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AGARD/SMP Review Damage Tolerance for Engine Structures 2. Defects and Quantitative Materials Behaviour

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AGARD Report No.769

AGARD/SMP REVIEW

DAMAGE TOLERANCE FOR ENGINE STRUCTURES

2. DEFECTS AND QUANTITATIVE MATERIALS BEHAVIOUR

Papers presented at the 67th Meeting of the Structures and Materials Panel of AGARD in Mierlo, Netherlands, 2—7 October 1988

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14. Abstract

Most current military and all civil engines are operated under "safe life" procedures for their critical components. Experience has shown that this philosophy presents two drawbacks:

a. The move towards designs allowing higher operational stresses, and the use of advanced high-strength alloys make it likely that a disc burst could happen (following a rapid crack growth) well before the statistically-based "safe life" has been achieved.

b. It is potentially wasteful of expensive components, since it has been estimated that over 80% of engine discs have ten or more low cycle fatigue lives remaining when discarded under "safe life" rules.

Damage tolerance being an alternative lifeing philosophy, the Sub-Committee on "Damage Tolerance Concepts for the Design of Engine Constituents" has therefore decided to conduct a series of four Workshops addressing the areas critical to Damage Tolerance design of engine parts.

The present report includes the papers presented during Workshop II dealing with Defects and Quantitative Materials Behaviour. It also includes the content of the discussions which followed the presentations.

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PREFACE

Most current military and all civil engines are operated under "Safe-life" procedures for their critical components. Experience has shown that this philosophy presents two drawbacks:

(a) The move towards designs allowing higher operational stresses, and the use of advanced high-strength alloys make it likely that a disc burst could happen (following a rapid crack growth) well before the statistically-based "Safe-life" has been achieved.

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It also includes the content of the discussions which followed the presentations. On behalf of the Structures and Materials Panel, I would like to thank the authors, the recorders of the discussions and the session chairmen whose participation has contributed so greatly to the success of the Workshop.

R.LABOURDETTE
Chairman, Sub-Committee on
Damage Tolerance Concepts
for the Design of Engine
Constituents

* * *

La totalité des moteurs civils et la plupart des moteurs militaires sont actuellement mis en oeuvre suivant les concepts de "durée de vie certaine" en ce qui concerne leurs parties vitales. La pratique de cette approche a mis en évidence les deux inconvénients suivants:

(a) La tendance à l'utilisation des moteurs sous contraintes mécaniques plus élevées et l'emploi d'alliages à haute résistance rendent possible l'éclatement d'un disque (à la suite d'une progression rapide de fissure) avant que la "durée de vie certaine", évaluée statistiquement, ait été atteinte.

(b) On observe également un gaspillage de pièces onéreuses, puisqu'on estime que 80% environ des disques retirés du service conformément aux règles de "durée de vie certaine" ont encore un potentiel supérieur à dix durées de vie en fatigue oligocyclique.

La Tolérance aux Dommages constituant une autre approche possible de la définition des potentiels de vie, le Sous-Comité "Concepts de Tolérance aux Dommages pour le dimensionnement des composants de moteurs" a décidé d'organiser une série de quatre Ateliers consacrés aux divers aspects de la Tolérance aux Dommages appliquée aux moteurs.

Le présent rapport contient les diverses présentations effectuées à l'occasion du deuxième d'entr'eux traitant des Défauts et aspects quantitatifs du comportement des matériaux. On y trouve également un compte-rendu des discussions qui ont suivi les diverses présentations.

Au nom de la Commission Structures et Matériaux, je remercie les auteurs, les rapporteurs de discussion et les présidents de sessions qui on grandement contribué au succès de cet Atelier.

R.LABOURDETTE
Président du Sous-Comité
"Tolérance au Dommages pour les
composants de moteurs"

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From USA and Canada
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Attn: SMP Executive
APO New York 09777

Tel:(1)4738 5790 & 5792
Telex:610176F

CONTENTS

DEFECTS, THEIR BEHAVIOUR AND RELATED DESIGN ASSUMPTIONS

R.H.JEAL
Chief of Materials and Mechanical Technology

Rolls-Royce plc
P O Box 31
Derby
DE2 8BJ
U.K.

1. Introduction

The design engineers traditional view of materials has been enshrined in the assumptions he makes - namely "All materials are homogeneous elastic isotropic media free from defects" Fig 1 (Ref.1). The metallurgical and manufacturing engineers job has been to produce materials that met those assumptions as nearly as possible and design criteria have been set to keep operating conditions within the limits of behaviour described by those assumptions.

Where the assumptions broke down, usually because of the presence of manufacturing aberrations or lack of understanding of operating conditions the method of manufacture or component design was changed to make the assumption valid again. If the component could not be designed within the criteria of the current material then a new material was developed with a higher yield stress or better creep properties so the operating conditions again fell inside the assumed envelopes.

As the designers of engines sought for greater efficiency and lower weight the stresses and temperatures imposed on components gradually rose. Under these more arduous conditions the basic design assumptions starting breakdown on three counts.

- "Defect free" material cannot be produced in spite of cleaner, better control manufacturing routes.
- Materials are not continuous isotropic media but do have a microstructure and directionally that changes their behaviour.
- Materials are not elastic, they exhibit plasticity, fatigue and creep effects which are critical in determining their behaviour.

This workshop, one of a series of four covering the structures and materials technology required to support the application of damage tolerance concepts to engine component examines how far materials deviate from the traditional assumptions of their nature and behaviour and discusses the changes required in design, analysis and manufacture practices to maintain safe efficient engines with the required integrity.

2. Current Material Behaviour Assumptions

Detailed observations over may years has shown that, under cyclic loading, all materials can go through four phases to failure.

a) crack nucleation.

b) stable growth of short cracks.

c) stable growth of long cracks.

d) final unstable failure.

These are shown diagramatically in Fig.2 (Ref 1).

The relative importance of these phases depends upon the nature of the material deformation mechanisms, the material structure, the nature of any discontinuities present and the applied loads, temperature and environment.

2.1 Crack Nucleation

This phase is dependant upon:

a) the presence of discontinuities in the material.

b) the nature of strain distribution within the material.

Discontinuities, if they are not already crack-like, hence bypassing this phase altogether, act as local 'stress' raises within the material concentrating the applied damage into local areas within the component. They are always present because of the basic material structure (grain boundaries, second phases, precipitates, twins etc) required to give the described mechanical properties and the deviations from the structural requirements created by the manufacturing process (eg. porosity, inclusions).

The high proof strength materials developed for engine components gain their properties by using various solid solution or precipitative hardening systems to limit the slip taking place. This tends to lead to strain localisation (Fig.3) which limits the volume of material absorbing damage.

When these two effects are combined (as in early powder materials) the matrix surrounding the discontinuity rapidly cracks (Fig.4).

A discontinuity is only a defect if its presence was unsuspected and its behaviour not accounted for.

2.2 Stable growth of short cracks

Once a 'crack like discontinuity' is present what happens next is dependant upon the material slip mechanisms, its structure, the size of the crack and the nature of the applied loads.

Whether a crack can be considered as short or not is shown in Fig 5.

In planar slip materials with cracks of a size similar to the microstructural features the short crack growth phase is one of the most crucial ones to component behaviour accounting for most of what has been called 'initiation' before an "engineering crack" appeared, but the growth mechanism has been so dependant upon microstructure (Fig.6) that the models used for life prediction were based upon totally false premises.

Once such a crack grows to a size which is large compared to the material structure (averaging out the crack tip effects) or the mechanics changes to away from pilar slip (eg. with a change in temperature) the general rate and direction of growth becomes dominated by the applied loads.

2.3 Stable growth of long cracks

This phase has been well studied since the 1960's when crack growth prediction techniques based upon linear elastic fracture mechanics were first developed. Crack growth becomes practically independant of material structure or even detailed alloy chemistry showing only efforts of alloy type and general structure (eg. ferritic steel, austenitic steel, nickel superalloy etc). The rate is very dependant upon applied loads and crack shape and size - the traditional model of materials therefore becomes more applicable again.

2.4 Final Unstable failure

Final failure takes place when the section can no longer sustain the load or crack growth becomes unstable under the condition described by the appropriate fracture mechanics approach - again little dependant upon material type.

3. Application of assumption to design and manufacture

An understanding of the modes of behaviour described above and the criticality of each one under various operating conditions for various materials allows for better definition of the criteria that can be used to optimise design and the materials required to meet those design criteria. These in turn lead to a better understanding.

a) the data and methods required to support design.

b) required specification and quality standards.

c) the process controls required in manufacture.

d) the inspection and release procedures required for components.

The aim must always be the match design intent and process capability to meet fully defined engineering requirements as illustrated in Fig 7.

3.1 Design database

Before correct data can be chosen the mode of behaviour expected for the conditions known to exist has to be identified and an appropriate analytical model chosen taking account of the material to be used.

Data appropriate to that model must be available obtained on relevant material to the appropriate quality standard.

It is highly unlikely that behaviour criteria can be set on laboratory data alone - component testing is normally essential to cover all influences.

3.2 Specification and Standards

The material to provide the defined criteria needs to be fully defined, both structure and standards taking account of the manufacturing processes concerned. Full material characterisation is essential.

3.3 Process controls

The on going standard of the material can only be ensured by identifying the critical process parameters and matching the product to th required specification. If it can't by controlled it shouldn't be relied on as a criteria.

3.4. Inspection and release procedures

These ensure that the process has been properly controlled and the product meets requirement. They should not be used as a "gate" to allow selected components through and reject the rest.

4. The workshop

The papers following this introduction cover the critical parameter introduced above.

a) Materials structure and manufacturing processes.

b) Component defects and their effect on behaviour (Discs and Blades).

c) Effect of structure and discontinuities on component behaviour.

Once we have established the behaviour of the real world around us we can examine the tools we require to optimise design, develop materials and ensure integrity.

a) Modelling methods.

b) Development of future technology.

Finally we examine the need for common approaches to prediction and data.

These papers and the associated discussion should give us a solid foundation based upon the real behaviour of materials to build our future techniques and criteria for safe optimised components.

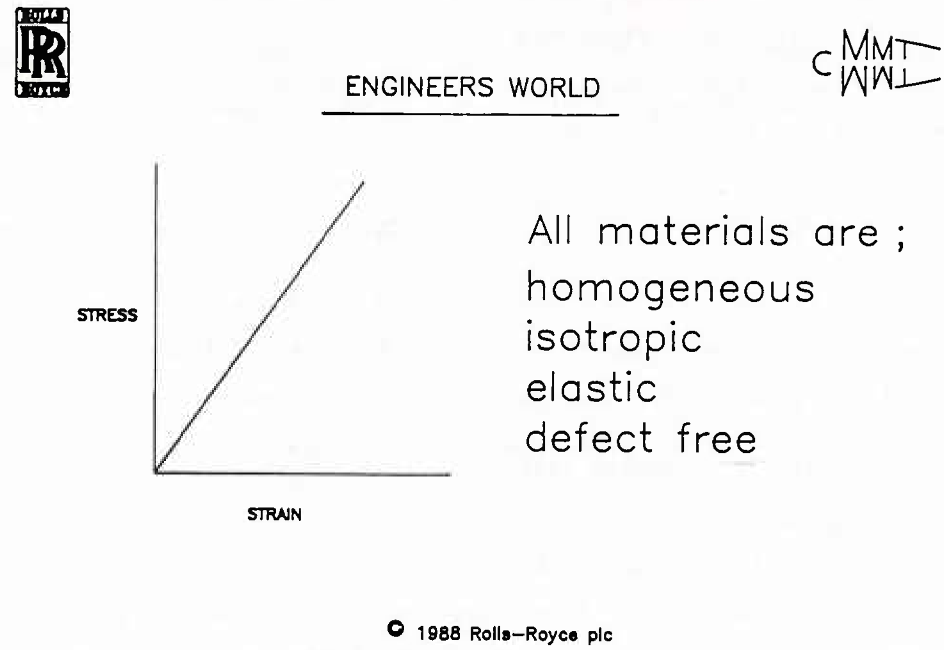

Fig 1 Engineering assumptions

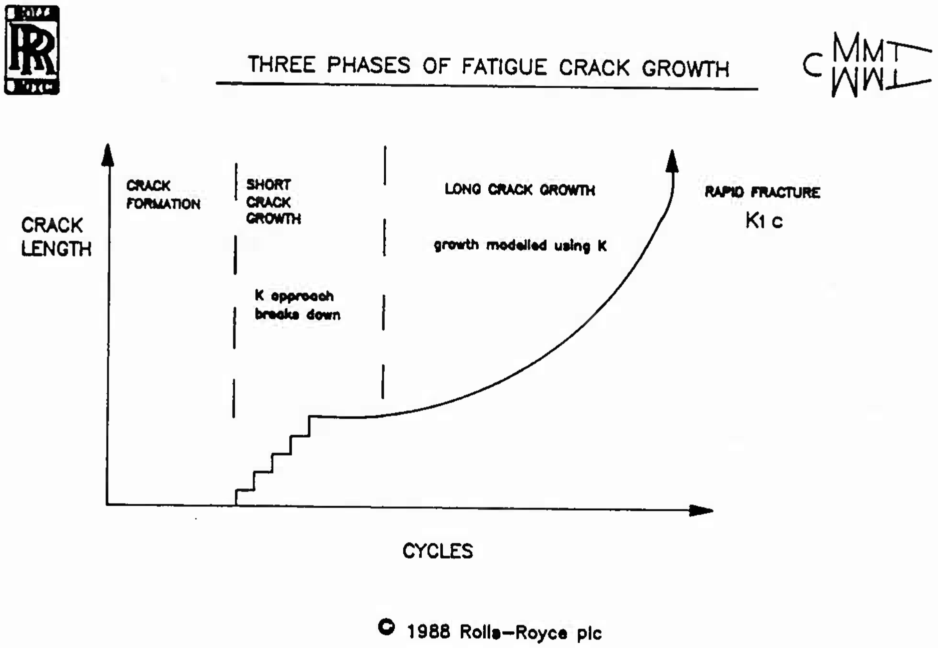

Fig 2 Phases of crack growth

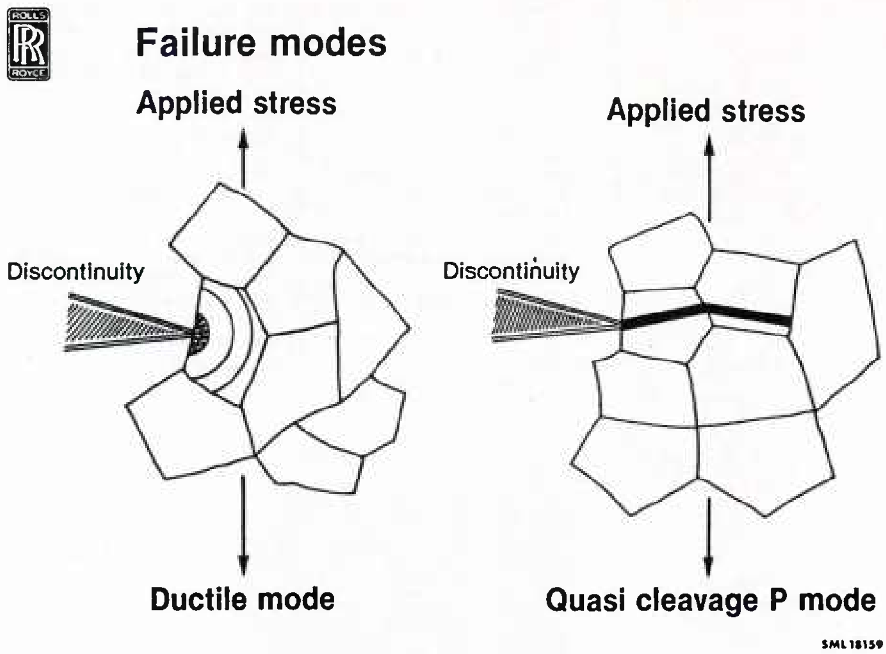

Fig 3 Crack tip failure modes

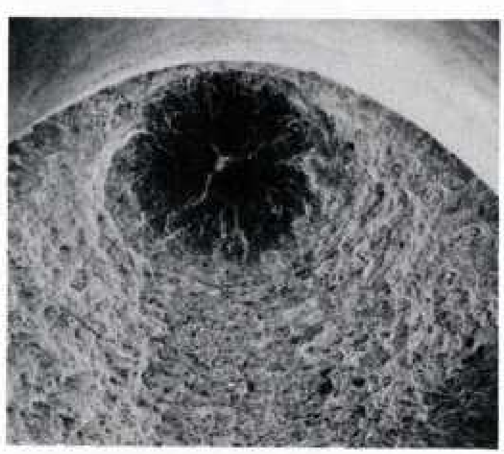

Fig 4 Subsurface failure in powder Ni–base alloy

MATERIAL MICROSTRUCTURE \ STRUCTURAL SECTION	LONG CRACK COMPARED TO SECTION SIZE DESIGNATION OL	SHORT CRACK COMPARED TO SECTION SIZE DESIGNATION OS
LONG CRACK COMPARED TO MICROSTRUCTURAL UNIT SIZE DESIGNATION LO	- CONVENTIONAL CONTINUUM MECHANICS APPROACH - LEFM - THROUGH - CRACK DATA IS APPROPRIATE E.G. COMPACT TENSION - LARGELY INDEPENDENT OF MICRO-STRUCTURE DESIGNATION LL	- CONVENTIONAL CONTINUUM MECHANICS APPROACH APPROPRIATE - SIMILITUDE WITH LL PROVIDED CLOSURE EFFECTS ARE TAKEN INTO ACCOUNT - TENDS TO GIVE SLOWER CRACK PROPN RATES IF EXPRESSED AGAINST NOMINAL Δ K. (IN LABORATORY AIR ENVIRONMENT) DESIGNATION LS
SHORT CRACK COMPARED TO MICROSTRUCTURAL UNIT SIZE DESIGNATION SO	- MODIFIED LEFM APPROACH REQUIRED WHICH INCORPORATES CRYSTAL ANISOTROPY - CRACK DIRECTION AND STEP LENGTH DEPENDENT ON SECTION GEOMETRY AND CRYSTALLOGRAPHY NO SIMILITUDE WITH OTHER MODES DESIGNATION SL (SPECIAL CASE - SINGLE CRYSTAL)	- INDIVIDUAL STRUCTURAL UNITS CAN BE TREATED AS "ANISOTROPIC CONTINUUM" - LEFM INVALIDATED BECAUSE THE CRACK SIZE IS COMPARABLE TO THE EXTENT OF CRACK TIP PLASTICITY - CRACK DIRECTION, STEP LENGTH DOMINATED BY CRYSTALLOGRAPHIC CONSIDERATIONS DESIGNATION SS

Fig 5 Definitian af crack types in planer slip materials

Fig 6 Microstructural influence on crack path in a Ti alloy

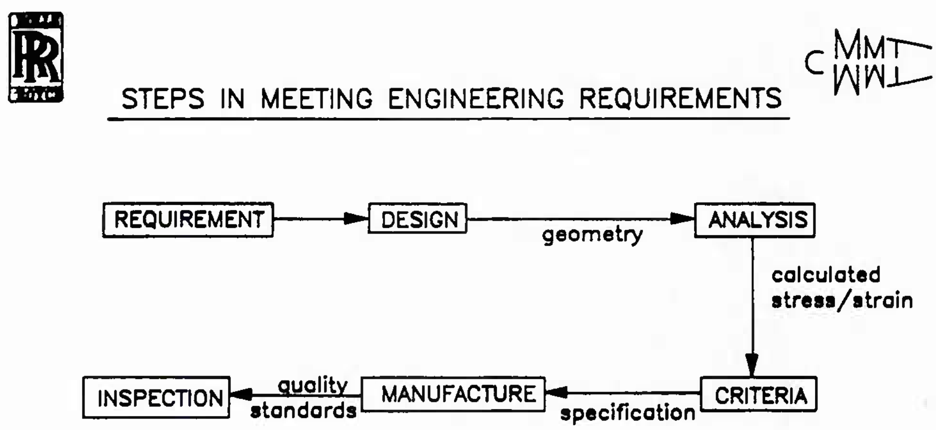

Fig 7 Steps in meeting engineering requirement

THE DAMAGE TOLERANCE OF DISC MATERIALS: METALLURGICAL FACTORS; RELATIONSHIP WITH PROCESSING PARAMETERS

by

E.J.Bachelet
SNECMA, Laboratoire Matériaux et Procédés
Service YKOE
B.P.81, Route Nationale 7
91003 Evry Cedex
France

1. INTRODUCTION

Until recently, the developments of discs material were conducted within the framework of design and lifing methods which essentially took into account the resistance to burst, fatigue crack initiation and creep, of these parts. Based on these criteria, alloy design and processing options, aiming at increases stresses or temperatures in operation, were often accompagnied, for metallurgical reasons which shall be reviewed, by a decrease in the intrinsic damage tolerance of discs materials. In the early eighties such an evolution, was the origin of difficulties which prompted basic research to understand and improve the damage tolerance ofthese materials, and more recently to develop more adapted materials . In parallel, evolutions in regulations put explicit demands to insure an adequate damage tolerance of the newer engines : ENSIP for instance.

