	13.5	19.6	4.3	---	---	1.37	2.95	---	0.06	---	0.019	0.0063
AP1	Bal	---	16.9	14.9	5.0	---	---	4.1	3.6	---	0.04	---	0.028	0.0002
Rene95	Bal	---	8.0	13.0	3.5	3.5	---	3.5	2.5	---	---	---	0.06	0.01
U720Li	Bal	0.08	14.57	15.92	2.98	1.35	---	2.44	5.18	---	0.042	---	0.023	0.016
RR1000	Bal	0.0-1.0	14.0-19.0	14.35-15.15	4.25-5.25	---	1.35-2.15	2.85-3.15	3.45-4.15	---	0.05-0.07	0.5-1.0	0.012-0.033	0.01-0.025
N18	Bal	---	15.4	11.1	6.44	---	---	4.28	4.28	---	0.019	0.50	0.022	0.008
ME3	Bal	---	18.2	13.1	3.8	1.9	2.7	3.5	3.5	1.4	0.05	---	.03	.03

Were relatively low, steel discs performed very well and were used extensively in gas turbine engine. Typical among the steel turbine disc alloys is FV535 (Table 2)

used in the Rolls-Royce Avon engines. Good strength is maintained up to about 450°C (Figure 3). However, strength and oxidation resistance fall rapidly at higher temperature. Hence by the middle of the 1960's steel turbine discs were being displaced by Ni-Fe systems such as IN718 and IN901. The high Ni content stabilised an austenitic crystal structure and precipitation hardening provided greater resistance to deformation at high temperatures and good microstructural stability. The 1970's saw further increases in the disc temperature to over 600°C where the long term stability of Ni-Fe alloys was poor. Ni alloys with increased precipitation hardening and higher thermal stability were introduced to extend disc capability above 600°C. Some typical alloys and compositions are shown in Table 2.

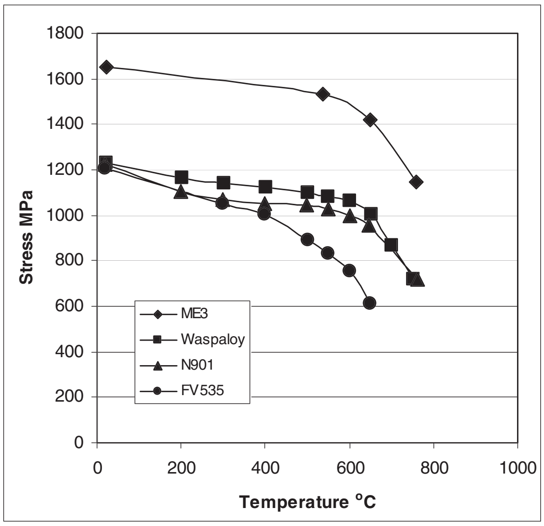

Fig. 3. Ultimate strength of typical disc alloys.

Following the series of Ni alloys from Waspaloy to ME3 it can be seen that the concentration of γ' forming elements (Al, Ti, Ta) and refractory elements (Mo, W, Ta) has been steadily increased. Both increase the strength of the alloys but the γ' precipitates are particularly beneficial for high temperature strength. Thus the required balance of properties was originally achieved with a well controlled grain structure to provide good fatigue resistance together with a simple uni-modal dispersion of fine γ' to ensure reasonable temperature strength capability (Figure 4a and b).

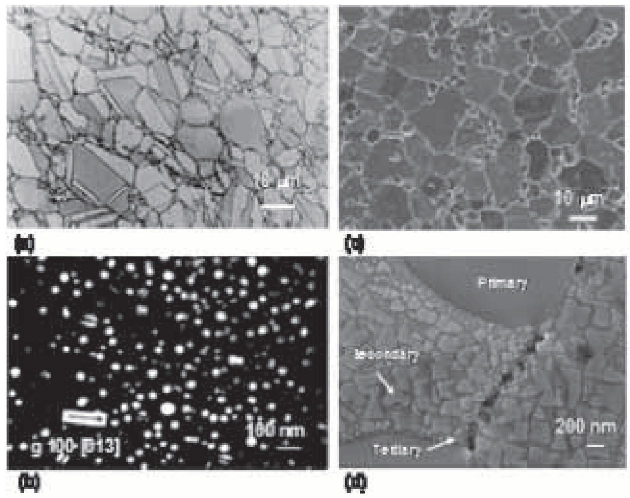

Fig. 4. Microstructures of disc superalloys (a) Waspaloy grain structure, (b) N901 γ' precipitates, (c) N18 grain structure and (d) N18 γ' precipitation.

However as engine temperatures increased, the alloys became more complex with bi-modal and tri-modal γ' distributions (Figure 4c and d) obtained through complex thermo-mechanical processing and heat treatment. In general the advanced alloys use a combination of hot deformation around the γ' solvus temperature combined with a sub-solvus solution treatment temperature which results in some coarse primary γ' particles (1-5µm) at the grain boundaries, which helps maintain a fine grain size (5-10µm). Subsequent ageing then gives a duplex γ' structure of secondary particles (100-500nm), which provides most of the temperature capability, together with fine tertiary precipitates (10-50nm) for additional strength.