The following paper will review the metallurgical factors which effect the damage tolerance of discs materials belonging to the classes of alloys presently used, Titanium-base and Nickel-base. The main challenges facing the materials designers shall be presented.

2. TITANIUM ALLOYS

2.1 General features

The low density and excellent specific properties of Titanium alloys warrant their extensive application for the discs of modern compressors (Fig. 1). Aside of the still widely used Ti 6-4 - lower temperature range, moderate strength - two broad lines of developments have been conducted :

. alloys with high specific strength for medium temperatures applications (400 to 450°C) : Ti 6.2.4.6. , Ti17 , IMI 550 , ...

. alloys for high temperature applications : Ti 6.2.4.2.S. , IMI 685 , IMI 829 , IMI 834 , Ti 1100, which extend the use of Titanium alloys up to about 600°C.

The cyclic behaviour of Titanium alloys has often been commented in contradictory terms : their basic fatigue properties may be excellent, with F/R ratios reaching .6 and their crack propagation behaviour (long cracks) measured on CT is caracterized with high thresholds and acceptable propagation rates; on the other hand they are credited with both a high scatter in fatigue properties (ref. 1), a high sensitivity to surface conditions, and their in-service use has sometimes proven to be deceptive, revealing an unexpectedly poor damage tolerance, for reasons not immediatly understood. This complex picture has been progressively elucidated by systematic studies establishing relationships between, processing conditions, microstructures and mechanical properties. A key point, in this picture appears to be the influence of microstructures upon crack behaviour.

This complex situation will be best discussed on the basis of a classification of Titanium alloys established upon the relative stability of the low temperature (hcp) and high temperature (bcc) phase : this classification reflects the phase equilibria, the alloy stability and therefore the usable range of temperature, the phase transformation conditions allowing or not effective microstructural control by thermo-mechanical processing, second phase hardening and on the whole the mechanical properties of the resultant products (fig. 2).

As regards processing conditions, two extreme options lead to two extreme types of microstructures with strong implications upon mechanical properties :

- (α + β) final processing and heat treating leads to generally fine microstructural features associated with higher ductility, higher L.C.F. strength, lower fracture toughness.

- (β) final processing or annealing leads to coarser microstructural features associated with higher fracture toughness and lower fatigue crack propagation rates (long cracks)

2.2 The damage tolerance issues

2.2.1. The high temperature near-α alloys : from β annealing to (α + β) TMP

A high content in α-stabilizers promotes the formation and stability of the α phase more resistant to creep deformation than the β phase. All high temperatures titanium alloys are from the near-α type, meaning that at operating conditions, there are minute amounts of β phase (a few %) at equilibrum. The β transus are correspondly high. The most favourable microstructures as regards creep resistance are the relatively coarse structures inherited from forging or solutionning from the all (β) field (ref. 2). The high temperature IMI 685 still widely in use in Europe for high temperature applications is based on this alloy / processing / microstructure association : after forging sequences, the alloy is solution treated above the β transus (1050°C O.Q.) and stress relieved. This normally leads to "moderate" grain sizes ($\simeq$ 1000 microns) and upon industrially realistic cooling conditions, plate-like α phase precipitates from the grain boundaries or within the grains in colonies (fig. 3). Cristallographic relationships do exist between a prior β grain and the α plates, and consequently different colonies of apparently different orientations belonging to a same prior β grain are also cristallographically related (ref. 3).

A high scatter appears in the fatigue lives of IMI 685 : fatigue life seem to correlate best with the size of the prior β grains which is seen to vary to a large extent - from 1 to 5 mm - under small variations in TMP (even at equivalent circumferential locations within a given part : fig.(4)). Faceted fracture is observed in IMI 685 both in monotonic and cyclic loading (ref. 4). DAVIDSON and EYLON have demonstrated that a single facet could cross several α plate colonies of different apparent orientation, along or near a common crystal plane (0001) within a prior β grain (ref. 5). Faceted fracture has been credited for the high fracture toughness (typically of the order of 75 MPa√m) and good propagaton behaviour of long cracks (ref. 6) high thresholds with acceptable crack growth rates. For a long time the discrepancy between L.C.F variability and overall excellent crack propagaton behaviour remained unexplained. A full understanding of the damage tolerance in these alloys was brought by the studies of short crack behaviour of BROWN and HICKS (ref. 7) : they measured high cyclic crack growth rates for a propagation along crystallographic planes through several α plate colonies belonging to a same prior grain, up to length of 3.5 mm. Very clearly, prior β grain size appeared to be the key microstructural feature controlling both L.C.F. and damage tolerance and the processing options could be reassed on this basis : when annealed above their transus, all titanium alloys experience rapid grain growth susceptible to lead to grain sizes all the more large that the β transus of the alloys are high. In the near-α alloys β-annealed containing large grains, the inhomogeneity of the deformation under cyclic loading rapidly induces L.C.F damage accumulation and crack formation. These cracks, rapidly grow transgranularly up to large sizes within the ex-β grains : both durability and damage tolerance of these alloys are seriously impaired. On the other hand, the relatively strong crack closure effects induced by the large facets are at the origin of high thresholds and low propagation rates of long cracks at least at low R-ratios (fig. 5).

In the further developments of higher temperature alloys, IMI 829 or IMI 834 (ref. 8), a solution treatment in the (α + β) field was adopted allowing effective β grain size control by pinning of the β grain boundaries by primary α particles. In order to maintien good creep and fracture toughness properties, the primary α content had to be as low as possible to allow, upon cooling, a high fraction of β to transform into plate morphology products favourable to creep and fracture toughness. The difficulties with these concepts lie in the rapid variation of the α phase content with temperature just below the β transus : it raises again the questions of process control and scatter. Isothermal forging is advisable. Variations in chemistry from heat to heat and even within heats should be kept to a minimum. In IMI 834, the composition of which was optimized for a more progressive α phase solutionning than in IMI 829, the new "compromise" was achieved with a significant loss in fracture toughness.

All the consequences resulting from β annealing, described above apply to (α + β) alloys : see for instance the cas of Ti 6-4 in ref. (9). This explains probably why, in spite of interesting results, β annealing of Ti 6-4 or Ti 6.2.4.2.S was never applied to engine discs.

2.2.2. High strength medium temperature (α+β) alloys : (α+β) T.M.P or (β) forging

Ti 6-4 as it is used today for discs applications may be considered as a prototype for (α + β) processing. When normally performed, the final conversion of billets and the forging operations followed by a partial solution + age heat treatment, break up the prior coarse grain and plates structure by several sequences of deformation / recrystallisation within the two phase (α + β) field, and lead to a fine grain structure. The primary phase is normally of rounded shape and its volume fraction and size is determined by the conditions of the partial solutionning treatment (generally VF ≃ 50%, d ≃ 10 to 20 microns : see fig. 6) : needles or more often plates of secondary α precipitate within the β phase as transformed products. The coarseness of that precipitation depends mainly upon the rate of cooling after partial solutionning.

The fine microstructural features associated with these conditions induce high ductilities, high fatigue strength (initiation) but on the other hand poor fracture toughness and relatively poor crack growth resistance : crack deviation mechanism and roughness induced closure are not operative. They also lead to lower creep strength than β-annealing. Damage tolerance problems of several origins have been recorded :

- when moving to higher strengths alloys for weight saving purposes the inverse strength / fracture toughness relationship seems unavoidable and rapidly leads to small critical defects difficult to detect (fig. 7) : clearly one is confronted with contradictory requirements between durability (fatigue initiation resistance) and damage tolerance. The issue is further complicated by the relatively poor hardenability of (α + β) alloys.

- in some alloys the damage tolerance has been reported to be strongly dependant upon heat treatment which may create an interface between the α and β phases with a poor resistance to short crack propagation (ref. 10).

The high strength alloys following these options, Ti 6.2.4.6. for instance, are relatively stable (α + β) basis. They have a good potential for medium temperature applications : however the compromise between mechanical properties are delicate to insure. Recent studies in these materials aim at improving creep strength, fracture toughness, damage tolerance by forging from the(β) field along the lines of the next paragraph.

2.2.3. Metastable (β) alloys

(α + β) alloys with a sufficiently high content in β-stabilizers remain in a metastable condition on rapidly cooling to room temperature from the all (β) field ; the β-stabilizer content may be, however, sufficiently limited to allow subsequent α precipitation by appropriate aging treatments, inducing strong hardening effects. Ti 17 is an exemple of the application of these concepts to engine discs. It is forged from within the all (β) field, precipitates some primary α upon cooling after forging and then receives a solution + age treatment intented to set the primary α volume fraction and the secondary α distribution (spacing, size). The grain structure is generally medium size (≃ 500 microns) warm-worked.

These options in processing and microstructural control allow to develop high strength alloys for intermediate temperatures applications with more favourable overall compromises on mechanical properties than high strength (α + β) alloys either (α + β) or β processed. They are in particular in a better position from a damage tolerance point of view as reflected for instance by they yield strength fracture toughness relationship (fig. 7)

- yield strength is controlled (ref. 11) by the secondary α phase spacing.

- fracture toughness gets benefit from the ex -β grain structure and from the needle shape of primary α phase (crack branching). For the same reasons long crack propagation rates is good (roughness induced closure).

- L.C.F. is good due to the fineness of the α phase precipitates.

Although processed from the (β) field, these alloys do not offer suffer from the problems mentionned when reviewing the near-α alloys : since β transus are low, β processing is performed at temperatures much lower than for the near-α alloys, and grain control appears to be effective. Also, the α phase does not precipitate in "colonies" of plates narrowly spaced, with cristallographic relationships between colonies but in a relatively fine and dispersed manner (let us note that these alloys have a good hardenability contributing to α phase precipitation control). Their limitation of course lies in the mediocre resistance of the β phase to high temperature deformation which has restricted their use to around 400°C.

3. NICKEL BASE SUPERALLOYS

Ni base superalloys are used as disc materials in the hottest section of HP compressors or in turbines where they operate at highly stressed areas in a range of temperatures extending approximately from 400 to 650°C. This range covers two relatively different sets of behaviour leading to potentially different sets of conclusions as regards the influence of microstructure and processing conditions upon damage tolerance :

- lower temperatures where interactions with the environment are relatively weak. The tendancies will be best analyzed by reviewing data obtained at 20°C

- higher temperatures where environmental influences upon crack growth may become extremely strong : data obtained at 650°C shall be reviewed. This area is perhaps the most challenging as regards damage tolerance of disc alloys.

The damage tolerance issues shall be best discussed by analysing the behaviour of superalloys in the three stages :

- crack incubation,

- short crack growth (SCG),

- long crack growth (LCG).

3.1. Lower temperature range

3.1.1. Long crack growth (LCG)

This subject has been extensively documented : in the range of ΔK of interest for modern discs, LCG is characterized from a mechanical point of view by the existence of a threshold stress intensity for crack growth and by a strong influence of R-ratio upon this threshold with high R ratio lowering it : see for instance fig. 8 from ref. (12) concerning Astroloy PM with a necklace microstructure.

As regards metallurgical parameters, a strong influence of microstructures on both threshold values and crack growth rates just above the threshold is reported. This microstructural influence appears however to vary with R- ratio : on Astroloy PM again, data from ref. (13) presented on fig. (9) shows that :

- at low R-ratios (0.1), large grains (50 µm) or a necklace microstructure (40 µm Rx, 5 µm URx) leads to threshold of respectively 12 and 10 MPa √m whereas fine grains (≃ 10 µm) give a threshold ΔK of approximately 5.5 MPa √m. Slightly above 12 MPa √m, FCG rates may vary by more than one order of magnitude.

- at high R-ratios (0.8), thresholds for all three microstructures fall within a much narrower range of values, from 5.5 to 6.5 MPa √m, and FCG rates vary within a 1 to 3 ratio.

These results and others on INCO 901, Waspaloy (ref. 13, 14) have been rationalized in terms of roughness induces closure : in the near-threshold range of FCG, Ni base superalloys exhibit strongly inhomogeneous deformation creating a facetted fracture path along (111) planes (ref. 15). Surface roughness is induced, contributing to crack closure as first suggested by Mc CARVER and RICHTIE (ref. 16) for Ni base superalloys.

The scale of the roughness is more or less directely related to microstructural features and primarily to the grain size as evidenced by VENABLES and al. (ref.13) : for Astroloy the size of asperities is of the order of the grain size in coarse grains (50 μm) as well as in fine grains materials (10 μm) : see fig. (10). INCO 901 with a coarse grains microstructure appears to deviate from this behaviour with width of asperities much lower than one grain : this is related by the investigators to the high amount of residual work left by thermomechanical processing within this material, which makes it relatively easy to activate various slip bands within one grain. Most interestingly, close values of thresholds stress intensities are found for Astroloy and INCO 901 of similar roughnesses but with grain sizes varying almost in a 1 to 10 ratio.

3.1.2. Short crack growth (SCG)

The physically small cracks (from a few tens to a few hundredths of microns) are the most relevant to damage tolerance of turbine discs since most of the life of the parts is spent in this range and also since they are at the limit of what is presently controllable. by N.D.T.

Again, we shall consider data on Astroloy (ref.12, 14, 17) concerning either bidimensional cracks (12) or natural tridimensional short cracks (14, 17). The main features of SCG behaviour are well known : thresholds are not observed or only at very low ΔK. At low R-ratios, crack growth in the low ΔK range tends to occur at much higher rates than observed for long cracks and comes close to rates measured at high R-ratios for long cracks : fig. 11 and 12 . R-ratio influence is reduced and microstructure influence is also minimal as evidenced by the data reported on fig. 13 from various sources : the scatter in FCGR for a variety of microstructural conditions, test specimens, maximum load lies within a factor of three.

The departure from the long crack behaviour may in some cases - (ref. 12) bidimensional short cracks - be fully explained on the basis of a closure argument : closure being negligible for very short cracks and progressively growing with crack length. In other cases, closure effects do not account for larger SCG rates in the low ΔK range even relative to da/dN versus effective ΔK curves. For long cracks : it has been suggested that the local conditions of plasticity at the tip of short cracks created much higher effective R values than the nominal ones (ref. 14).

This overall picture has led some authors to define a microstructure insensitive regime for short crack propagation (ref. 14). Recent evidence (ref. 18) tends to show that great variations in CGR may be found for synthetic alloys differing widely in some of their microstructural features : lattice mismatch, antiphase boundary energy, volume fraction of γ' precipitates, sizes of grains and γ'.

The conclusion of this latter study is that low crack growth rates are promoted by planar reversible deformation. At this point, it should only be stated that such a requirement would be contradictory with the conditions leading to good LCF behaviour. Clearly more work is needed to understand these effects and to assess their engineering significance as regards alloy design and microstructural optimization.

3.1.3. Crack incubation

Crack initiation in superalloys cyclically tested may occur at two qualitatively different kinds of sites :

- at heterogeneites such as inclusions, clusters of carbides, pores, ... which may be considered as "defects" : this topic is reviewed in another paper of this conference,

- in the grains, generally by the formation of slip bands along (111) planes at the surface of the test specimens.

The frequency of failures at these sites depends upon several parameters some of them being intrinsic to the material - grain size, distribution of defects - , and some of them being extrinsic : stress level, state of stress, ... Initiation at defects is especially relevant for high strength PM materials whereas initiation at slip bands is the dominating mode of failure in medium strength conventionally processed (cast + wrought) materials. Numerous studies in INCONEL 718, Waspaloy, INCO 901 have shown that grain size is the primary microstructural parameter controlling LCF lives. In a context of life assessment of the rotating parts based upont LCF this has progressively led to the development of very fine microstructures (< 5 µm), high strength materials (static and creep), with excellent LCF resistance. The high operating stresses of the discs designed with these new materials restricted their damage tolerance potential. However at the lower temperatures, with a limited influence of microstructure upon short crack growth, the crack propagation rates were sufficiently low to make that concept a viable one.

3.2. Intermediate temperature damage tolerance (up to 650°C)

3.2.1. General features

The problems associated with creep-fatigue crack growth at intermediate temperatures have been widely investigated in the last ten years (ref.19, 20) as they appear to be critical for the design of damage tolerant rims of modern disc engines. The phenomenon has been studied essentially on long cracks. Its main features are the following :

- a transition from a transgranular to an intergranular mode of fracture when the cyclic frenquency of tests is reduced or when hold times are introduced within cycles,

- a dramatic increase in CGR in the intergranular regime of crack propagation when cyclic frequency is reduced or hold times increase : see for instance fig. (14) from ref. (20),

- the intergranular mode of CCG appears to be environmentally (oxygen) controlled since it does not occur in vacuum, or inert environments (He..).

- a wide spread of the data is obtained for different alloys, and even for different heats and microstructures of a given alloy. The relevant variables appear to be chemical composition, microstructure - mainly grain size - strength.

3.2.2. Transgranular regime :

GAYDA and al. have reduced a large set of data concerning CGR at a relatively high frequency (.33 Hz) of Astroloy, Rene 95 and IN 100 of varying grain sizes (3 to 500 µm) and yield strength (680 to 1260 MPa) in a relationship which shows the beneficial effects of increasing grain sizes and yield strength : fig. (15) from ref. (21)

$$\frac{da}{dN} = A\ (\sigma_y^{-1}\ d^{-1/2})\ \Delta K^M$$

The inverse yield strength - crack growth rate relationship is consistent with most low temperatures models of crack propagation.

In this regime chemical composition would appear to have no influence other than via strength.

3.2.3. Intergranular regime :

The variability of the phenomenon with metallurgical parameters is very large as evidenced from data collected from various authors and from work at SNECMA reproduced on fig. (16).

Several parameters have to be considered to account for that variability :

Grain size remains an important factor as demonstrated by J.P. PEDRON and A. PINEAU (ref.28) on INCONEL 718 and reported in fig.14. Most favourable microstructures appear to be large grains and necklace microstructures tested in the L.T. directions as they favour crack branching. Microstructures caracterized by heavily elongated grains are often seen to induce deviations of large cracks propagating in a plane transverse to the direction of grain elongation. GAYDA and al. confirm this predominant effect of grain size in Rene 95 ; see fig. (17) from ref. (21). Clearly from their data, yield strength does not appear to be a prime parameter.

Other variables investigated are precipitate size (ref.22) : it would appear that overaging improves FCGR.

Chemical composition is expected to play an important role as suggested by the environmental dependance of the phenomena. This seems indeed the case . BAIN discuss the role of main elements (Co, Mo being beneficial) and of minor additions such as B or Zirconium (ref.22). 100 ppm of boron for instance are seen to lower the crack growth rates in a high strength superalloy by an order of magnitude in the work of PAINTENDRE and al. (ref.23). In Waspaloy part of the microstructural effects have been rationalized in terms of the dependance of Boron concentration in the grain boundaries with grain size (ref.24). The best commercially available alloys appear to be Waspaloy and Astroloy. Recent developments have been aimed at developping compositions especially optimized for resisting to environmentally assisted creep-fatigue crack growth (ref. 25) : such work is extremely rewarding and points out to the need of more studies to understand the interplay of chemical and micromechanical factors in the crack growth behaviour of Ni-base superalloys.

3.2.4. The short crack picture

Short crack in creep-fatigue situations has been much less studied than long cracks. Fig.18 presents some results obtained at SNECMA concerning the influence of microstructure upon the CGR of Astroloy in creep fatigue conditions. It can be seen that a 1 to 6 variation in CGR is induced by microstructural changes from a coarse grain structure (Hip + Forge) to an ultra-fine microstructure (Extrude + isoforge, $\simeq$ 3/5 µm). The necklace microstructure (Hip + Forge) being almost to the level of the large grain structure.

One point needs further investigations : the behaviour at low ΔK. PELLOUX and al. (ref.26) on creep fatigue of Astroloy, DIBOINE and al. (ref.27) on creep crack growth of INCONEL 718 have clearly demonstrated the influence of test procedures on stage I of creep crack growth : more work needs to be done to investigate the nature of the thresholds observed for long crack propagation : Are the cracks arrested for mechanical or chemical reasons ? Down to which level of ΔK, may the cracks be restarted by "pure" fatigue cycling ? How do short cracks behave in these conditions ?

3.3 Processing of superalloys for damage tolerance

The evolution of the operating conditions brings the discs of the newer engines in the range of temperatures where environment assisted crack growth occurs. From the previous review, it may be seen that the locus of microstructural optimization has to be displaced from where it was, based upon low temperature LCF (durability) requirements.

Fine grains microstructures are reconsidered and new concepts of thermomechanical processing are being explored. We shall briefly review three of them :

3.3.1. T.M.P for necklace structures

As seen previously, the necklace microstructure offers a peculiarly interesting compromise between LCF and damage tolerance. Work has been performed on PM Astroloy densified by hipping and conventionally forged : hipping above the γ' solvus allows in this material to reach a uniform grain size of around 50 µm. Subsequent forging below the γ' solvus induces a recrystallization process which initiates from the particle boundaries. A well controlled amount of strain may lead to a partial recrystallization and a necklace microstructure : warm-worked grains surrounded by fine grains.

Difficulties with this concept lie with the relatively rapid evolution of the microstructure with applied strain : too little will leave a coarse grains structure, too much a fine recrystallized structure.

3.3.2. Super solvus solution treating of wrought products

Many attemps have been done to increase the grain size of wrought products (forged or extruded) by super-solvus solution treatment. The grain growth process appears to be difficult to control under industrial conditions of temperature control. Critical grain growth is frequently observed : see for instance fig. (19).

3.3.3. Sub-solvus grain growth

Recently, a new concept of grain growth control by sub-solvus solution treating was put forward by SNECMA : it is based upon an alloy development leading to a superalloy, N18, with an elevated γ' solvus temperature - 1190°C - and a progressive solutionning of γ' within a large temperature range : see fig. (20) from ref. (25).

Normal heat-treating practice allows to reliably set, during solution treating, a residual volume fraction of primary γ' very uniformely dispersed which effectively controls grain boundary migration, resulting in a grain size comprised between 10 to 15 µm. Moreover the alloy composition was optimized in order to minimize sensitivity to environment in a creep-fatigue situation. The result is an excellent compromise between strength and damage tolerance : see fig. (21).

4. CONCLUSION

In the past, the neglect of damage tolerance has led in occasions to inadequate concepts of alloy / microstructure / processing, for disc applications. The resulting problems in operationnal service have initiated numerous studies which helped to clarify these concepts in the case of titanium alloys and of superalloys operating at relatively low temperature (550°C).

To-day, the evolution of regulations, compels the metallurgists to integrate the requirement for D.T. materials in their thinking. However new challenges are created by the operation of modern engines at higher rim temperatures than before. Progress will only be attained by the elucidation of the mechanisms of the interaction of short cracks growing in creep-fatigue conditions in superalloys under oxidizing environments.

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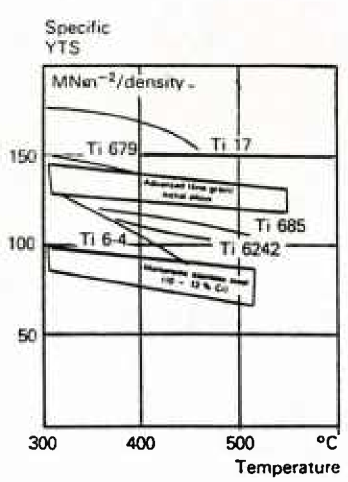

Figure 1-a : 0,2 % yield strength of Ti-Alloys.

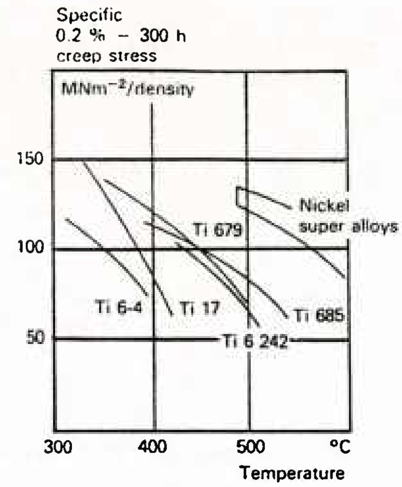

Figure 1-b : 0.2 % strain creep of Ti-Alloys.

Figure 1 : Intermediate and elevated temperature properties and applications of Ti-Alloys

β- TRANSUS	850 — 900	950 — 1000	1050 °C
ALLOY CLASS ...	β- METASTABLE	α + β	NEAR α
ALLOYS	BETACEZ Ti17	Ti Ti6.4 IMI 550 Ti 6242 IMI 685 IMI 834	
USUAL TMP SEQUENCES	β - FORGE (α + β) HEAT-TREAT	(α+β) FORGE AND HEAT-TREAT	FROM β HEAT-TREAT TO (α+β) TMP
IDEAL MICROSTRUCTURES	β MATRIX + BIMODAL α PRECIPITATION NEEDLE - SHAPED	EQUIAX PRIMARY α WITHIN β TRANSFORMED MATRIX	FROM NEEDLE OR PLATE α IN EX β GRAINS TO A STRUCTURE WITH A SMALL AMOUNT OF EQUIAX PRIMARY α .

FIGURE 2 TITANIUM ALLOYS FOR DISC APPLICATIONS MAIN TRENDS

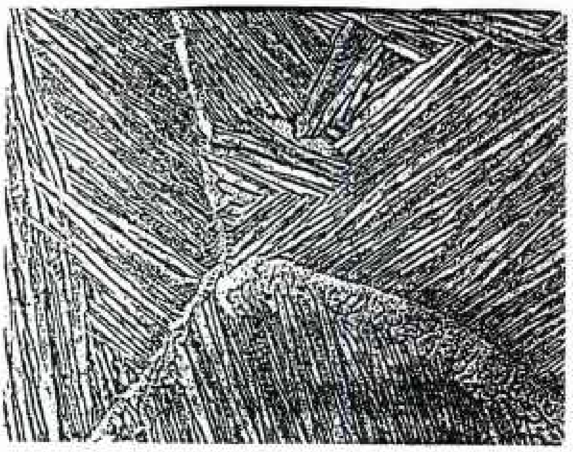

Figure 3 : IMI 685 - β Annealed
Typical microstructures

$\Delta\varepsilon = .8$ $\theta = 450°C$ $R_\varepsilon = 0$

N_R (CYCLES)	COMMENTS
13 627	SURFACE INITIATION
8 734	OU LARGE GRAINS
17 714	
15 482	
187 933	
154 271	
110 308	INITIATION ON NORMAL
194 154	SIZE GRAINS

Figure 4 : DISC cut-up LIVES FOR SPECIMEN TAKEN AT EQUIVALENT RADIAL AND AXIAL LOCATIONS.

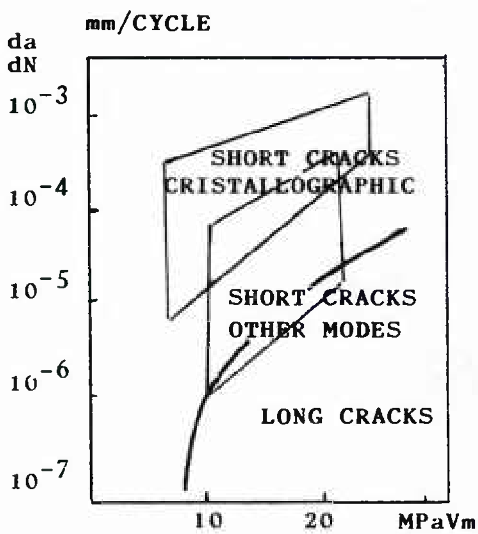

Figure 5 : IMI 685 20°C . 25 Hz
Fatigue crack growth rates from ref.7

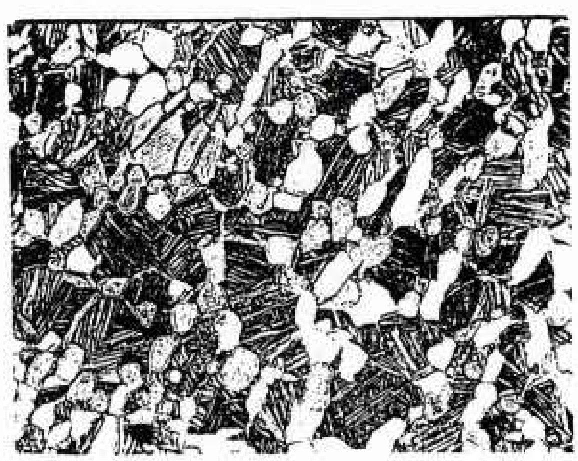

Figure 6 : Ti 6.4 α + β TMP
Typical microstructures

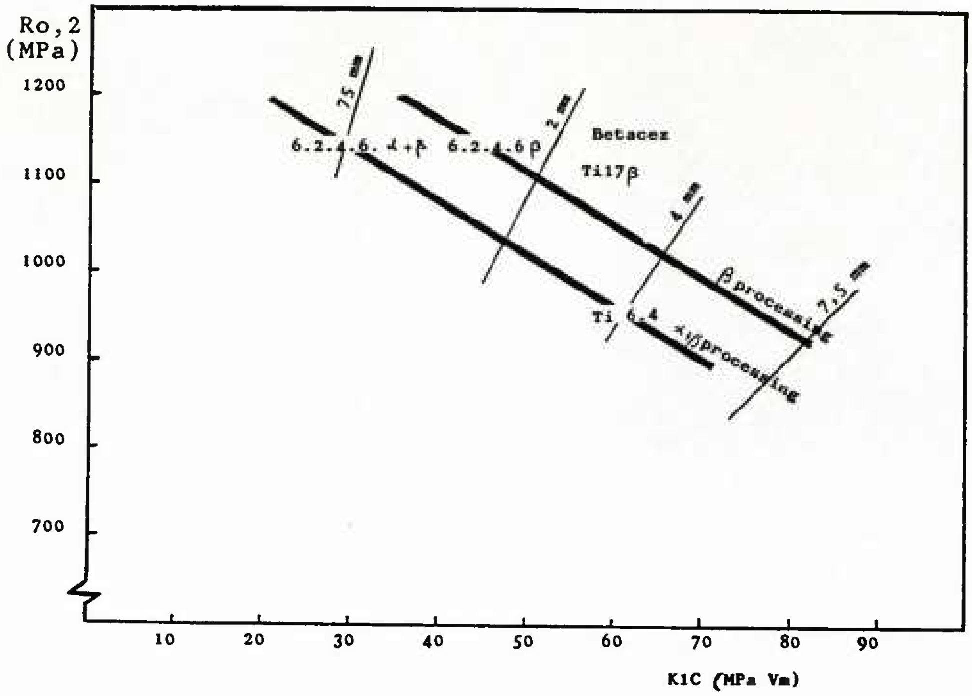

Figure 7 : TITANIUM ALLOYS FOR DISC APPLICATIONS
FRACTURE TOUGHNESS VERSUS YIELD STRENGTH 20°C
CRITICAL DEFECT SIZES

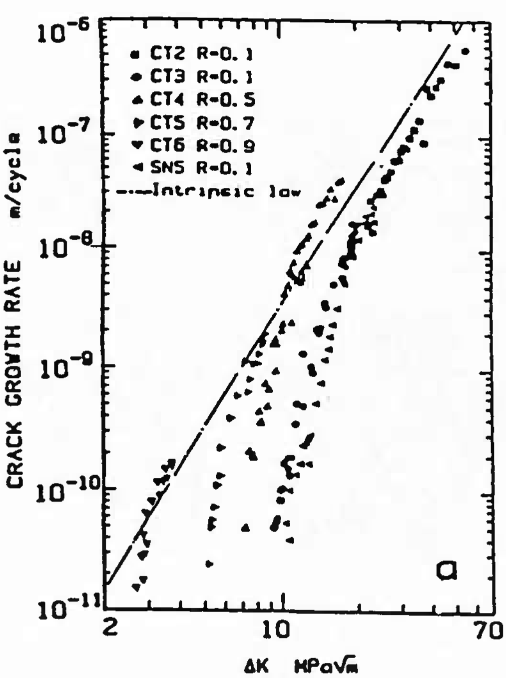

Figure 8 : Astroloy, Necklace microstructure - Fatigue crack growth rates 20°C 30 Hz influence of A-ratio from ref. 12

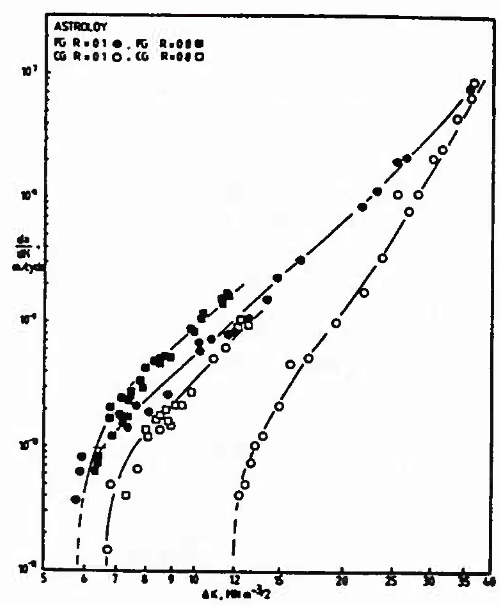

Figure 9 : Astroloy
FG Fine Grain 50 um
CG Large Grain 10 um
Fatigue crack growth rates 20°C 50 Hz - Influence of microstructure from ref. 1

ROOM TEMPERATURE THRESHOLD VALUES

	ΔK_{Th} (MPa $\sqrt{m}$) R = .1	ΔK_{Th} (MPa $\sqrt{m}$) R = .8
ASTROLOY CG	12	6.7
ASTROLOY FG	5.8	5.8
NIMONIC 90/C	10	6
NIMONIC 90/F	7.8	5.1

SURFACE ROUGHNESS IN CRACK WAKE

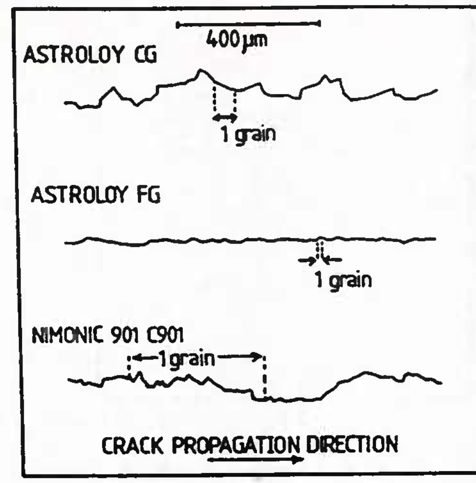

Figure 10 : Relationship between thresholds and crack surface roughness (from ref. 13)

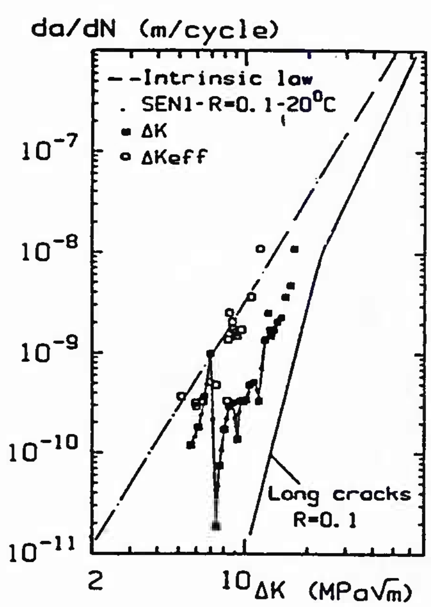

Figure 11 : Astroloy Necklace 20°C R = 0.1 30 Hz Fatigue crack growth rates of short crack from ref. 12

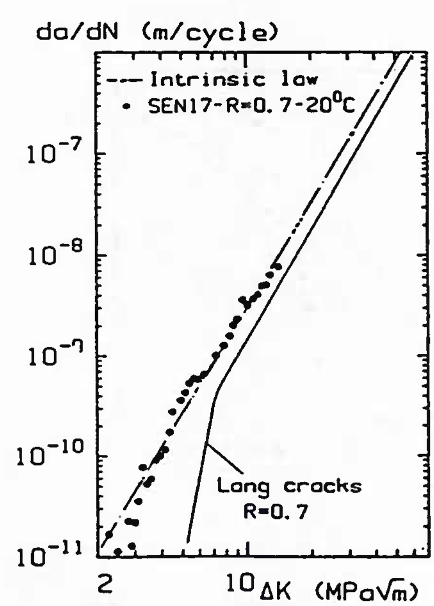

Figure 12 : Astroloy Necklace 20°C R = .7 30 Hz Fatigue crack growth rates of short crack from ref. 12

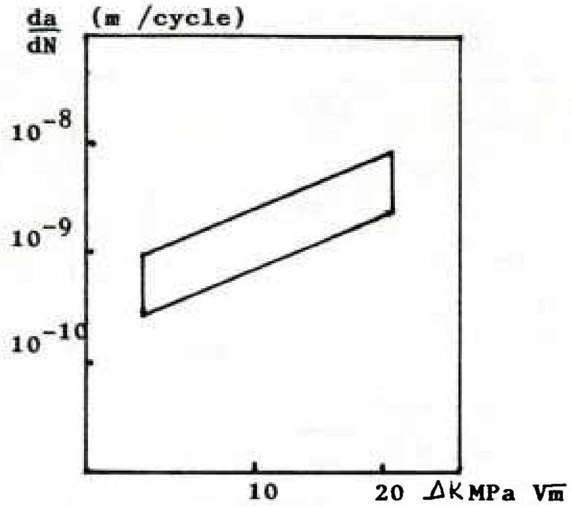

Figure 13 : Fatigue crack growth rates of short cracks data covers various microstructures (Necklace, coarse grains, fine grains), processing routes and laboratories