The use of the relatively low temperature processing around the γ' solvus stems from the requirement for a fine grain size and as a consequence the latest alloys (e.g. ME3) contain carbon and boron to provide fine grain boundary precipitation to reduce grain growth.

At temperatures above the γ' solvus. The combined benefit of a fine grain structure and the higher volume fraction of γ' available for precipitation can be seen in the high low temperature strengths achieved (Figure 5).

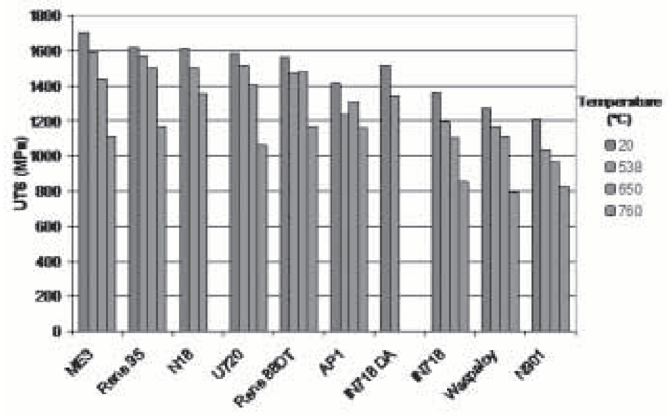

Fig. 5. Tensile strength of some Ni disc alloys

3. PROCESSING OF TURBINE DISC ALLOYS

The compositions of the disc alloys in Table 2 are very complex and the melting, casting and thermomechanical processing all require careful control to achieve the high level of properties demanded in discs, free from defects and segregation.

Vacuum induction melting (VIM) is commonly used to produce the first melt. Vacuum processes are essential due to the high reactivity of many of the alloying elements and to reduce the concentration of volatile trace elements. However the solidification of induction melted ingots is difficult to control which leads to excessive dendritic segregation. Consequently the VIM ingot is recast by the vacuum arc remelting (VAR) process. Control of the melt rate and the size of the melt pool allow considerable refinement of the cast microstructure and reduced segregation. In some highly critical applications an electroslag remelting (ESR) process may also be introduced between the VIM and VAR processes to reduce particulate oxides in the melt. The melt processing stages are absolutely critical since any inhomogeneities introduced at this stage cannot be removed by subsequent processes.

The cast ingot is subjected to a series of forging and heat treatment operations to produce the final disc forging and this is illustrated in the flow diagram below (Figure 6). Typical manufacturing operations after the production of the VAR ingot consist of an extensive homogenisation treatment at just below the alloy solidus, often preceded by a slow ramp to avoid localised melting associated with solidification segregation. This is then followed by extensive side forging (cogging) of the ingot at relatively high temperatures (~1100-1150^{o}C) to reduce the diameter and to break up the as cast structure. Further side forging is then carried out with a descending temperature schedule at temperatures in the vicinity of the γ' solvus combined with reheats to cause recrystallisation. This ultimately provides a billet with a refined uniform microstructure suitable for closed die forging. The disc is produced using tightly controlled temperature and strain rate conditions using two or three sets of dies, depending on the complexity of the shape. The development of the required microstructure during these operations and the final heat treatment is critical to the generation of the mechanical properties needed for subsequent safe service life of the component. Consequently the disc is then subjected to both non-destructive evaluation and mechanical tests on samples machined from part of the forging while more extensive tests are carried out on a cast/heat treatment batch basis. Subject to satisfactory inspection and test results the disc is then finish machined to the final shape required.

The general trend since the introduction of gas turbines for aero-engine applications has been towards increased size for improvements in thrust and efficiency which has led to a demand for increased ingot and forged billet diameters. However the control of grain size and particularly segregation become more difficult as ingot size increases. This difficulty has been compounded by the increased complexity of the alloys and as a consequence powder metallurgy techniques have been developed to facilitate the manufacture of the latest generation of superalloys. In this process VIM melted material is gas atomised into a powder which is subsequently consolidated into billet using either extrusion or hot isostatic pressing. The material is then forged into discs in the conventional way.

Clearly the whole manufacturing route is a very complex process and optimisation by empirical 'trial and error' methods is extremely expensive and very time consuming. Consequently a suite of process models has been developed to accelerate development and minimise the number of trial forgings required to optimise a given material and component shape.

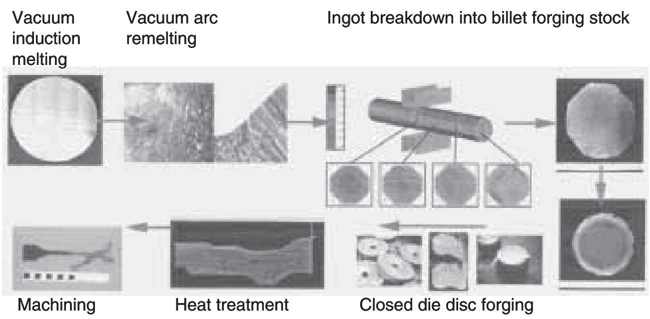

Fig. 6. Flow diagram of typical turbine disc Thermo-Mechanical Processing

4. OVERVIEW OF THE VARIOUS MODELS

The modelling tools developed can be divided into four main areas which cover the different functions of the manufacturing route;

(i) Solidification simulation for the prediction of ingot structure (e.g. dendrite arm spacing), segregation effects (e.g. freckle) and defects (e.g. inclusions and whitespot). The aim here is to provide better melt control, through current flow, melt rate and electrode feed rate, to ensure higher ingot consistency.