Astroloy 20°C

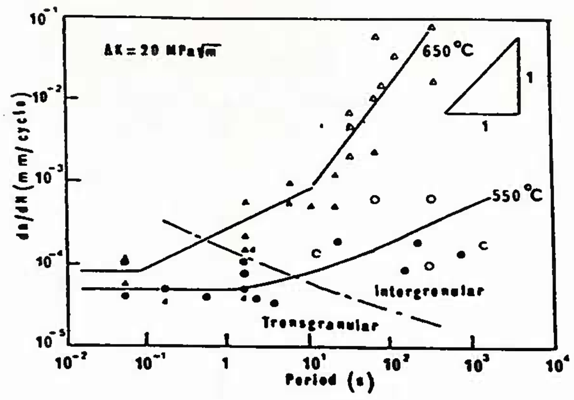

Figure 14a : Inconel 718 Fatigue crack growth rates versus Fatigue cycle period. data compiled in ref. 20

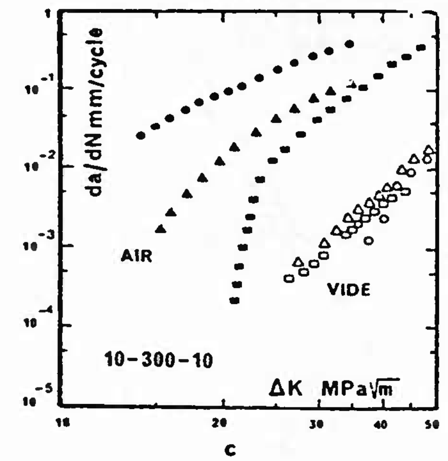

Figure 14b : Inconel 718 650°C Creep-fatigue crack growth

● Fine grain

▲ Coarse grain

■ Necklace

from ref. 28

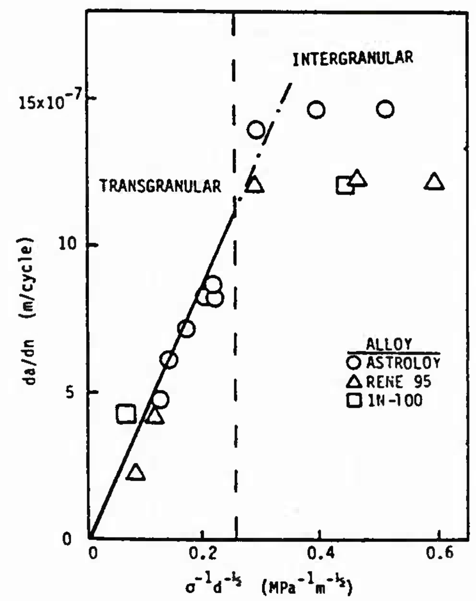

Figure 15 : Fatigue crack growth rates at 650°C, 33H3, R = .1 , ΔK = 30 MPa √m. Grain size from 3 to 500 um Yield strength from 680 to 1100 MPa from ref. 21

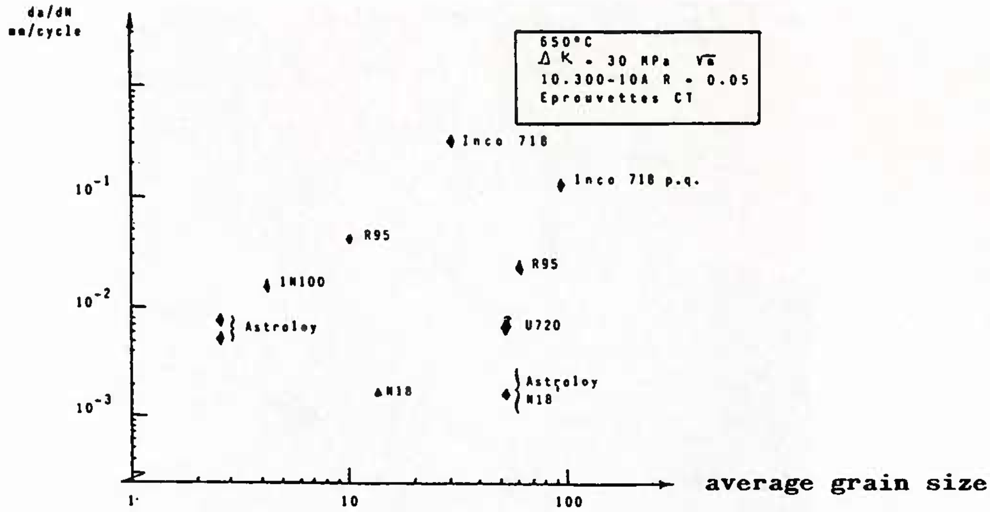

Figure 16 : Fatigue crack growth from various alloys SNECMA data + litterature

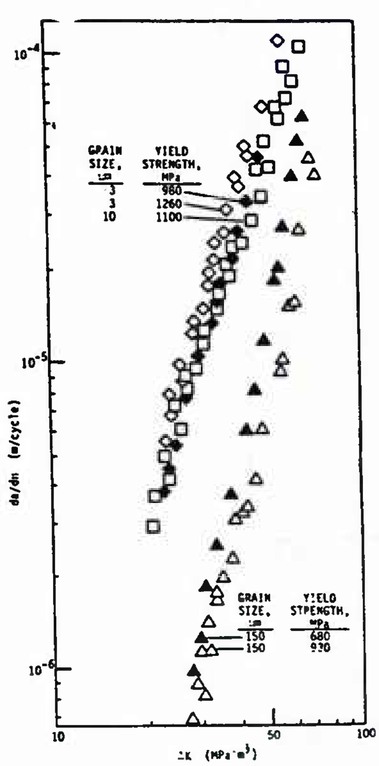

Figure 17 : Creep-fatigue crack propagation data for Rene 95. Tests were run in air at 650°C using a 120 sec. dwell at maximum load. From ref. 21

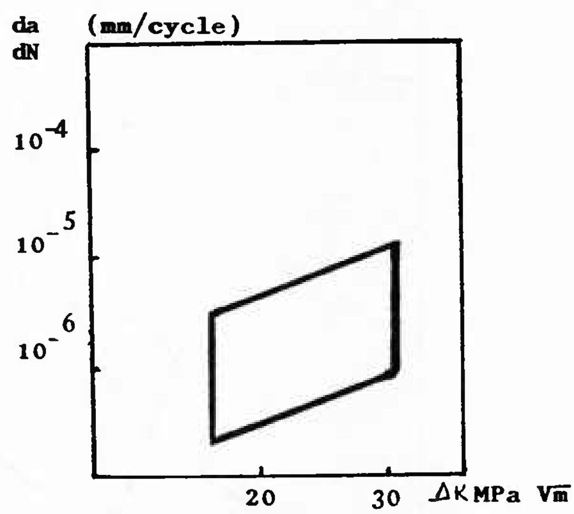

Figure 18 : Fatigue crack growth rates of short cracks data covers various microstructures (Necklace, coarse grains, fine grains), processing routes and laboratories

Astroloy 650°C

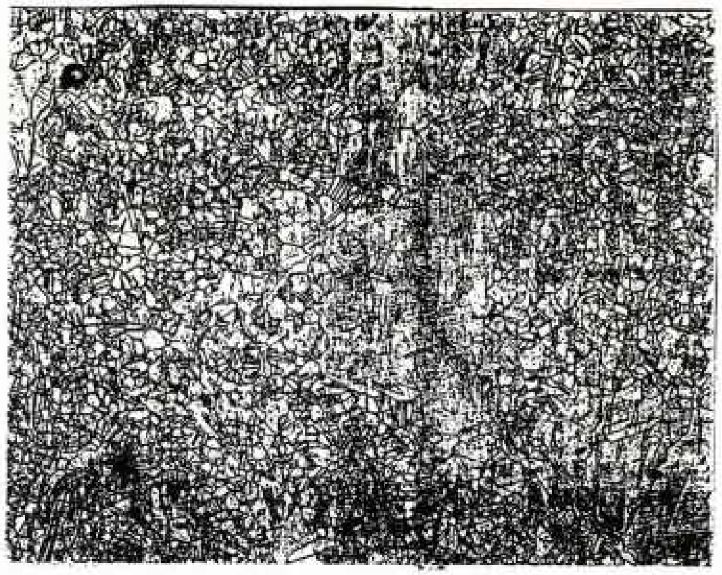

Figure 19 : Critical grain growth in
Astroloy super solvus solutionned

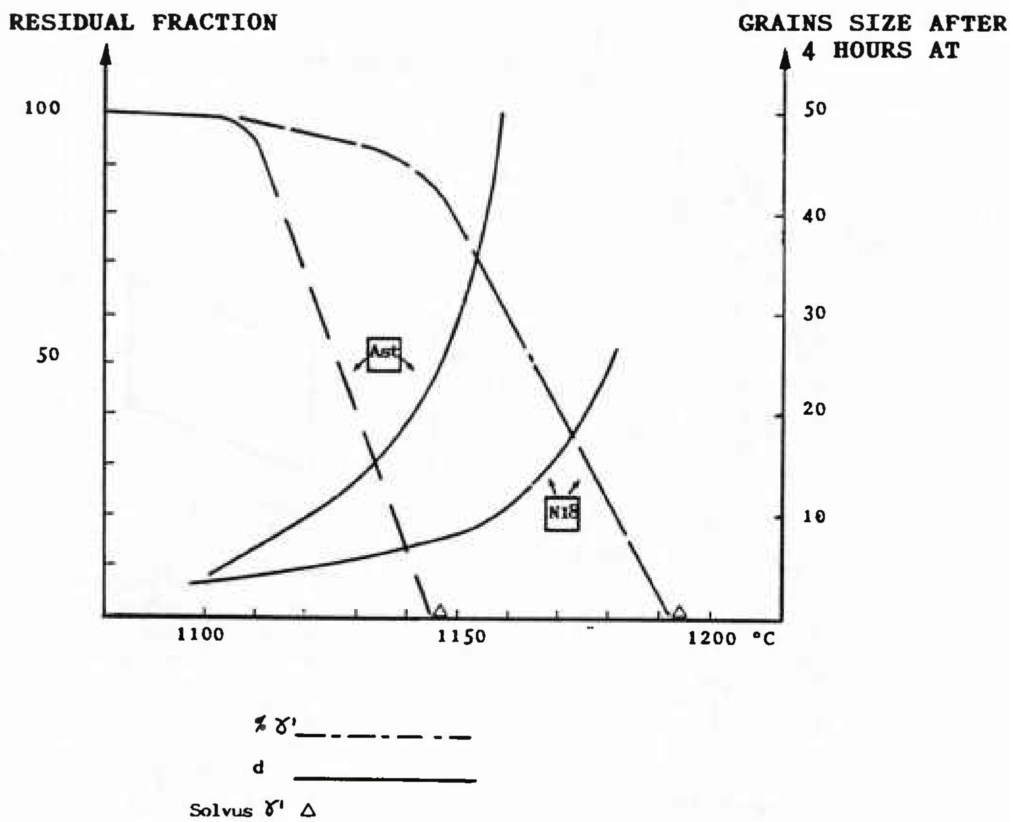

Figure 20 : Controled grain growth by progressive solutionning of primary γ' 4 hours at θ

REVIEW OF COMPONENT DEFECTS, THEIR RELATIONSHIP TO MANUFACTURING PROCESS AND THEIR EFFECT ON MATERIALS BEHAVIOUR (FOR CRITICAL ROTATING PARTS)

A C Pickard

Rolls-Royce plc
PO Box 31
Derby, DE2 8BJ - UK

SUMMARY

All engineering materials contain defects. These may range from microstructural inhomogeneities through to gross abnormalities which are independent of the underlying microstructure. The key to understanding the importance of these defects is the determination of their effect on full scale component behaviour.

In general, the most dangerous defects are those which, when subjected to cyclic loading conditions, behave as propagating cracks, with little or no crack nucleation life. "Danger" is a relative concept here, however; for instance, where the density of such defects is high, and the size uniform, tests on specimens and components may automatically take into account the presence of the "worst defect". If the equivalent initial crack size is small, use of the material will result in a component with a viable life.

When the presence of a defect is an infrequent occurrence, however, it is unlikely that specimen and component tests will cover the behaviour of the defect. In these circumstances, it is necessary to assess the risk of failure from defects and to ensure that this is acceptably low for in-service components.

INTRODUCTION

The traditional Engineering approach to designing with materials is to treat a component as "defect-free" if non-destructive inspection of the part results in no indications being detected. This approach is acceptable for low yield strength, tough materials, where local yielding around defects tends to suppress the formation or growth of cracks, and the material may be considered to be "defect tolerant". This is particularly true where a good standard of inspection is used, resulting in the rejection of components containing gross defects, which could behave as cracks even at the low levels of stress to which the material is subjected.

There has been an increasing trend over the years to move to higher yield strength materials and to use these materials at considerably higher stress commensurate with their improved tensile capability. This has been the case in particular for materials for aerospace applications, where there is a continuous pressure to reduce component weight by operating at these higher allowable stresses. Unfortunately, almost without exception these materials, when subjected to the higher stress levels allowed by traditional design procedures, cease to behave in a "defect tolerant" manner; in addition, the absence of detected indications during non-destructive inspection cannot be taken to mean that the material is "defect-free". This lesson has in general been learned the hard way; fracture mechanics methods were originally developed to understand the defect sensitivity of high strength materials used at high stresses, in particular where this had resulted in a catastrophic in-service failure.

The modern approach to designing with materials is to start from the premise that all engineering materials contain defects; the key to predicting component life and behaviour is then to develop an understanding of the tolerance of the material to the presence of these defects. In the case of cyclic loading, this corresponds to understanding the nucleation and propagation behaviour of fatigue cracks originating at or near defects.

DEFECT TYPES

Table 1 illustrates the variety of defects which may occur in a component, categorised into subsurface and surface, microscopic and macroscopic types. If the material cannot be considered to behave in a "defect tolerant" manner, all of these types have to be taken into account during the calculation of component cyclic life.

All of the subsurface defect types identified may occur in deep subsurface, near surface and surface-intersecting forms. In general, it is found that near surface defects are potentially the most dangerous. This is because these are harder to detect during non-destructive inspection of the component than surface intersecting defects, and deep subsurface defects tend to have longer crack nucleation and propagation lives than near-surface defects. This aspect is discussed in more detail later, when some aspects of defect modelling are considered.

The boundary between microstructural features which may be described as "defects" and those which may be considered as a normal part of the microstructure is usually defined by observation of crack nucleation behaviour in the material. If cracks are normally observed to nucleate at and grow from a particular microstructural feature, for instance carbide particles, that feature must be considered to be a "defect" in the microstructure. Figure 1, taken from reference 1, illustrates the point and shows the formation of cleavage cracks below a notch in a mild steel specimen (1). Microcracks are observed to nucleate in carbide particles and to propagate into the matrix by a cleavage mechanism; the carbide particles, which are a normal part of the microstructure of the material, may be considered as defects for this particular fracture mechanism, and statistical correlations have been produced between carbide thickness and fracture energy for this type of material (2).

The boundary between microscopic and macroscopic defects is normally defined by considering the relationship of the defect size to microstructural dimensions. From a component life prediction viewpoint, however, a more convenient definition of this boundary is obtained by considering the frequency of occurrence of the defect type. For most materials, microscopic defects, when they occur, are normally a regular feature of the microstructure. Hence a full scale component test will "sample" a large number of microscopic defects, and the behaviour of the test component can be taken to be representative of the behaviour of components in service. A prerequisite of this "read-across" is that sufficient controls must be applied to the material manufacturing process to ensure consistency of the defect distribution statistics.

Macroscopic defects are, for most materials, an infrequent occurrence. The presence of gross inclusions, segregation or porosity, or of forging cracks, in a material is normally associated with a loss of material manufacturing process control, which while it may affect a batch of components, is usually observed in practice to be a "one-off" problem. The normal lifing approach in these circumstances is to "inspect out" the worst defects and to perform a risk analysis to ensure that the probability of component failure from the particular defect type is "extremely remote". This approach is discussed in more detail later.

For a well controlled component manufacturing route, with appropriate precautions taken during component storage and engine assembly, macroscopic surface damage marks and machining imperfections should be a rare occurrence. In these circumstances, the statistical risk analysis approach may be used. An alternative approach which has been developed in recent years is the damage tolerance method, whereby components are subjected to non-destructive inspection before and/or during service with the aim of identifying and removing those which show surface damage or evidence of cracking.

The correspondence between the definitions of macroscopic and microscopic defects based on microstructural size comparisons and on the life prediction approach is normally good. There are some exceptions, however. A case in point is for damage - prone materials, where the risk of a component containing surface damage is high. In these circumstances, rig testing of a number of full - scale components will automatically take account of the effects of a range of representative damage marks, and the macroscopic damage can be treated, from a life prediction point of view, as if it were a microscopic defect.

COMPONENT LIFE PREDICTION IN THE PRESENCE OF DEFECTS

Table 2 indicates the variety of control and life prediction approaches which are usually applied for the various types of defect discussed above.

These fall into three general categories:

(a) Process Control and Tests on Representative Components

This approach is used where the density of defects is sufficiently high to ensure that the tested component will contain a representative sample, and the size range is sufficiently well controlled to ensure that the scatter between "best" and "worst" ($+3\sigma$ and -3σ) component lives is not excessive.

The Predicted Safe Cyclic Life (PSCL) approach has been used for a number of years to control the cyclic lives of critical rotating parts in aircraft gas turbine engines. The method, which is described in detail elsewhere (3, 4, 5) uses a rig test or set of tests on components representative of those in service to derive the safe cyclic life of the component. The scatter in component lives is taken to follow a log-normal distribution with a life ratio of 6:1 between best ($+3\sigma$) and worst (-3σ) results. For a single rig test, the tested component is assumed to be taken from the lower 95% of this distribution, and a life factor of 4 is applied to the rig test result to predict the worst (-3σ) in-service component life. For multiple rig tests, reduced scatter factors are applied to the mean or minimum life (3). Tests may be performed at stresses higher than those present in-service; the test life is then factored accordingly based on the low cycle fatigue properties of the material from which the component is made. The life end point is taken either as a fraction of the life to rapid fracture (the preferred method) or as the life to formation of an "engineering" crack.

The databank approach (4, 5) has been developed in recent years as an extension of the PSCL approach. The results of a number of specimen and full scale component tests on different features (disc bores, fillet radii, rim features, holes etc.) are combined, using a knowledge of the fatigue behaviour of the material, to produce a fully correlated fatigue life prediction databank. Minimum component lives are then predicted directly from this databank, thus removing the need for an individual component test.

For a number of materials, it has been shown (4, 5) that fracture mechanics methods may be used to obtain this correlation, by invoking the concept of an "effective initial flaw size" associated with each fatigue test result. Statistical analysis of the effective initial flaw size distribution then allows the maximum probable value to be derived; minimum component life for a given feature then corresponds to the crack propagation life from this maximum effective initial flaw size origin. In the case of damage sensitive materials, the effective initial flaw sizes correlate well with the sizes of the damage marks present at failure origins on components.

Both the rig test and databank approaches rely on the testing of a representative sample of material, whether the fatigue origin is "conventional" fatigue crack nucleation by a surface slip mechanism or fracture of a defect. Testing of full scale components is considered to be essential in ensuring that a representative sample is evaluated; the relative stressed volumes of components and laboratory specimens are such that lives based on specimen testing alone are unlikely to be representative of in-service behaviour.

(b) Control of Material Cleanliness Combined with a Risk of Failure Analysis

When the rate of occurrence of defects is low, the probability of a component containing a defect being selected for rig test is remote. In these circumstances, method (a) cannot be used and a risk of failure analysis has to be performed instead.

Probability of failure from a defect can be considered as the combination of six statistical factors (6):

i) The probability of a component containing a critical size flaw.

ii) The probability of the flaw not being detected during non-destructive inspection.

iii) The probability of the flaw falling in a zone of the component which is sufficiently highly stressed to result in failure of the component from that flaw within the required life.

iv) The probability of the flaw showing no nucleation life (ie. behaving as a crack of the same size as the flaw from the first cycle).

v) The probability of the crack growing from the flaw at a faster rate than assumed.

vi) The probability of the fracture toughness of the material being lower than anticipated.

Full evaluation of these factors clearly requires a detailed knowledge of the behaviour of the defects and the component material, together with a full characterisation of the non-destructive inspection technique applied. The rejection rate at inspection is an important parameter in this approach, and is a measure of the cleanliness of the material.

Studies of defect failure probabilities in full scale components have shown (6) that the probability of a component containing a critical size flaw is very powerful in determining the overall risk of failure. This emphasises the need for clean melting routes for materials where fatigue and fracture behaviour are critical for components manufactured from those materials.

(c) Damage Tolerance

The Damage Tolerance method relies on the use of fracture mechanics to predict the behaviour of cracks growing from origins with size fixed by material process control or non-destructive inspection. The approach is described in detail elsewhere (4, 5, 7).

The Damage Tolerance approach is aimed at reducing the risk of failure from surface damage or subsurface defects, within the Predicted Safe Cyclic Life, by ensuring that an adequate crack propagation margin is present in components. This margin may be obtained either by designing to sufficiently low stresses to ensure that the fracture mechanics life, with safety factors applied, is longer that the service life of the component, or by performing in-service inspections and rejecting components with indications larger than those used in the fracture mechanics life calculations. The statistics of the definition of the crack propagation origin size are a major concern in the damage tolerance approach. The evaluation of the reliability of non-destructive inspection methods is a subject of current debate.

The damage tolerance approach tends to be pessimistic as a method of predicting total life, principally because no account is taken of crack nucleation life at defects. For those materials that are sensitive to surface damage or other defects, however, there is a strong link between the "process control" damage tolerance approach and the material databank method for determination of the PSCL. The effective initial flaw size approach can be considered as a form of damage tolerance lifing in these circumstances, with process control origin sizes based on "fatigue" sampling of the material.

DEFECT MODELLING

There is a continuous pressure to improve the performance to weight ratio of gas turbines for aerospace use. One way of minimising weight is to use high strength materials at high stress levels. In general, however, it is found that the sensitivity of materials to fatigue failure increases with applied stress. In these circumstances, it is essential that an understanding is obtained of the mechanisms of crack nucleation and growth from defects. Two avenues may then be pursued for increasing component life:

(a) Development of materials which are less sensitive, at a given stress level, to the presence of defects.

(b) Identification of the most harmful defects and modification of the material manufacturing process route to avoid these types of defect.

Modelling of defect behaviour is a critical part in the development of this understanding; three examples are given of models which have been developed in recent years:

(i) Porosity Models

Figures 2 and 3 show some stress concentration factors which have been derived using 3D finite element stress analysis methods for a variety of surface and subsurface pore shapes. The spherical subsurface pore (figure 2, case 1) exhibits a stress concentration factor of 2.057, which is close to the analytical solution for a spherical pore in an infinite medium (2.047) (8).
The flattened pores have an aspect ratio of 4:4:1, and consist in cross section of a parallel sided central region with a semicircular end (figure 2 cases 3 to 5); the stress concentration factors clearly depend on the orientation of the stress field. The elongated pores shown in figure 3 have an aspect ratio of 4:1:1, and consist of a cylindrical section with hemispherical end. The surface - intersecting case is representative of a blunt gouge in a component and illustrates that significant stress concentration occurs, in this case lower than but broadly comparable with that for a two - dimensional semicircular notch in a sheet (2.988) (9).

(ii) Soft Inclusion in a Hard Matrix

Figure 4 illustrates two finite element meshes used to analyse the behaviour of a soft, yielding inclusion in an elastic matrix. The first model is of a spherical inclusion, the second of a flattened inclusion with aspect ratio 4:4:1. Figure 5 examines the variation of peak stress in the matrix adjacent to the inclusion with remote stress (in the x direction) for the two 3D models and for a comparable two - dimensional model of a cylindrical inclusion. At remote stress levels below the yield stress of the inclusion, no stress concentration is seen (with identical elastic properties for the matrix and inclusion). As the remote stress is increased, however, the inclusion yields and a stress concentration effect is seen in the matrix adjacent to the inclusion. The level of concentration varies with the form of the inclusion and follows the same sequence as the stress concentration factors for the equivalent shapes of pore (see above).

Fatigue cycling of an inclusion of the form represented by the flattened inclusion mesh (figure 4(b)) resulted in failure from an origin in the matrix at the predicted peak stress position.

(iii) Hard Inclusion in a Soft Matrix

Figures 6 to 9 illustrate the results of a two dimensional finite element analysis of the yielding distribution around an elastic, circular inclusion in a soft matrix (10). In this case, the strains are developed as a result of the thermal mismatch between the inclusion and the matrix on cooling from a temperature representative of the solution treatment conditions for the matrix material. Three analyses are illustrated; these are aimed at establishing the effect on the yielding distribution of the proximity of the defect to a free surface. In the deep subsurface case (figure 6), the plastic strain distribution is uniform around the inclusion. As the inclusion is brought nearer to the surface, however, an asymmetry develops in the yielding (figures 7 and 8), with progressively more plastic strain being observed in the ligament between the inclusion and the free surface. Figure 9 shows the stress distributions in the inclusion and matrix for the near-surface case. It can be seen that the inclusion is subjected to a significant bending stress, with a high tensile peak on the side nearest to the free surface. This compares with a uniform compressive stress in the inclusion in the deep subsurface case.

The peak tensile stress in the inclusion in the near surface case may explain the greater propensity of near surface ceramic inclusions to crack and hence to act as fatigue crack nucleators, when compared to deep subsurface inclusions; the examples (c) below of defect behaviour in practice illustrate this point well.

The above examples clearly show the potential for the use of finite element stress analysis methods to understand the behaviour of defects in materials. The approach is still in its infancy, however, and requires considerably more investigation before it can be used with confidence as a predictive tool.

DEFECT BEHAVIOUR IN PRACTICE

A number of examples are given below of cracking and failure of rig - tested components from surface and subsurface defects. These show that the defect usually, but not always, behaves as a crack from the start of cycling. Current investigations are in progress to understand the differences which result in the presence of a significant nucleation life during failure from a defect.

a) Failure from Machining Damage

Figure 10 illustrates cracking from a poorly chamfered hole in the diaphragm of a rig tested Ti 6Al 4V compressor disc. The view shows the corner between the bore of the hole and the disc diaphragm. The crack origin is a step between an area of deep and light chamfering, the step height being of the order of 75 μm. The effective initial crack size required to explain the observed growth of the crack in the number of cycles applied is also 75 μm radius, implying that the damage mark has behaved as a crack from the start of cycling. Figure 11 shows another machining damage mark, again in the chamfer of a disc diaphragm hole in Ti 6Al 4V, which has acted as a fatigue crack origin. The damage mark appears to be a tear in the surface. Again, the correlation between effective initial crack size and damage mark size appears to be good (both 75 μm radius), suggesting that the damage mark has again behaved as a crack from the start of cycling.

Figure 12 shows cracking in an incorrectly blended radius between the diaphragm and drive arm of a compressor disc in Ti 6Al 4V which was subjected to rig testing. This type of failure can often be difficult to model because of the multi - origin nature of the cracking and the subsequent interaction effects. Hence great care is normally taken to avoid mismatched blend radii in highly stressed regions of components.

Figure 13 illustrates cracking from a machining damage origin in a Nickel base superalloy turbine disc, during rig test (11). The crack appears to have grown from a grain which was cracked during the machining operation; the cracking was detected during inspection and subsequently broken open to reveal the fracture face shown. Crack propagation life predictions have been performed and show a ratio of predicted to actual life of 0.92. This is consistent with the machining damage behaving as a crack of the same size from the start of cycling, although it is possible that there is a small nucleation life present.

In all of the cases shown above, the probability of this type of machining damage occurring is sufficiently high that tests on full scale components are representative of the behaviour of the worst in - service parts. Testing of representative components, combined with process control, may thus be used to determine Predicted Safe Cyclic Lives in the presence of these defects.

(b) Failure from Dents, Scratches and Other Impact Damage

Figure 14 shows cracking from an impact damage mark (a dent) on the bore/rear hub face corner of a Ti 6Al 4V compressor disc which has been subjected to rig testing. The dent is of the order of 250 μm diameter, although the depth is difficult to estimate. The effective initial crack size to explain the observed propagation behaviour is 40 μm radius, indicating that there may be some nucleation life in this case, although this clearly depends on the depth of the damage mark.

Figure 15 illustrates a damage mark which (surprisingly) has not propagated as a crack. This scratch is of the order of 12.5 mm long, but is shallow; it was observed on the rear hub face of a Ti 6Al 4V compressor disc which was subjected to rig testing.

As in case (a) above, the frequency of occurence of surface damage during full scale component testing is usually high. Hence tests on representative components, combined with inspection to remove the worst examples of surface damage, may be used to determine Predicted Safe Cyclic Lives. Alternatively, Damage Tolerance methods may be used to remove from service those components which show signs of surface damage or crack growth from damage marks during non-destructive inspection.

(c) Failure from Non-Metallic Inclusions

Figure 16, taken from reference 11, shows the growth of a crack from a just subsurface ceramic defect in a Nickel base superalloy disc during rig testing. During manufacture, a defect was detected in the bore using ultrasonics at -11dB to -12dB level. The disc was rig tested until failure occurred in the rim region. The bore crack shown was then opened. The ratio of predicted to actual life using corner crack propagation data is 1.07, illustrating that the defect behaved as a crack from the start of cycling.

Figure 17, taken from reference 5, shows a bore subsurface inclusion in another Nickel base superalloy disc. This was tested for 20,000 cycles before the disc was sectioned to reveal the inclusion. No cracking was found associated with this inclusion.

The defect modelling results described above suggest that the proximity of the inclusion shown in figure 16 to the bore of the disc resulted in cracking of the inclusion in the early stages of the test,or possibly even before testing commenced. The inclusion shown in figure 17 is not close to any free surface, and hence the non-symmetric yielding effects observed in the modelling of the near surface case will not apply; the lack of cracking and consequent high life supports the behaviour predicted from the defect model.

The types of subsurface and surface inclusions shown in figures 16 and 17 are a relatively rare occurrence in the standard of material used for aeroengine critical parts. Hence it is necessary to use probabilistic fracture mechanics, combined with control of material cleanliness, to demonstrate that the risk of failure from these types of defect is extremely remote.

Figure 18 summarises the comparison of actual and predicted lives for failures from defects in Nickel base superalloy components during rig testing. In general, it is observed that machining damage origins behave as cracks from the start of cycling. Forging cracks (a type of surface-breaking inclusion) also tend to propagate from the first cycle, but inclusion origins do on occasion show significant nucleation life.

CONCLUSIONS

1. The various types of material defects have been categorised and discussed.

2. Three methods have been identified for control and prediction of component lives in the presence of defects:

(a) Process control and tests on representative components.

(b) Control of material cleanliness combined with a risk of failure analysis.

(c) Damage tolerance

3. Process control/tests on representative components are used when the density of defects is sufficiently high to ensure that the tested component will contain a representative sample, and the size range is well controlled. The importance is emphasised of performing tests on full scale components with representative surface finish/condition, rather than on small laboratory specimens.

4. Cleanliness control/risk of failure analysis is used when the rate of occurrence of defects is low, and the probability of a component containing a defect is remote. The importance is emphasised of the need for clean melting routes as a means of reducing the risk of failure.

5. Damage tolerance is a means of reducing the risk of failure from surface damage or subsurface defects, within the Predicted Safe Cyclic Life. The statistical reliability of detection of the non-destructive inspection technique used is critical to this approach.

6. The use of material life prediction databanks is highlighted as a means of combining the results of a number of tests to increase the sample size on which minimum life predictions are based. The effective initial crack size approach is identified as a good method of obtaining this correlation of test results. The relationship with the process control damage tolerance method is noted.

7. Defect modelling is identified as an area for future development to understand the nucleation and growth of cracks from defects. The significance of nucleation life is highlighted. Future materials development will need to take into account the tolerance of the material to defects, both surface and subsurface.

8. A number of examples are given of defect behaviour in rig tested components. The importance is again emphasised of testing full scale components to obtain an understanding of the fatigue behaviour of material with representative standards of surface finish and damage.

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TABLE 1 DEFECT TYPES

TYPE → / LOCATION ↓	MICROSCOPIC	MACROSCOPIC (Gross Abnormalities)
SUBSURFACE	- Microscopic inclusions (carbides, nitrides, borides etc.) - Microporosity - Deformation structures retained from forging (heavy slip bands etc.)	- Gross inclusions (ceramics, oxides, HIDs, LADs, etc.) - Gross segregation - Gross porosity
SURFACE	- Cracked grains (machining damage) - Cracked twins (machining damage) - Surface intersecting microscopic inclusions (carbides, nitrides, borides etc.) - Surface intersecting microporosity	- Scratches - Dents - Plucks and tears - Corrosion pits - Fretting marks - Machining marks/ imperfections - Forging cracks - Surface intersecting gross inclusions (ceramics, oxides, HIDs, LADs, etc.) - Surface intersecting gross segregation - Surface intersecting gross porosity

TABLE 2 CONTROL AND LIFING METHODS FOR DIFFERENT DEFECT TYPES

TYPE → / LOCATION ↓	MICROSCOPIC	MACROSCOPIC
SUBSURFACE	- Process control + tests on representative components	- Control of material cleanliness + risk of failure analyses
SURFACE	- Process control + tests on representative components - Process control + Damage Tolerance	- Control of material cleanliness + risk of failure analyses - Damage Tolerance (NDI) - Tests on representative components

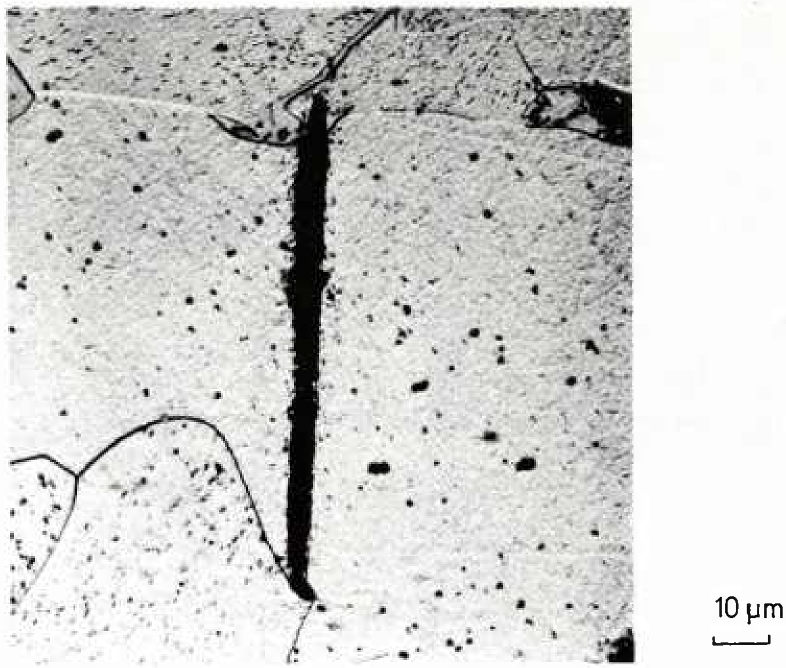


Figure 1. Formation of cleavage cracks by carbide fracture ahead of a notch in a mild steel specimen (1) (Courtesy JF Knott)