(ii) Billet breakdown modelling for the optimisation of pass and bite sizes and reheat schedules during side forging to provide economic production of billet with a uniform microstructure.

(iii) Closed die forging simulation to eliminate forging defects (e.g. laps and folds) and to define the forging conditions needed to produce the required microstructure and mechanical properties.

(iv) Modelling the evolution of residual stresses during heat treatment, to reduce cracking problems, and during machining, to control distortion in the final component.

The properties achieved in the disc are predominantly dictated by the casting and forging operations and consequently the modelling activities discussed here cover (i) to (iii).

4.1 INGOT MODELLING

The modelling of solidification during VAR and the prediction of grain structures and dendrite formation requires simulation over a wide range of length scales and consequently macro-modelling is carried out to determine heat transfer, fluid flow and electromagnetism [4-5] in the arc melting process while meso-modelling using both deterministic and cellular automata approaches are used for nucleation and grain growth [6].

The main goal of the homogenisation heat treatment of a VAR ingot is to dissolve intermetallic phases and to reduce segregation. However significant grain growth can occur during the long soak times at high temperatures that occur during this process. A cellular automaton (CA) model has been developed for simulating the grain growth that accompanies the homogenisation process of the VAR ingot. The CA model is based on discretely solving the classical rate equation of grain boundary movement at a mesoscopic scale. The effects of boundary curvature as well as mis-orientation of grains are considered for calculating boundary velocity.

Figure 7 shows the results of simulations of the grain structure produced after solidification after VAR and the development of grain structure at different stages of the homogenisation process. It can be seen that the models have predicted the actual behaviour with reasonable fidelity.

4.2 MICROSTRUCTURE MODELLING

A number of models of the recrystallisation behaviour of IN718 have been developed [7-10] which are based on empirical observations of the relationships between strain, strain rate and temperature and the recrystallised volume fraction and grain size. The model described below is that developed by Evans [7] and used by Dandre et al [8].

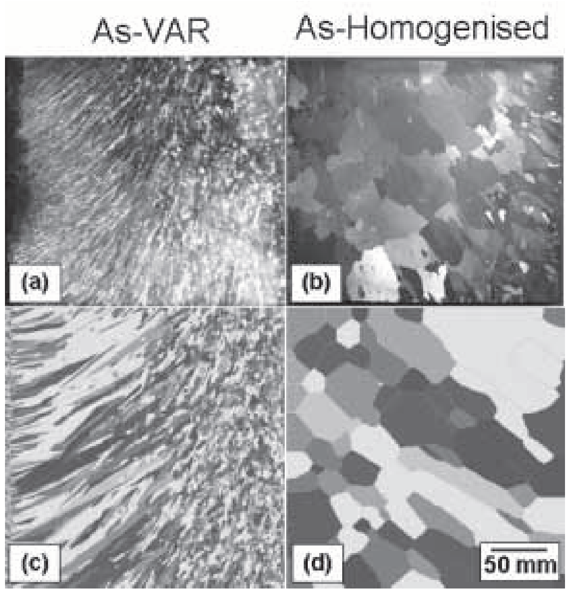

Fig. 7. Comparison of as VAR and homogenised IN718 ingot after [4] - (a-b) experiment (c-d) model

In this model recrystallisation commences when a critical strain is reached. This strain (ε_c) is defined by an Arrhenius based relationship of the form shown in Eq. 1.

$$\varepsilon_c = B.r^m.\exp(Q/R.T) \qquad (1)$$

Where B, m, Q are material constants, R is the gas constant, T is temperature and r is strain rate.

The expressions for recrystallised grain size and the recrystallisation rate ($dV/d\varepsilon$) both use identical forms to Eq.1 (with different values for the material constants) and the volume fraction recrystallised V_2 after time increment Δt is then given by Eq. 2.

$$V_2 = 1-(1-V_1)\exp((-dV/d\varepsilon).r.\Delta t \qquad (2)$$

where V_1 is the volume fraction at the beginning of the increment.

The grain size distribution (Figure 8) can then be readily determined from the summation of the incremental volume fractions recrystallised at a given grain size. The model is straightforward and can be easily coded into a subroutine in a finite element solver.

A more physically based model for recrystallisation in low stacking fault energy materials has recently been published [11-13] which uses a combination of dislocation density and grain nucleation and growth to define the microstuctural response to thermo-mechanical processing. The model also has the advantage that the reaction kinetics can be modified to take into account the effect of precipitation, which can occur during hot working below the gamma prime (Waspaloy) or delta (IN718) solvus temperatures. The inputs required are time, strain and temperature history data and an initial starting microstructure which can be either a uniform grain size or a distribution to facilitate multi-stage operations.