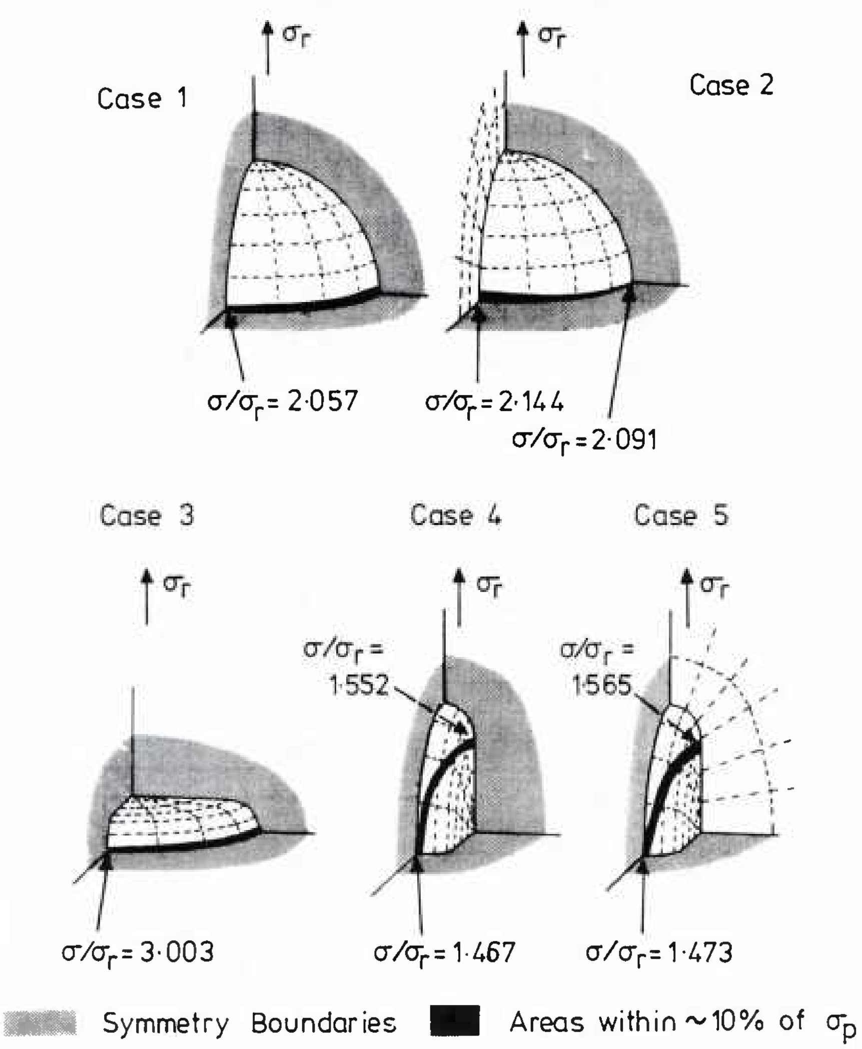


Figure 2. Stress concentration factors for surface and subsurface pores

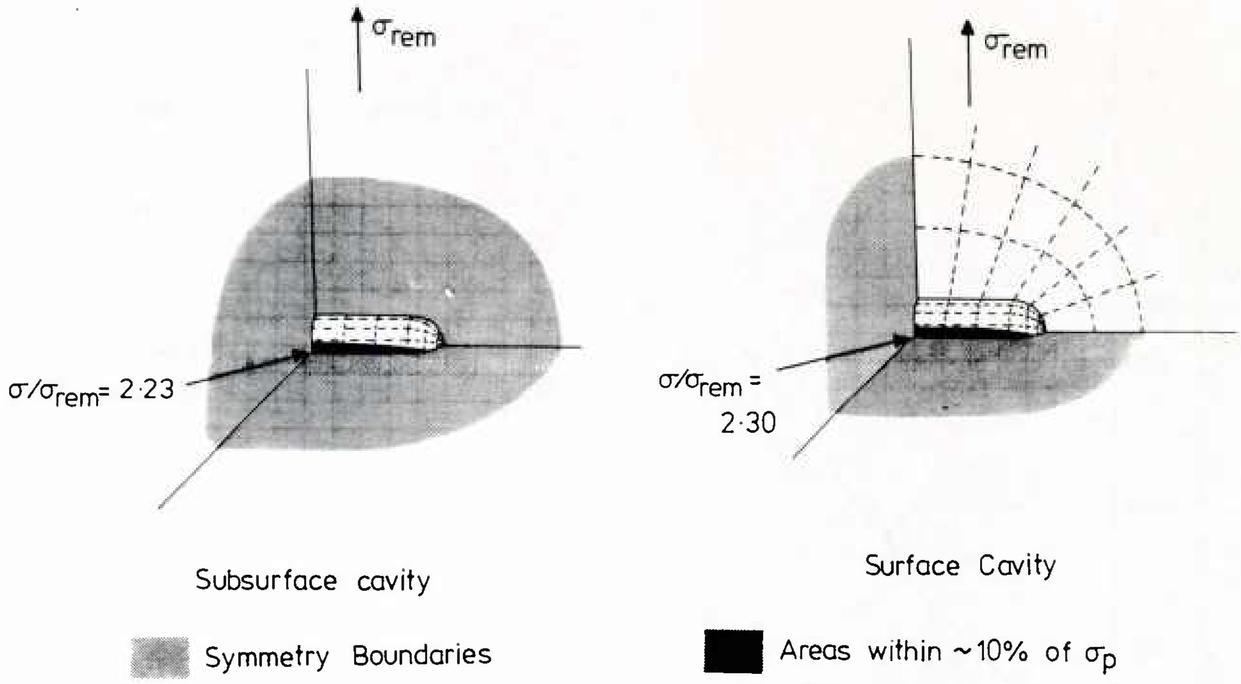


Figure 3. Stress concentration factors for surface and subsurface pores

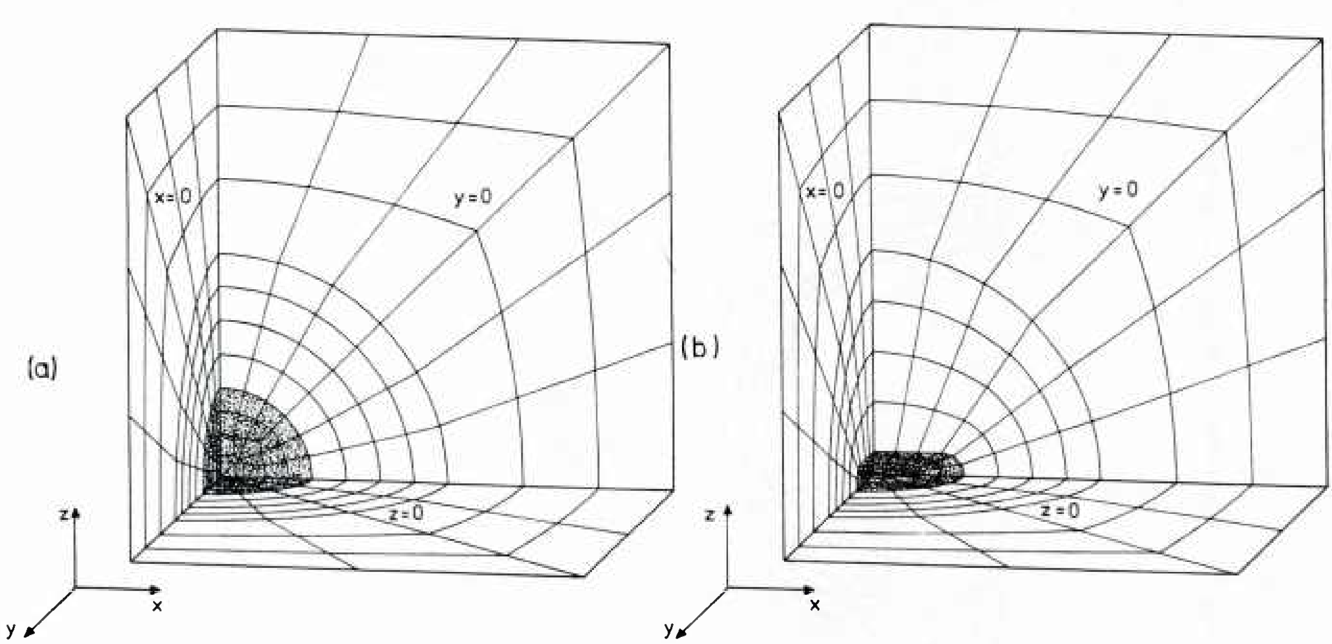


Figure 4. 3D finite element meshes for analyses of soft inclusions in hard matrices

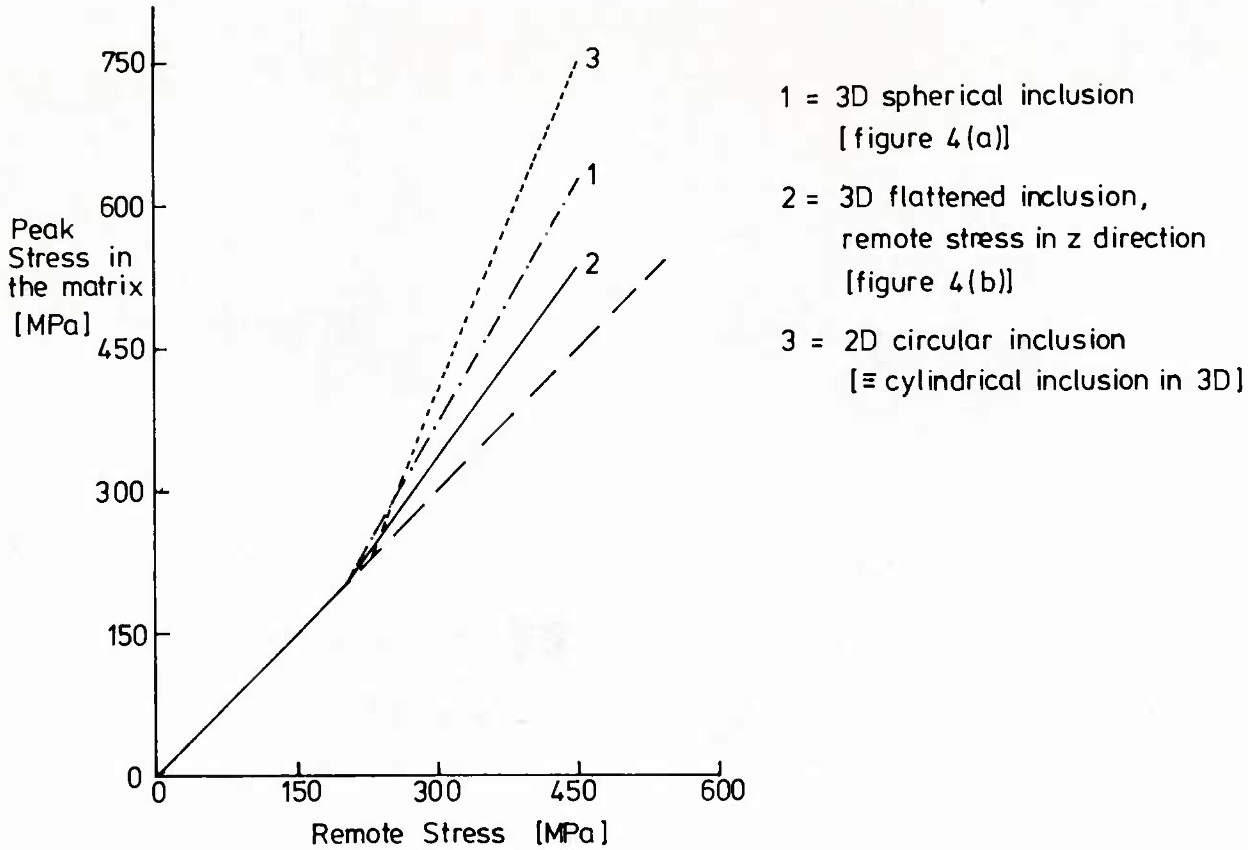


Figure 5. Peak stress in the matrix adjacent to a soft inclusion

Contour	ε_p (%)
1	0.28
2	0.57
3	0.85
4	1.14
5	1.42
6	1.70
7	1.99

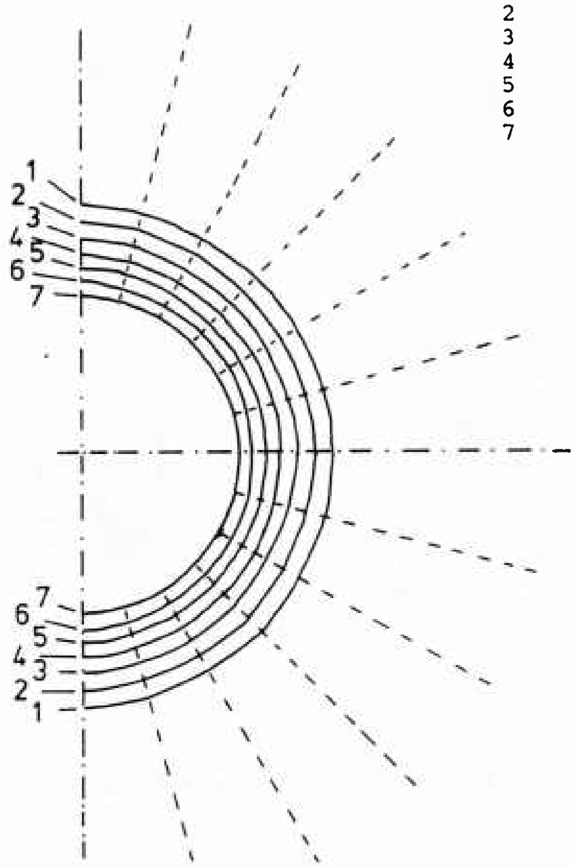
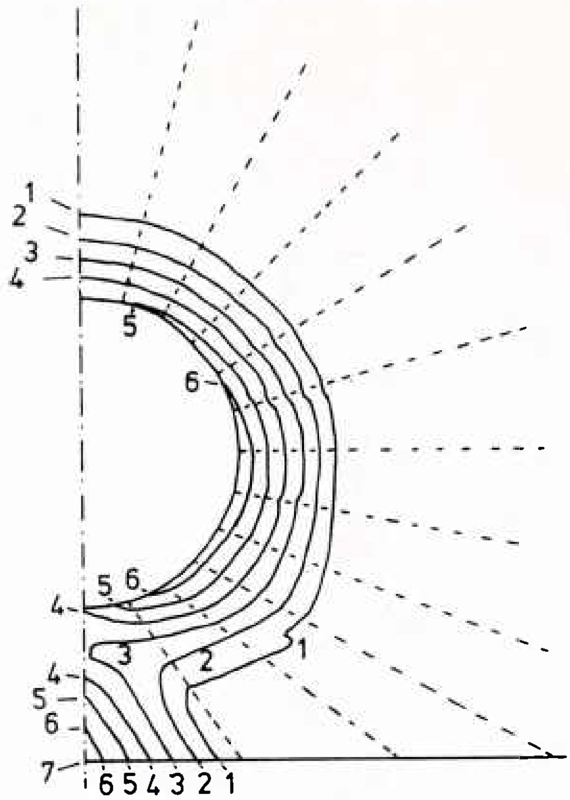


Figure 6. Subsurface inclusion - equivalent plastic strain contours



Contour	ε_p (%)
1	0.19
2	0.48
3	0.77
4	1.06
5	1.34
6	1.63
7	1.92

Figure 7. Near-surface inclusion (limb/radius ratio = 1.0) - equivalent plastic strain contours

Contour	ε_p (%)
1	0.28
2	0.73
3	1.18
4	1.63
5	2.09
6	2.54
7	2.99

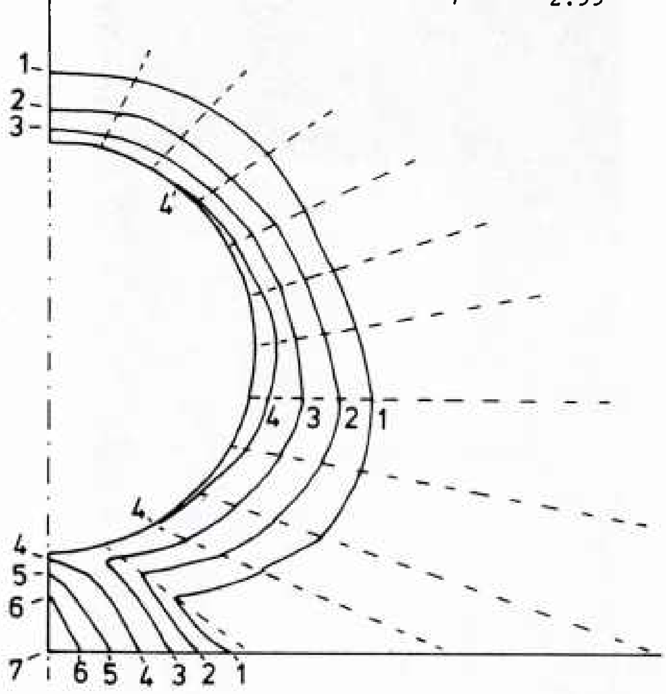


Figure 8. Near-surface inclusion (limb/radius ratio = 0.5) - equivalent plastic strain contours

Contour	σ_x (MPa)
1	-1130
2	-795
3	-460
4	-125
5	210
6	545
7	880

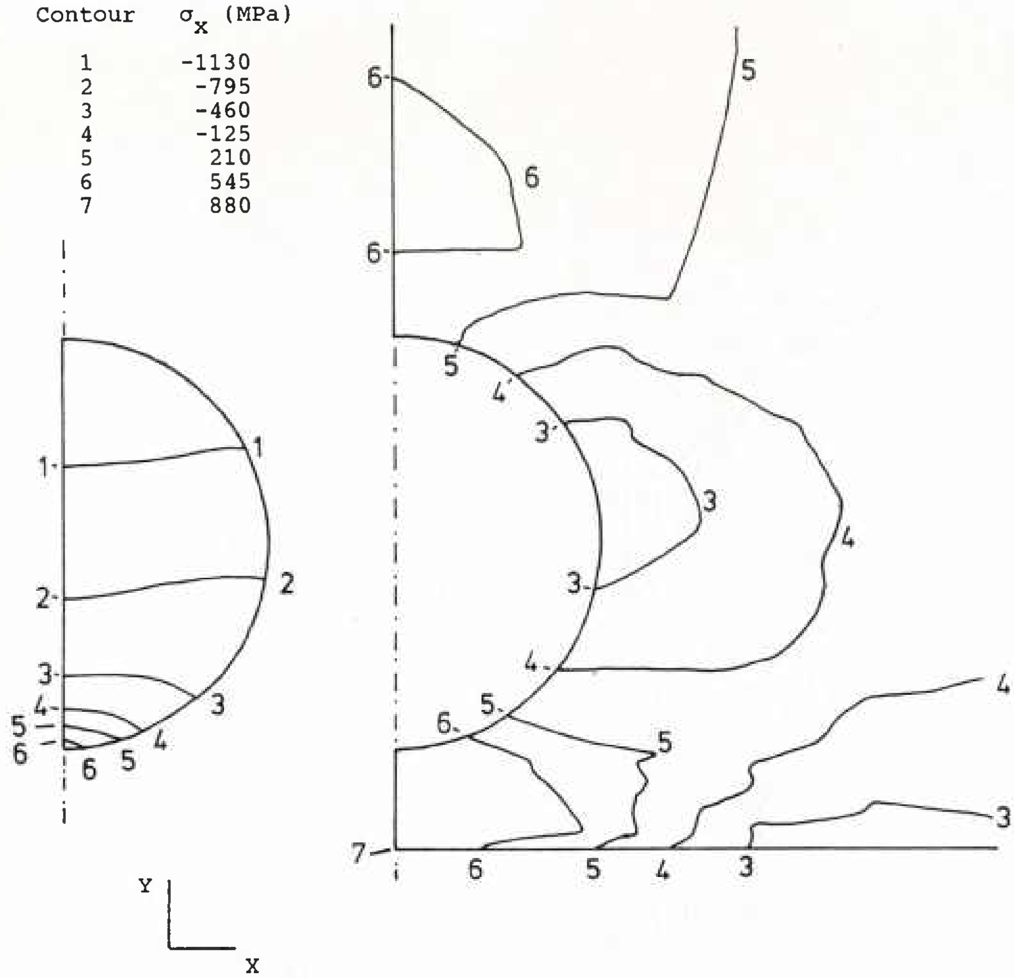


Figure 9. Near-surface inclusion (limb/radius ratio = 0.5) - x direction stress contours