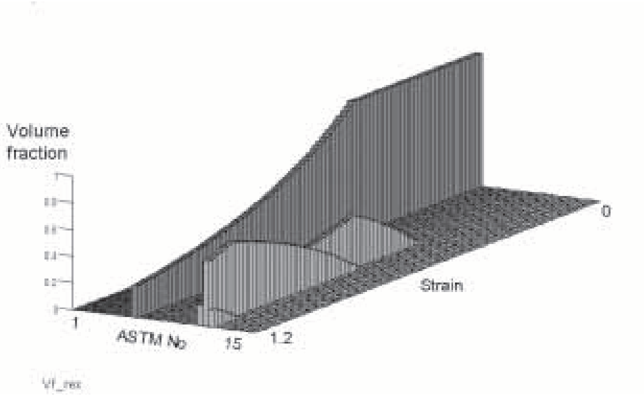

Fig. 8. Predicted grain size distribution in IN718 as a function of strain at 1000°C.

The model simplifies the microstructure by grouping similar individual grains into grain sets and assumes that new grains nucleate on existing grain boundaries. Within a grain set the model assumes all grains have the same dislocation density and diameter. As new grains are nucleated new grain sets are added to the model with information defining the grain boundaries on which they were nucleated

The grain sets are split into those which are recrystallising and those that are growing where the growth kinetics are both defined by simple Arrhenius functions. The recrystallising grain growth rate is also factored by the remaining volume available for growth and the dislocation density differences between grains. The known sizes and dislocation densities of the grains in the distribution allow the length of grain boundary that is available for nucleation to be calculated and hence the number of new grains to be nucleated (Figure 9).

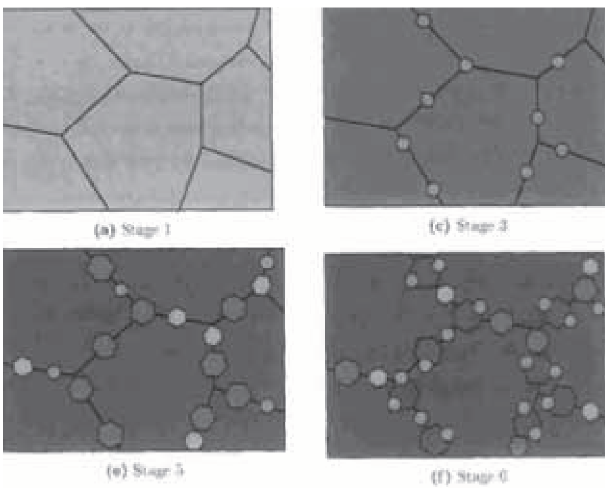

Fig. 9. Schematic diagram showing the evolution of new recrystallised grains on existing boundaries.

Precipitation can significantly reduce grain growth and the nucleation rate of recrystallised grains and consequently it is important to model both the precipitation and the dissolution particularly as forging is often carried out in the vicinity of the delta solvus. This is achieved through combining time-temperature transformation data with a Scheil calculation [14] to define the precipitation kinetics while dissolution is handled simply by assuming that a time of 5s above the solvus will cause the precipitates to go into solution.

Reasonable predictions of microstructural evolution can be made for both open die ingot breakdown operations and closed die forging (Figure 10) such that process routes can be simulated sufficiently well as to allow significant reductions in the physical trials needed for process development.

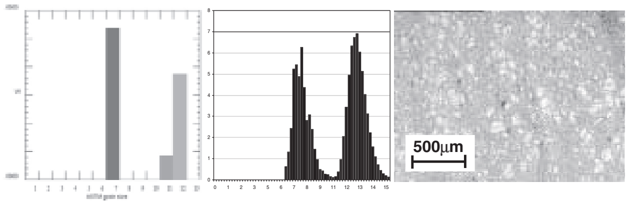

Fig. 10. Grain size predictions (ASTM No.) and micro-structure in an IN718 disc forged at 1000°C.

4.3 MODELLING OF MECHANICAL PROPERTIES.

The emphasis on modelling recrystallisation and grain size in superalloys is due to the need to provide uniform microstructures and so ensure consistent mechanical properties. However the relationship between microstructure and properties in superalloys is a complex function of grain size, gamma prime particle distribution and size and grain boundary precipitation. Thus the prediction of the properties resulting from a given process route is not simple as relatively small changes in alloy chemistry and process conditions can give rise to significant microstructure and property variation. However the composition control and uniformity of billet microstructure realised in the powder metallurgy disc alloys results in more reproducible behaviour, when compared to conventional ingot melted material, which facilitates the production of controlled microstructures. Forging and solution heat treatment of P/M superalloys at sub-solvus temperatures results in a structure where the grain size is determined by the initial powder particle size and a coarse dispersion (CD) of primary gamma prime particles at the grain boundaries [15]. The flow softening behaviour, which is commonly observed in forging these materials, is associated with the formation of increasing amounts of this coarse dispersion and fine recrystallised grains. This can result in the formation of a "necklace" type microstructure of CD regions around the initial grains which increases with deformation (Figure 11). The effect of this microstructural evolution on tensile properties is relatively small with the finer structural unit size providing slight increases in strength. However the effect on creep is more significant with a much greater dependence on CD (Figure 12a) where the finer structures result in lower lives. The effect on fatigue crack growth is more complex with indications that intermediate CD fractions of ~0.5 result in a minimum in crack growth rate (Figure 12b). The model described in [15] allows the prediction of CD to be carried out within finite element based forging models which provides the possibility of manipulating the deformation strain, temperature and strain rate to produce discs with controlled microstructures and optimised mechanical properties.