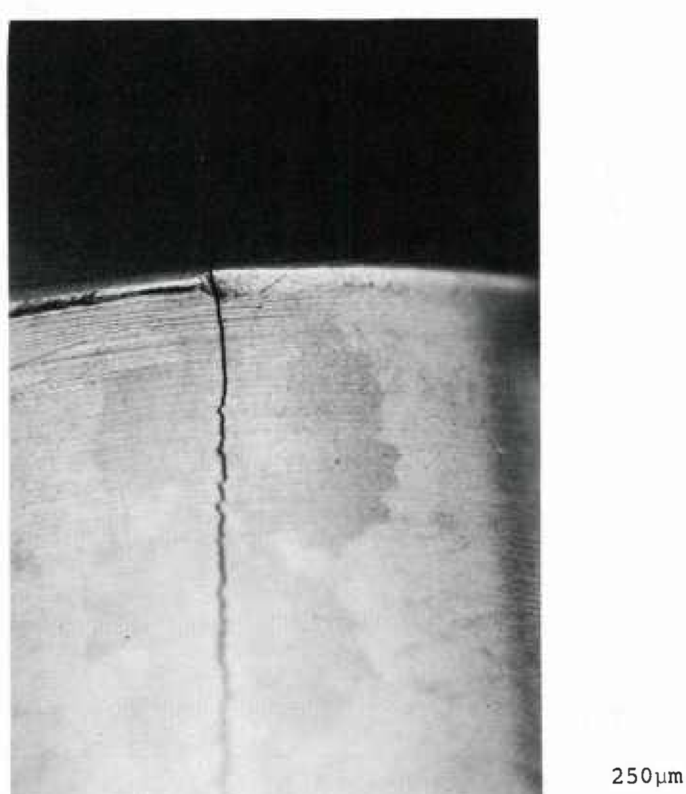


Figure 10. Crack growth from a poorly chamfered hole, Ti6Al4V

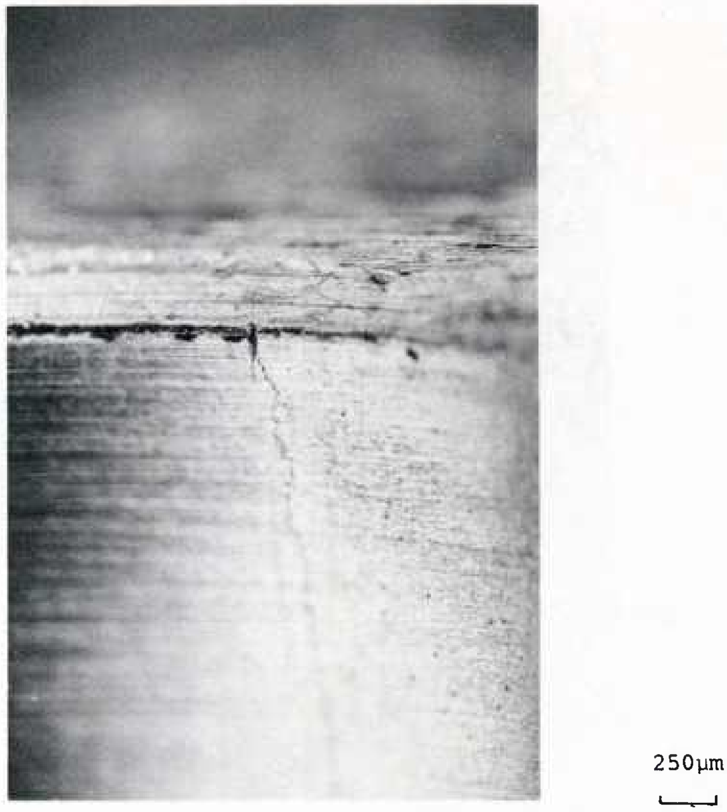


Figure 11. Crack growth from a machining mark on a hole corner, Ti6Al4V

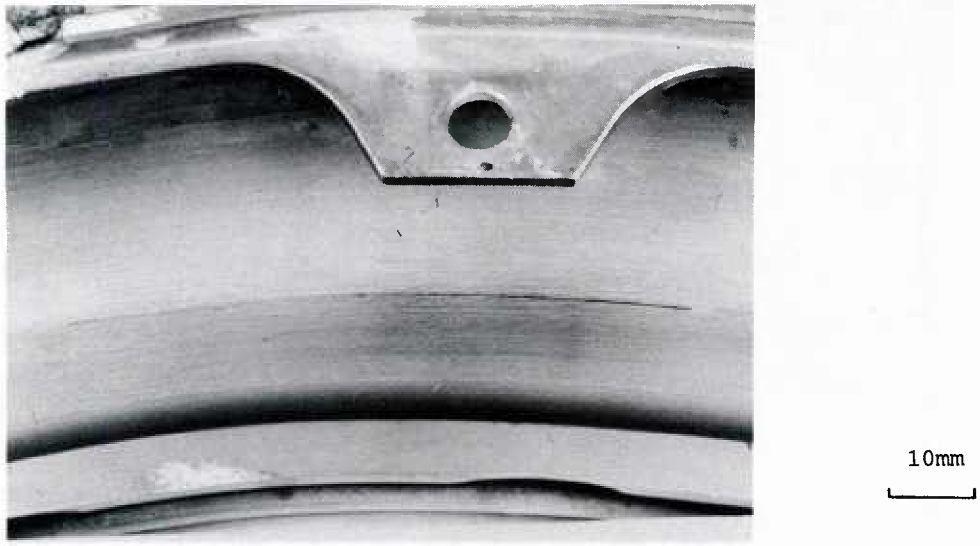
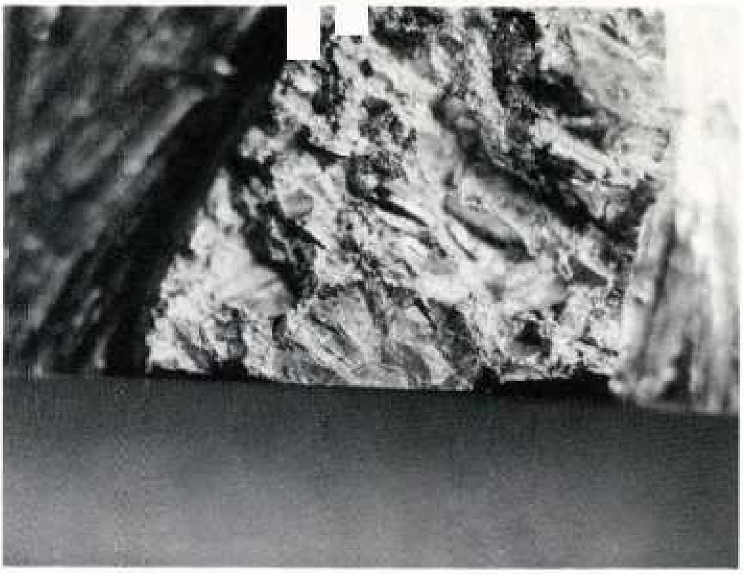


Figure 12. Crack growth from an incorrectly blended radius in a disc diaphragm, Ti6Al4V



250µm

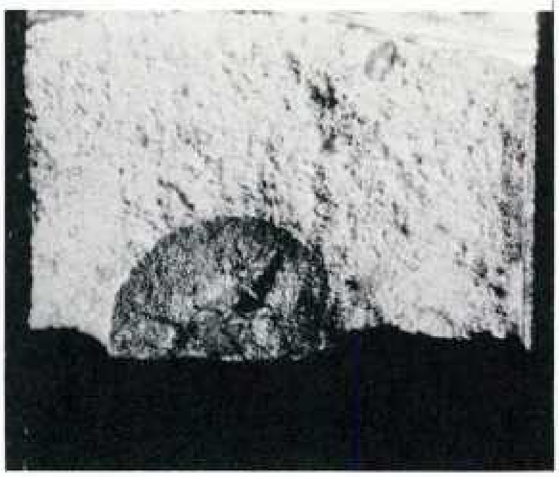
Figure 13. Crack growth from a machining damage origin, Nickel base superalloy

250µm

Figure 14. Crack growth from an impact damage mark, Ti6Al4V

2.5mm

Figure 15. Surface scratch in a Ti6Al4V disc



1mm

Figure 16. Crack growth from a near-surface inclusion, Nickel base superalloy

250µm

Figure 17. Subsurface defect, Nickel base superalloy

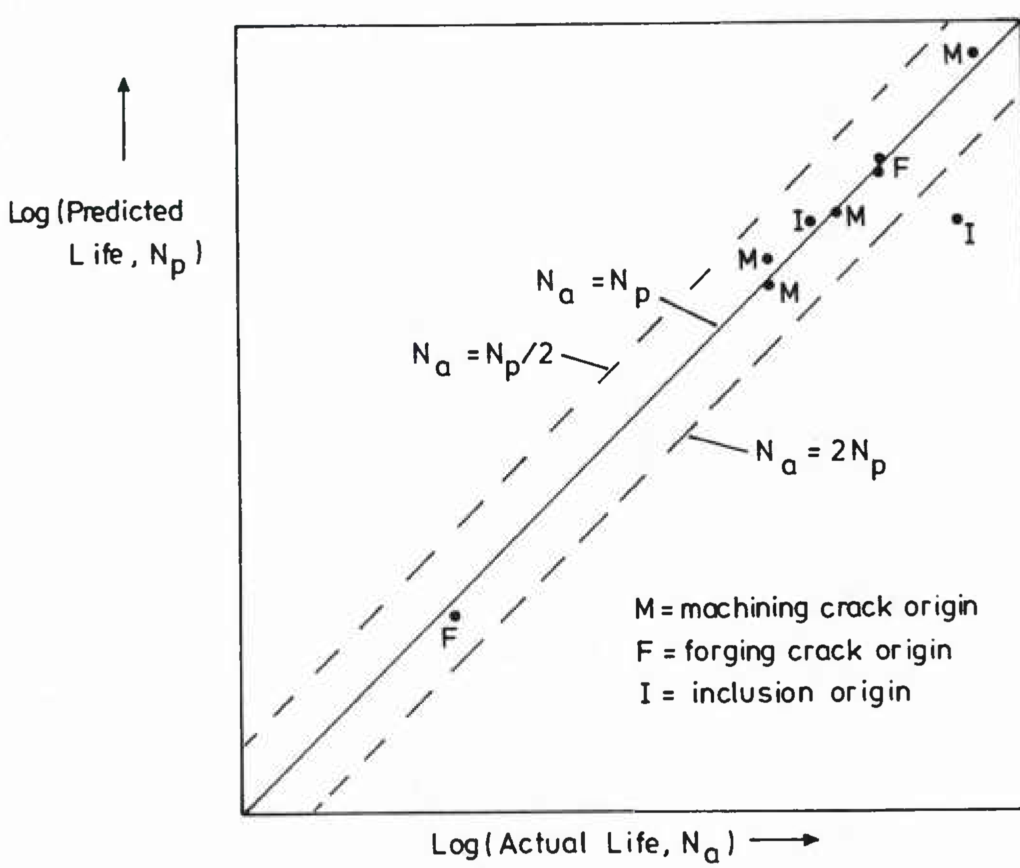


Figure 18. Comparison of actual and predicted lives for failures from defects, Nickel base superalloy

LES DEFAUTS DANS LES ROUES DE TURBINE

COULEES MONOBLOC

par
D. FOURNIER
TURBOMECA
BORDES 64320 BIZANOS
FRANCE

1. INTRODUCTION

La conception d'une pièce aéronautique est généralement le résultat d'une optimisation suivant 3 paramètres :

- fiabilité
- performance
- coût

Dans les turbines à gaz, cette optimisation, amène parfois à des choix différents dans le cas des moteurs de faible puissance par rapport à ceux de forte puissance. Fig. 1.

Nous présentons ici une technologie spécifique aux petites turbines à gaz : les roues de turbines coulées monobloc. On peut estimer à ce jour que plus de 300 000 roues coulées monobloc ont été fabriquées. L'expérience en vol est de l'ordre de 100 millions d'heures.

On rencontre ces pièces sur les turbomoteurs d'hélicoptère des turbopropulseurs et des réacteurs de faible et moyenne puissance. Les plus répandus sont : Allison 250 - Garrett TPE 331 - Lycoming LTS 101 - Turboméca ARRIEL 1A.

Les avantages et les inconvénients de cette technologie sont résumés dans le tableau ci-dessous :

AVANTAGES	INCONVENIENTS
- Faible coût - Simplicité - Possibilités géométriques (plus de pales sur un même diamètre) - Réduction des fuites	- Défauts nombreux (procédé) - Techniques de contrôle peu performantes

2. EXIGENCES TECHNIQUES

Une roue est représentée sur la figure 2. C'est une pièce de haute intégrité, la rupture ne peut être envisagée en service. Les conditions à remplir, les plus critiques pour l'utilisation sont :

Pour les pales : tenue en fluage de 900 à 1000 °C
résistance à la corrosion

Pour le moyeu : résistance à la fatigue oligocyclique à 500°C.

3. CHOIX DE L'ALLIAGE

Les alliages les mieux adaptés sont les superalliages base nickel en particulier : IN 713 LC, IN 792, MAR M 247. Ces alliages sont également employés pour la réalisation d'aubes de turbine. A TURBOMECA nous avons sélectionné le MAR M 004 qui est une modification de l'INCO 713 LC par addition de 1,3 % d'hafnium dans le but d'améliorer la ductilité. Le point critique pour l'alliage est d'obtenir des valeurs suffisantes de ductilité et de ténacité dans la zone la plus massive : le moyeu.

<u>Structure métallurgique</u> : le procédé de fonderie cire perdue conventionnel conduit à une solidification colonnaire grossière fig. 3.

<u>Microstructure</u> : Cet alliage est durci par la phase γ', la fraction volumique atteint 60 %. On a également une quantité importante d'eutectiques γ/γ' en nodules.

Propriétés mécaniques : A la température ambiante les propriétés mécaniques sont relativement faibles, mais elles restent au même niveau jusqu'à 650°C. Dans le domaine de température de 500 à 650°C c'est un alliage performant.

Charge de rupture $\geqslant$ 780 MPa
Limite élastique $\geqslant$ 700 MPa
Tenacité K1C $\geqslant$ 70 MPa $\sqrt{m}$

Les propriétés de fatigue oligocyclique sont montrées sur la figure 4 en comparaison à celles de l'INCO 718 forgé qui est un des alliages actuels les plus performants.

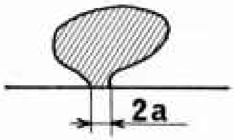
Le mécanisme d'amorçage des fissures de fatigue est visible sur la figure 5. L'amorce est intergranulaire provoquée par les incompatibilités géométriques de déformation d'un grain à l'autre. La propagation est transgranulaire suivant les plans de glissement.

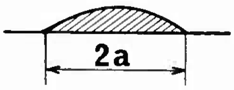
4. CONTROLES

Les roues monobloc sont des pièces critiques, il est donc nécessaire d'avoir une procédure très performante. Ces contrôles sont destructifs et non destructifs. La nature et la fréquence sont indiqués sur le tableau de la figure 6.

5. DEFAUTS

Aucun défaut supérieur à 0,2 mm n'a été détecté sur l'une des 250 coupes diamétrales de roues examinées. Tous les défauts importants (> 0,3 mm) ont été détectés par courants de Foucault dans l'alésage des pièces. Sur le tableau de la figure 7 nous avons indiqué la nature et la fréquence des défauts importants. Figures 8, 9, 10, 11 on peut observer quelques uns des plus typiques après des essais au banc partiel ou sur moteur. Les clichés ont été pris au microscope électronique à balayage après ouverture des fissures. Dans tous les cas l'essai a été arrêté avant l'éclatement. Nous avons procédé de la manière suivante : quand on a une indication de défaut en courants de Foucault on prend une réplique. Si le défaut est confirmé la roue est mise en essai au banc ou sur moteur. Nous avons constaté qu'il était quasi impossible de prévoir la propagation de ces défauts ; la dimension apparente ne permet pas d'estimer la profondeur et la forme de la fissure comme indiqué sur le schéma :





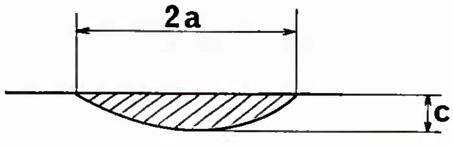
La taille apparente est : 2a.

6. PROPAGATION D'UNE FISSURE SEMI ELLIPTIQUE

Dans le cas d'une application hélicoptère on peut prendre comme objectif de durée de vie : 3000 cycles.

Nous avons calculé en collaboration avec l'Ecole des Mines de Paris, le comportement d'une fissure semi elliptique dans une roue monobloc.

Schéma de la fissure :



Nous avons utilisé les courbes de vitesse de fissuration de la figure 12 en prenant à chaque fois une courbe de vitesse minimale et une de vitesse maximale.

6.1. Défaut initial provoquant la rupture en 3000 cycles

Nous avons d'abord déterminé la taille du défaut initial provoquant une ruture en 3000 cycles. Les résultats sont reportés figure 13 sous forme graphique. En prenant la courbe de fissuration minimale la profondeur du défaut initial est de 0,91 mm avec la courbe de fissuration maximale la profondeur du défaut initial est de 0,31 mm.

6.2. Influence de la taille du défaut initial

Sur la figure 14 on peut voir l'évolution du nombre de cycles de fissuration en fonction de la taille initiale du défaut. On constate un facteur 3 sur la durée de vie suivant que l'on considère un régime de fissuration lent ou rapide. Ceci souligne l'importance de la détermination de la courbe de vitesse de fissuration. On peut remarquer que pour cette géométrie de fissure les courants de Foucault permettent de garantir une durée de vie en fissuration de 4500 cycles (Défaut initial 2a = 0,3 mm).

6.3. Influence de la forme de la semi-ellipse

Sur la figure 15 nous avons tracé la taille de la fissure à l'éclatement (a) en fonction de l'élancement de la demi ellipse ($\frac{a}{c}$). On constate que le facteur le plus important est la profondeur. La partie débouchante de la fissure (2 a), c'est-à-dire la partie détectable avec les moyens actuels, permet uniquement d'affirmer la présence d'une fissure sans pouvoir quantifier sa nocivité. Ceci implique actuellement de rebuter toute roue sur laquelle on a détecté une fissure.

7. CONCLUSIONS

- Il existe des défauts de taille importante dans les roues monobloc de fonderie. Leur nombre est très faible, (5 défauts pour 1000 pièces).
- Leurs natures et leurs positions sont connues.
- Le matériau des roues monobloc, l'alliage MAR M 004 est tolérant au dommage.
- Une procédure de contrôle spéciale permet de garantir l'utilisation de ces pièces.

Des améliorations sont cependant souhaitables au niveau :

- Du procédé pour générer moins de défauts de grande taille
- Des techniques de contrôles non destructifs pour déterminer la profondeur des fissures.

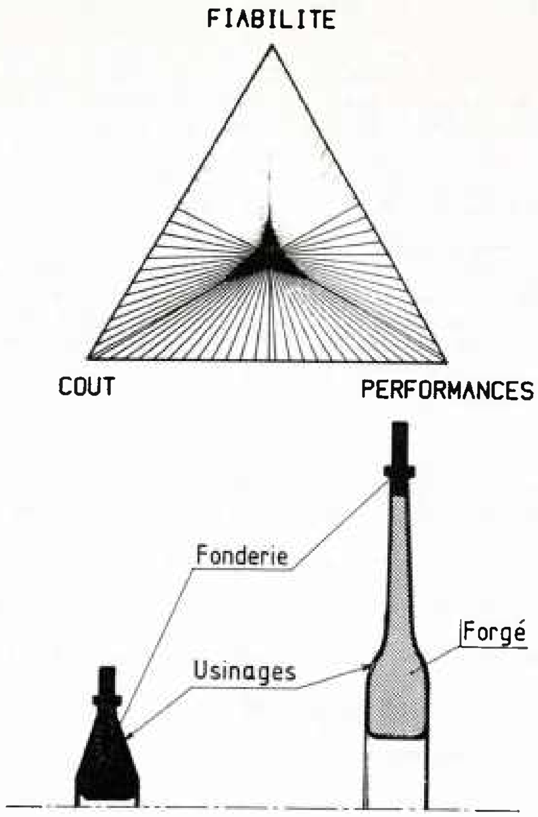


Figure 1

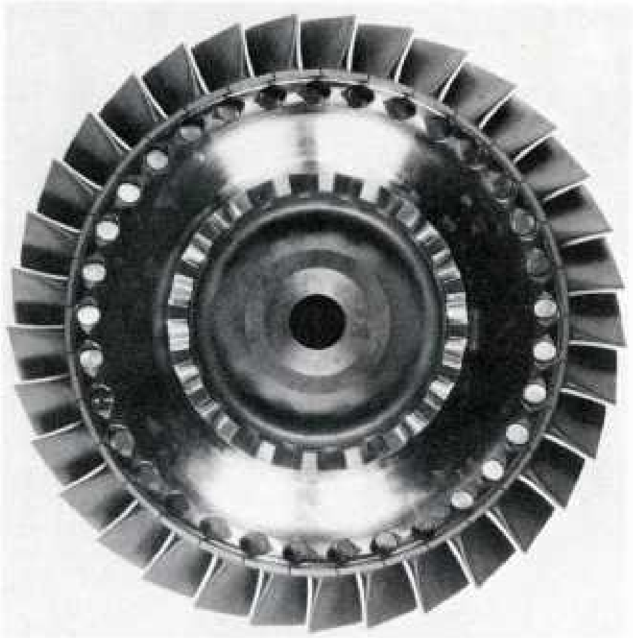


Figure 2

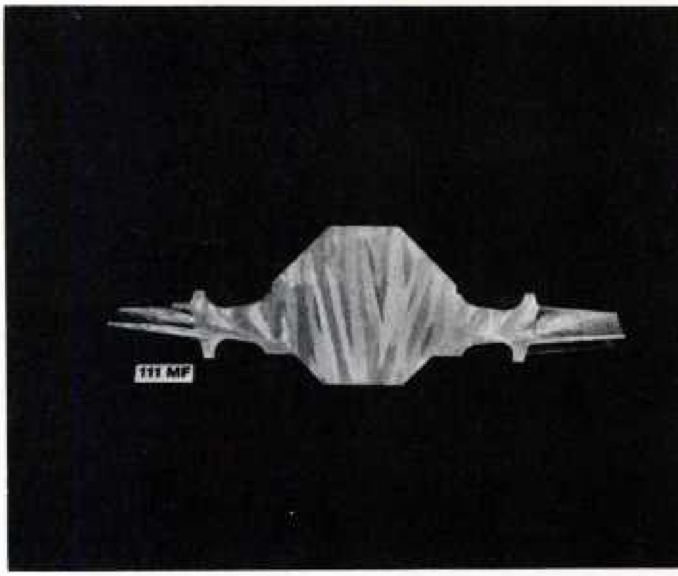
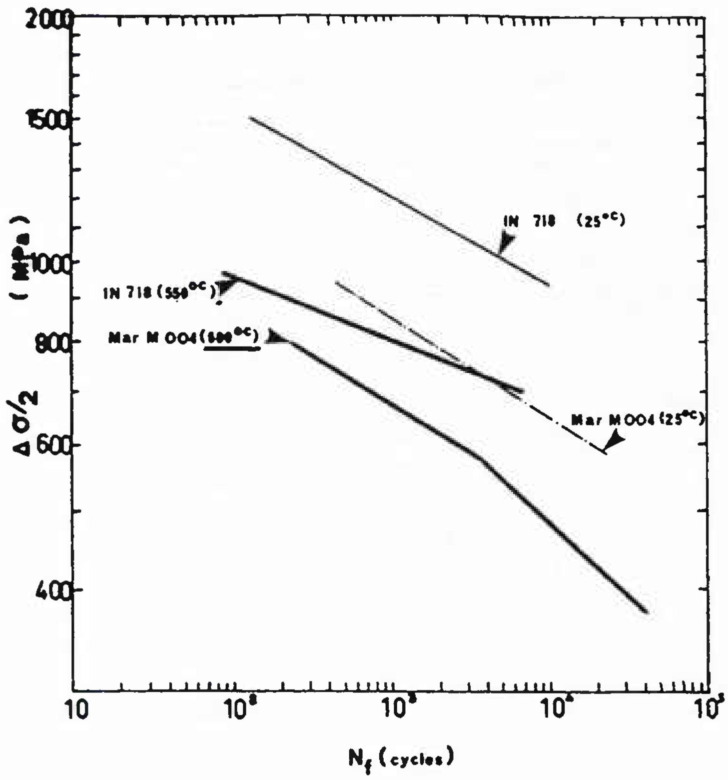