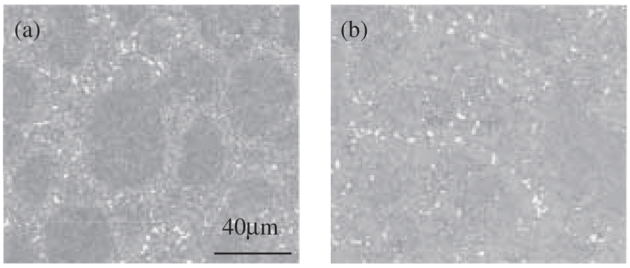

Fig. 11. Necklace microstructure in isothermally forged Nimonic alloy AP1, (a) 0.5 CD, (b) 0.74 CD.

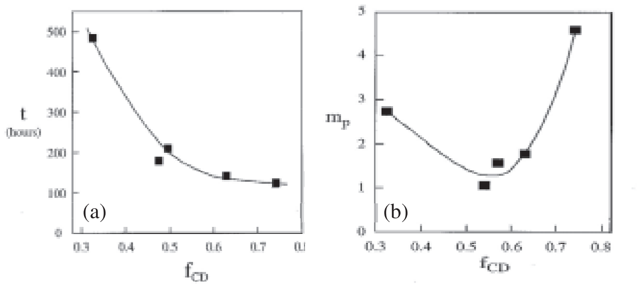

Fig. 12. Mechanical properties of Nimonic alloy AP1 as a function of CD fraction (a) Stress rupture life at 610 MPa and 730°C, (b) Paris exponent of crack growth rate at 600°C.

5. PROPERTIES OF HIGH STRENGTH DISC ALLOYS

The successful development of a high strength Ni disc alloy requires a balance of high strength with resistance to fatigue crack initiation and crack growth. Precise control of the microstructure is essential to achieve the optimum balance of properties. Figure 13 plots the first half cycle of 2 strain control fatigue tests of the alloy U720Li [16]. The standard U720Li alloy was heat treated below the γ' solvus and has a grain size of about 5μm. This ensured a high 0.2% yield stress of 1050MPa. A higher, super-solvus heat treatment coarsened the grain size to about 20μm and reduced the yield stress to 980MPa.

There was a small benefit of the coarser grain size on the strain control LCF life at 650-700°C but the fine grain material experienced a higher stress range for a given cyclic life, and hence it was more resistant to fatigue in terms of stress.

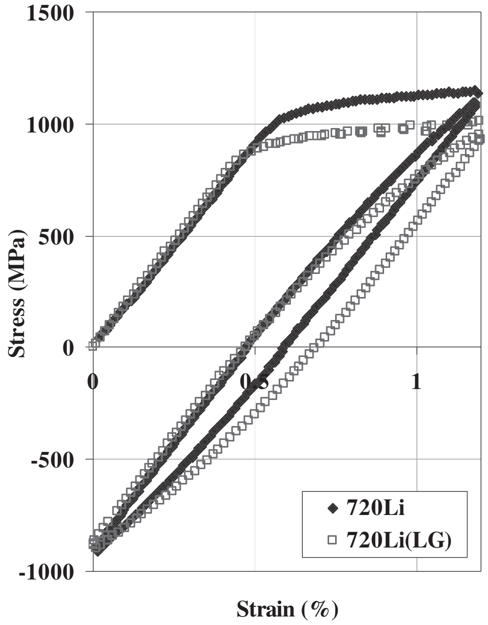

Fig. 13. Initial half cycle and stabilised hysteresis loop for stain control fatigue of U720Li at 650°C.

Fracture mechanics suggest that this small increase in life may be due to the effect that grain size has on the fatigue crack growth rate. Coarse grained U720Li showed a consistent reduction in crack growth rate, Figure 15. At 650°C the fine grain crack growth rate is twice that of the coarse grain. This difference is even greater at 725°C. Some data for Waspaloy with a grain size of 50μm confirms the benefit of a larger grain size at 650°C. The alloy RR1000 has been specifically developed to reduce crack growth rate by manipulating the γ' mismatch, SFE and APBE. The alloy can match the excellent crack growth rate of Waspaloy but with a finer grain size and higher strength [16].

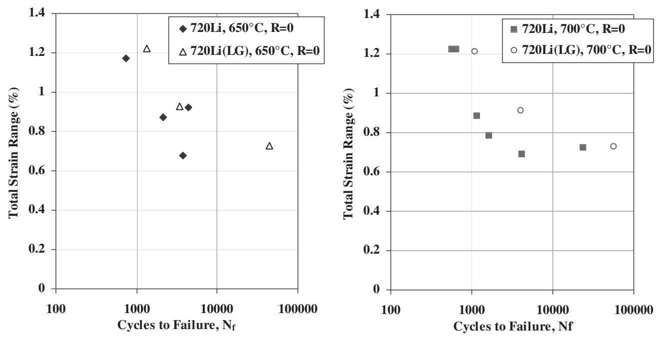

Fig. 14. Strain control LCF data for U720Li [16].

Data presented in Figure 15 was obtained with a trapezoidal wave form, 1second ramps with a 1s dwell at the maximum and minimum load. A further factor to consider is the issue of time at temperature and time under load. RR1000 was tested with a 20s hold on peak load [17]. A small increase in the crack growth rate was observed at 650°°C but at 725°C the 20s dwell increased the fatigue crack growth rate by nearly 10 times.

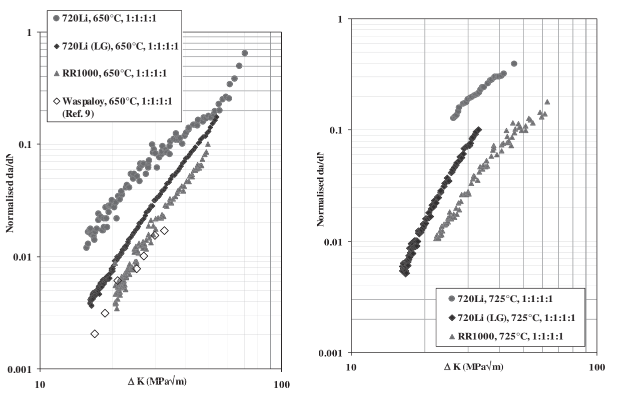

Fig. 15. Fatiguecrack growth plots at 650°C and 750°C [16]

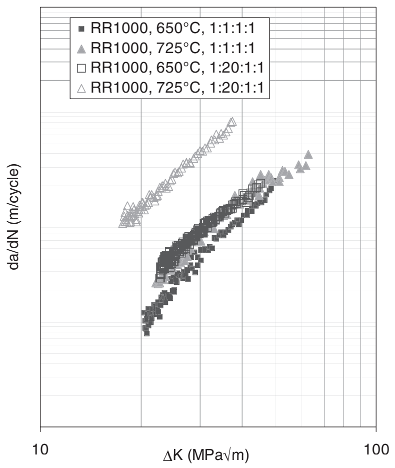

Fig. 16. Effect maximum load dwell on fatigue crack growth rate [17].

There has been much debate as to whether the increase in high temperature fatigue crack growth rate with dwell should be attributed to time dependent deformation (creep) or environmental interactions at the crack tip. Nickel disc alloys do creep at temperatures above 650°C under high loading. As an example, the creep properties of U720Li are shown in Figure 17.

The coarse grained U720Li had a clear creep advantage over the fine grain at higher temperatures and this suggests that the fine grained alloy should have a faster fatigue crack growth rate at elevated temperature due to creep. Indeed this effect can be seen in the data presented in Table 3 [18]. At 725°C the the crack growth rate of the fine grained alloy is nearly 3 times faster. There is also a large increase in the crack growth rate with a 20s hold at the maximum stress which is consistent with creep damage.

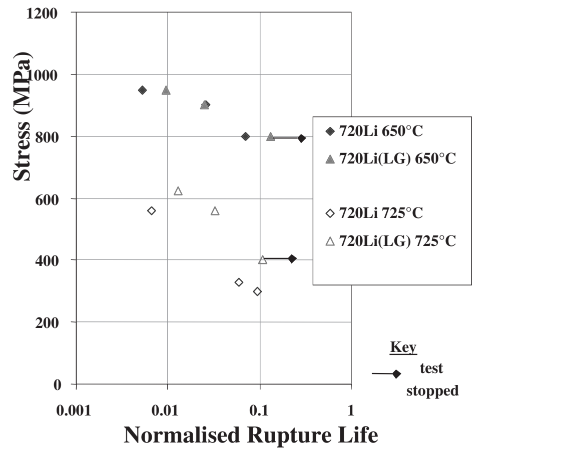

Fig 17. Creep rupture results for alloys tested at 650°C and 725°C [16].

However, tests under vacuum do not show a high dependence on the hold time, so environment is also an important factor. Indeed, in the case of U720Li the enviromental effect is dominant since in vacuum the 20s dwell only increased the crack growth rate by a factor of 2. The high crack growth rate for the fine grained alloy with a 20s dwell is largely due to oxidation at the grain boundaries promoting integranular failure. A coarse grained structure has less boundary area and so will be less sensitive to the environmental attack.

Table 3 – Crack growth rate of U720Li in um/cycle with a ΔK of 30 MPa/‾ m.

		650°C		725°C	
Dwell secs	**Grain size**	**Air**	**Vacuum**	**Air**	**Vacuum**
1	Coarse	0.7	0.2	1.5	0.3
20	Coarse		0.2	4.0	1.0
1	Fine	1.3	0.23	1.5	0.4
20	Fine	1.8	0.25	50.0	0.8

6. OPTIMISATION OF HIGH PRESSURE TURBINE DISC DESIGN.

The preceding section has provided examples of the improved high temperature properties that can be achieved in coarse grained Ni disc alloys. However, this is achieved at the expense of a reduced strength and load control fatigue life at lower temperatures. The engine component has many critical areas but these can be broadly sub-divided into 2 zones. The disc rim operates at the peak temperature, and the disc bore that experiences peak stresses at only a few hundred degrees. Traditional design has concentrated on optimising the material for the peak bore stresses and using a fine grained alloy with the maximum strength. However for new engines, where the rim temperature exceeds 700°C, this approach results in a heavy design due to the relatively poor creep and environmental resistance of the fine grained rim. Significant weight and cost savings would be possible if the disc had a larger grain size at the rim – a dual microstructure disc.

Many processes have been proposed for the production of dual microstructure discs [19]. All operate on the same basic principle, heat the rim above the γ' solvus to allow a controlled coarsening of the grain size while keeping the bore below the γ' solvus to retain the finer structure. However, the practical implementation this heat treatment process needs to be kept as simple as possible, while maintaining the close control on thermal conditions necessary for accurate microstructural control. Rolls-Royce plc and Ladish Co. Inc. have developed a process that uses an insulated heat sink to create a controlled temperature distribution across the disc forging during the solution heat treatment process. This is shown schematically in Figure 18. The central section of the forging is surrounded by an insulated block containing a heat sink, but the rim of the disc is exposed. When the pack is placed in a furnace above the γ' solvus the rim heats up quickly, the γ' dissolves and the grain size can increase. However the bore of the disc heats up much more slowly and the pack is removed from the furnace when the bore has reached the sub-solvus heat treatment temperature, thus preserving the fine grain size of the forging.

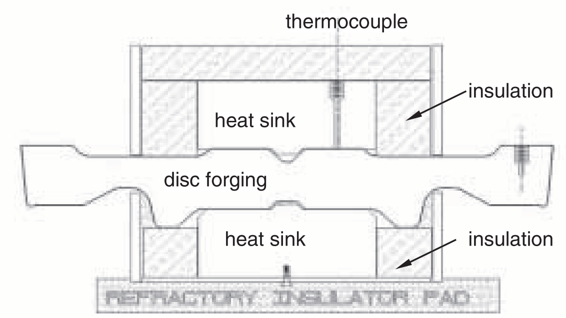

Fig. 18. Diagram of the heat treatment pack for development of a dual microstructure [19].

Although the concept of this process is quite simple, the practical implementation requires a very precise knowledge of the unsteady time/temperature cycle experienced by the disc material and the influence that this has on the microstructure throughout the disc (Figure 19). It is necessary to calculate the thermal profile thoughout the heat treatment and model both the dissolution of the γ' and the kinetics of the grain growth to ensure that the optimum microstructure is achieved throughout the part. The position of the coarse to fine grain transition in the disc is critical to achieving the lowest mass, most efficient design.

Beyond the dual microstructure concept there is is also the possibility of grading the chemical composition of the disc. Several proposed techniques for the manufacture of dual alloy discs are described in References 20-21. Figure 20 shows an example of a dual alloy disc made by HIP bonding a coarse grain U720LC rim to a fine grain P/M U720 bore.

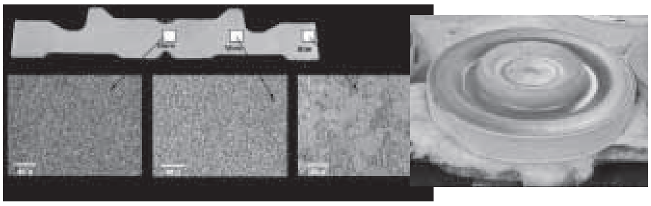

Fig. 19. Finished forging with dual microstructure (coarse grained rim, fine grained bore)

A next step may be to bond the single crystal blades to the disc to form a turbine blisc. This removes the parasytic mass of the fir tree root fixing, thus reducing the stress on the disc and facilitating a lighter, more efficient design. This is illustrated in Figure 20 which shows a demonstration blisc component produced with single crystal MC2 HIP diffusion bonded to a U720 bore. Blisc technology has found widespread applications in titanium compressor parts where the weight and cost savings are substantial

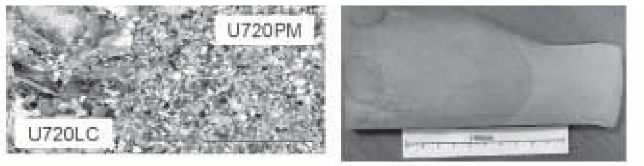

Fig. 20. Microstructure and bondline in a dual alloy disc with a coarse grain rim and fine grain bore [20].

Fig. 21. Demonstration blisk with MC2 single crystal dummy blades integrally bonded to a fine grain U720 hub.

7. CONCLUDING REMARKS

The metallugical technology applied to high temperature gas turbine discs has always been at the cutting edge. A unique combination of severe stress/temperature cycles, long life requirements, minimum weight and the safety critical nature of the components has driven alloy and process design to the limit. Recent developments have seen a move away from cast and wrought nickel alloys to higher strength powder alloys that combine excellent strength with creep, fatigue and environmental resistance. The complexity of the alloy systems and the high sensitivity to process conditions have necessitated the development of materials models that provide strong support for the design at all stages, from the initial alloy design through to the prediction of final machining stresses and estimation of in-service performance. The operating temperature of turbine discs in advanced engines exceeds 700° so that creep and environment interactions have a large influence on the fatigue life.

The modelling tools have enabled a far better understanding of the the interactions between the alloy, the process and the properties such that it is now possible to design components that have their microstructure modified locally to resist the dominant local damage mechanisms. This paper has demonstrated this through the example of a dual microstructure turbine disc where the properties of the component are varied to best resist high and low temperature loading.

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REFERENCES

[1] "The Jet Engine", Published by Rolls-Royce plc, London 2005

[2] R. C. Reed, "The Superalloys: Fundamentals and Applications". Cambridge University Press, 2006

[3] M. R. Winstone and G. F. Harrison, "Fatigue Lifing Methodologies for Aeroengine Discs and the Effect of Material Strength", 9th Portuguese Conference on Fracture, Setubal, 18-20 February 2004, 445, Portuguese Society of Materials.

[4] A. Kermanpur, D. Evans, R. Siddall, P. D. Lee and M. McLean, "Effect of Process Parameters on Grain Structure Formation during Vacuum Arc Remelting of INCONEL718", Int. Sym. on Liq. Metal Proc. & Casting 2003, Nancy, France, Sept 22-25, (Ed. Lee, P.D., Mitchel, A., Bellot, J-P, Jardy, A.), 281, SF2M, Paris

[5] W. Wang, P. D. Lee and M. McLean., Acta Mat. **51**, (2003) 2971.

[6] A. Kermanpur, W. Wang., P. D. Lee and M. McLean, Mat. Sci. and Tech. **19** (2003) 859.

[7] S. C. Medeiros, Y. V. R. K. Prasad, W. G. Frazier and R. Srinivasan,, Mats. Sci. and Eng., **A293** (2000) 198.

[8] B. Antolovich, and M. Evans, "Predicting Grain Size Evolution of UDIMET Alloy 718 during the Cogging Process", Superalloys 2000, 39, TMS Warrendale, PA

[9] R. W. Evans, "Microstructural Modeling of Near Net Shape Forgings", Proc. Conf. EUROMAT '89', Deutsche Gesellschaft fur Metallkunde.

[10] C. A. Dandre, S. M. Roberts, R. W. Evans and R. C. Reed, Mat. Sci & Tech, **16** (2000) 14.

[11] R. P. Guest, "The Dynamic and Meta-Dynamic Recrystallisation of the Nickel Base Superalloy Inconel 718", PhD Thesis University of Cambridge, (2005)

[12] R. P. Guest and S. Tin, "Dynamic and Metadynamic Recrystallisation of IN718", Superalloys 718 and Derivatives 2005), 373, TMS Warrendale PA

[13] Guest, R.P. and Tin, S., "Modelling Microstructural Transformations of Nickel Base Superalloy IN718 during Hot Deformation", Superalloys 718 and Derivatives 2005, 385, TMS Warrendale PA

[14] H. K. D. H. Bhadeshia, Metal Science, **16** (1982) 159.

[15] P. L. Blackwell, J. W. Brooks and P. S. Bate, J. Mat. Sci. & Tech., **14** (1998) 1181.

[16] I. M. Wilcock, D. G. Cole, J. W. Brooks and M. C. Hardy, "Creep/Fatigue Behaviour of P/M Disc Alloys". Charles Parsons Turbine Conference 2003, Dublin, Ed. Strang, A. et al, 749, Institute of Materials Minerals and Mining, London.

[17] I. M. Wilcock, M. C. Hardy and M. B. Henderson, "Elevated Temperature Fatigue Crack Growth Behaviour of a Nickel Disc Alloy", AeroMat 2002, Orlando, FL, ASM International.

[18] N. J. Hide, M. B. Henderson and P. A. S. Reed, "Effects of Grain and Precipitate Size Variation on the Creep-Fatigue Behaviour of Udimet 720Li in Both Air and Vacuum", Superalloy 2000, Ed. Pollock, T.M. et al, 495, TMS Warrendale PA,

[19] J. Lemsky., "Dual microstructure heat treatment for advanced turbine engine component", TMS/ASM Spring Symposium, Niskayuna, New York, 23 May 2006

[20] M. B. Henderson, H. Berlet, G. McColvin, J. C. Garcia, S. Peteves, J. Montagon, G. Raisson, S. Davies, I. M. Wilcock, and P. Janschek, Manufacture and spin test evaluation of bimetallic gas turbine components, Charles Parsons Turbine Conference 2003, Dublin, Ed. Strang, A. et al, 605

[21] H. Burlet, M. B. Henderson, G. McColvin, J. C. Garcia, S. Peteves, H. J. Klingelhoffer, M. Maldini, U. E. Klotz, T. Luthi, and I M. Wilcock, 2003, Microstructural and mechanical performance of diffusion bonded bimetallic model discs, Charles Parsons Turbine Conference 2003, Dublin, Ed. Strang, A. et al, 735-748