Figure 3



VARIATION DE LA CONTRAINTE A DEMI-DUREE DE VIE $\Delta\sigma/2$ EN FONCTION
DU NOMBRE DE CYCLES A RUPTURE Nf POUR DIFFÉRENTS MATÉRIAUX
a) MarM004 20°C ; b) MarM004 800° C ;
c) INCO 718 20° C ; d) INCO 718 550° C

Figure 4

Figure 5

CONTROLES SUR LES ROUES MONOBLOC

	CAROTTES	ETAT BRUT DE COULEE	ETAT USINE
Micro/Macro	100%	100%	
Propriétés Mécaniques	10% Traction 10% Fluage	1 par lot de coulée	
Rayons X Conventionnel Rayons X Microfocus		Sur chaque pale	
Contrôle Ultra-sons		0.5 mm à 8mm de profondeur ou 1.0 mm à 32mm de profondeur Variation de structure (Basltique/Equiaxe)	
Contrôle par ressuage	100%	100%	100%
Courants de Foucault			Moyeu 100% Surface crique > 0.1mm Film d'oxyde > 0.3 mm

Figure 6

DEFAUTS DETECTES

5000 roues inspectées (Mar M004)

Définition : Discontinuité nocive pour l'application

* Pour les petites roues, ils sont génrslement localisés dans le moyeu
* Ces défeuts sont dûs au procédé de coulée

	Pourcentage de roues (%)	Taille maximale (mm)
Film d'oxyde (Hf , Nb)	0.5	3
Porosités	0.1	2
Ségrégation Phases aciculaires Hf , S , Zr	0.05	0.2

Figure 7

Figure 8

Figure 9

Figure 10

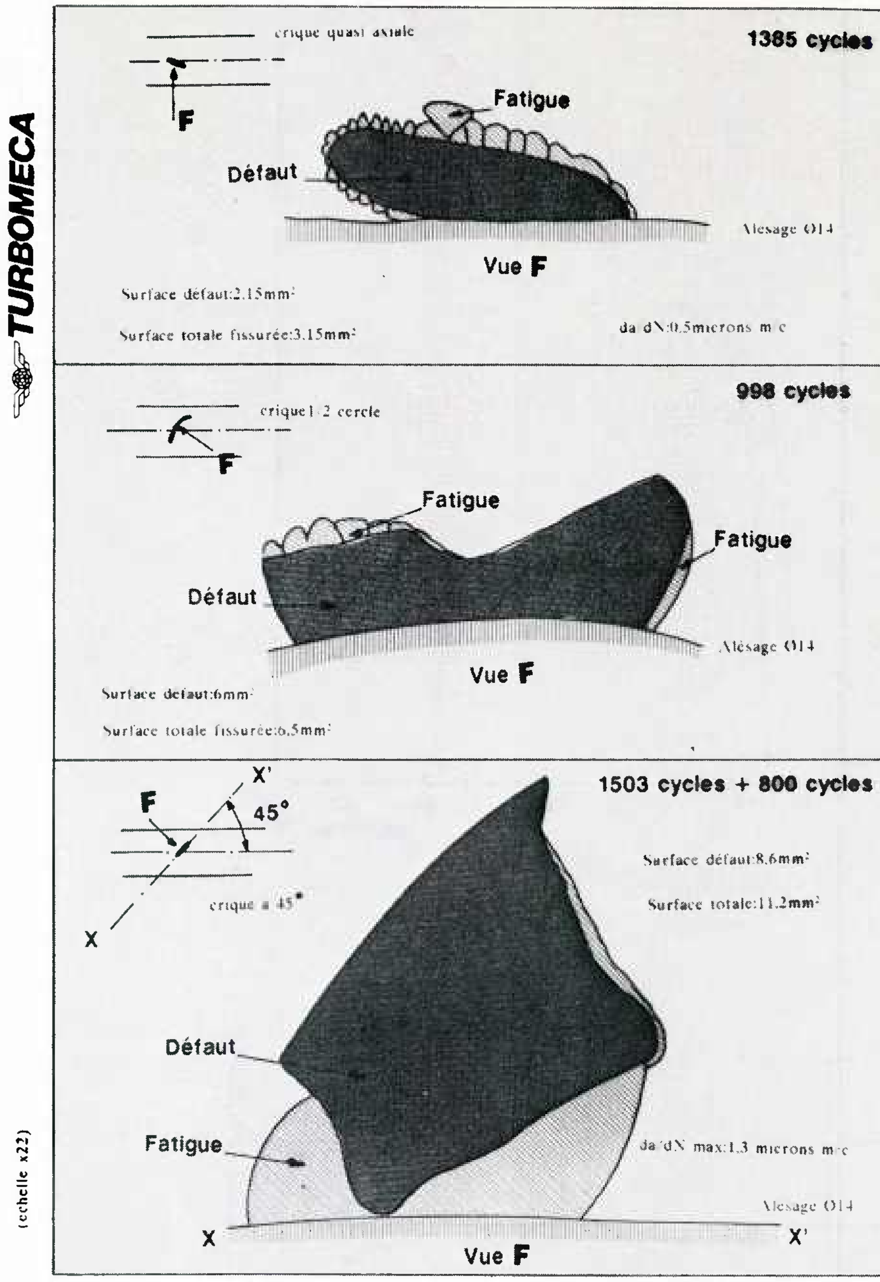

TURBOMECA
crique quasi axiale
1385 cycles
F
Fatigue
Défaut
Alésage O14
Vue F
Surface défaut:2.15mm²
Surface totale fissurée:3.15mm²
da/dN:0.5microns m/c
998 cycles
crique 1/2 cercle
F
Fatigue
Fatigue
Défaut
Alésage O14
Vue F
Surface defaut:6mm²
Surface totale fissurée:6.5mm²
X'
F
45°
1503 cycles + 800 cycles
Surface défaut:8.6mm²
Surface totale:11.2mm²
crique à 45°
X
Défaut
Fatigue
da/dN max:1.3 microns m/c
Alesage O14
X
X'
Vue F
(echelle x22)

Figure 11

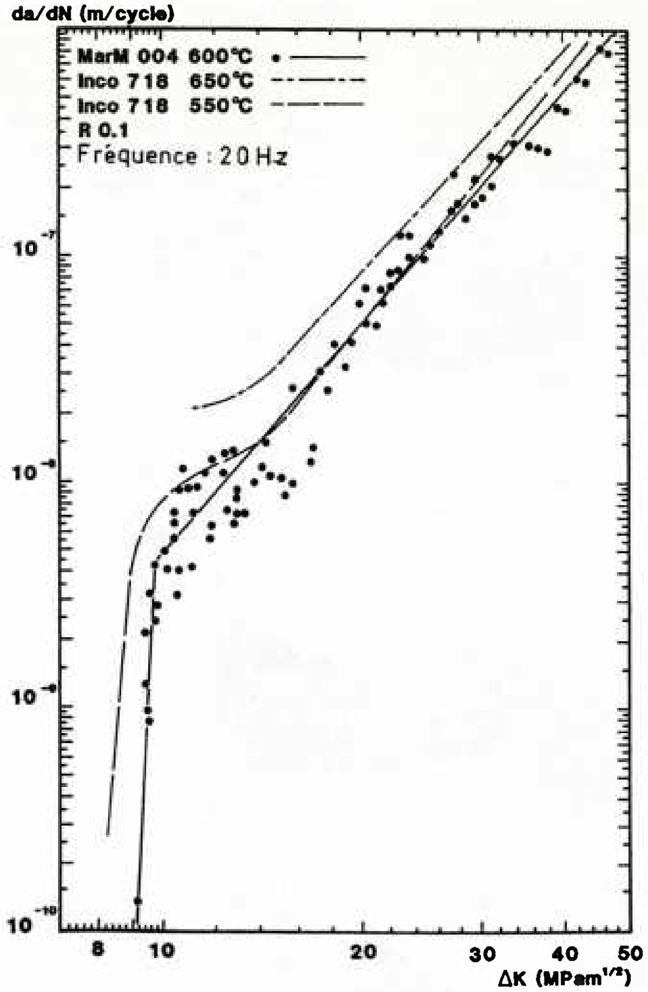
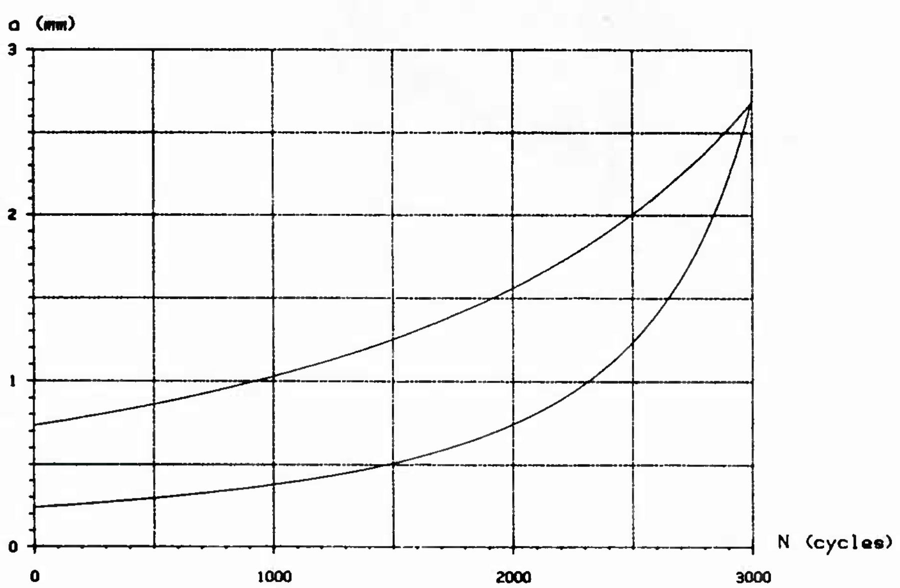
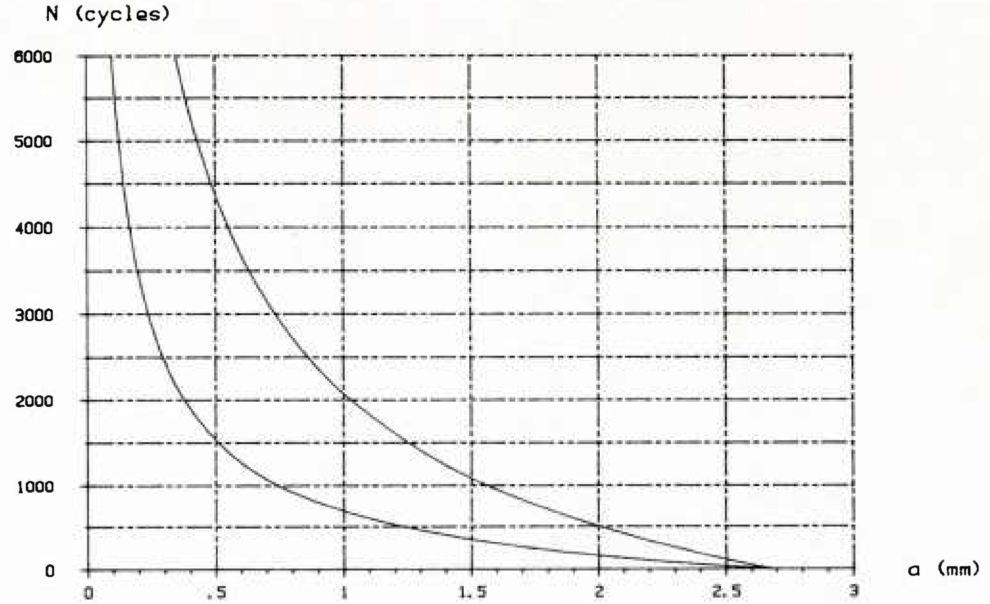


Figure 12



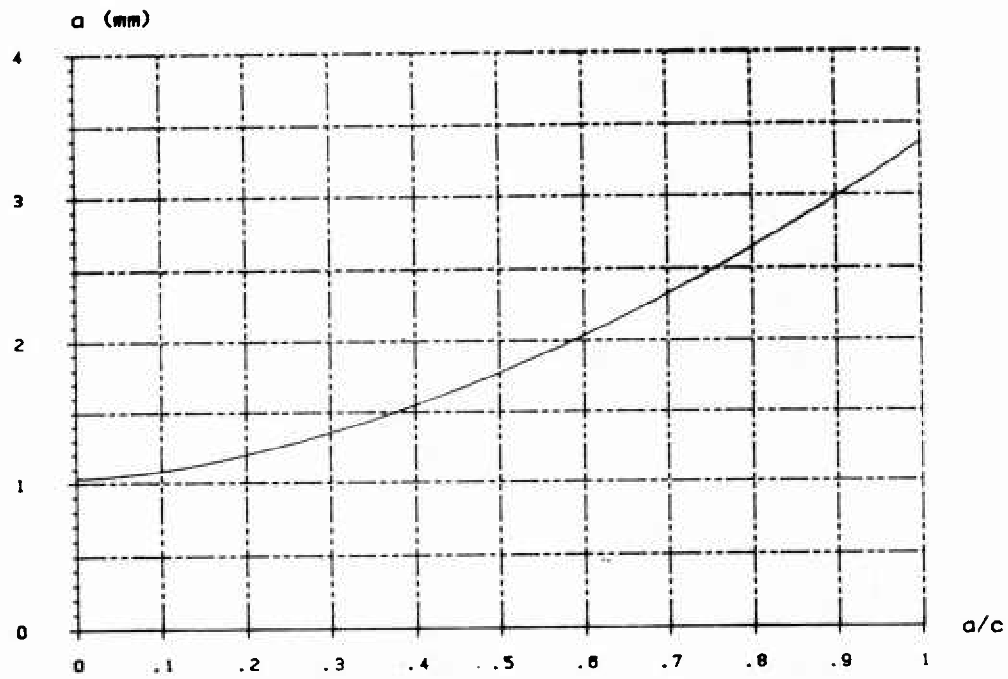
EVOLUTION DE LA TAILLE DE LA FISSURE CRITIQUE
Fissure semi-elliptique (a/c=0.82)

Figure 13



NOMBRE DE CYCLES A RUPTURE EN FONCTION DE a
Fissure semi elliptique a/c=.82

Figure 14



PROFONDEUR DE LA FISSURE A L'ECLATEMENT
EN FONCTION DU RAPPORT a/c (ROUE DE TURBINE 2EME ETAGE)

Figure 15

CRACK PROPAGATION FROM ARTIFICIAL DEFECTS IN TYPICAL ENGINE DISK MATERIALS; INTERPRETATION OF EXPERIMENTAL RESULTS BY MEANS OF 3D J INTEGRAL.

Aldo Frediani
Dipartimento Aerospaziale
Università "La Sapienza"
Via Eudossiana 16.00185 Roma
ITALY

Roberto Galatolo
Dipartimento di Ingegneria Aerospaziale
Università di Pisa
Via Diotisalvi 2, 56100 Pisa
ITALY

SUMMARY

This paper shows the results of crack propagation tests carried out on corner cracked specimens built in Powder Metals, AP1 and AF115, at high temperatures and different values of the stress ratio R. Analysis of the final crack shapes shows that discrepancies exist between the crack shapes expected on the basis of the behaviour of the Stress Intensity Factor and the actual crack shapes; similar conclusions are obtained in other significant examples. The threedimensional J-Integral (or GJ) is, then, introduced; the GJ values are computed along the crack fronts of circular shaped cracks and the actual cracks. It transpires that the values of GJ are virtually constant along the fronts of the actual cracks. Finally, future possibilities of application are outlined.

1. INTRODUCTION.

The introduction into service of new materials is necessary in order to improve the performance of the gas turbine disks for aircraft applications.Turbine disks are critical components of engines because of the presence of high temperatures, high mean stress levels and variations of stresses, (especially in the central bore). At present, considerable interest is being shown by Industry in nichel-based superalloys, obtained by means of the Hot Isostatic Pressing (HIP) technique, for their optimum mechanical characteristics up to high temperatures. Together with forging, the HIP methodology allows us to obtain a component which is very near the final shape; besides, different grain sizes can be obtained by suitably varying the parameters of the forming process (temperatures, pressures, cooling rates, temperature sequences, etc.). The main disadvantage of the HIP technique is the presence of small-size ceramic particles (a typical size is 100 µm), from which fatigue cracks can propagate up to failure. Total fatigue life (or "Actual life") is the sum of the following stages: (a) development of an initial crack from a defect or inclusion and (b) further regular crack growth to failure.
It is not simple to correlate the duration of stage (a) to the size and type of defect; besides, Fracture Mechanics do not allow us to interprete results in the short crack range. Therefore, because of the difficulties for certyfing the life spent in stage (a) of a Damage Tolerance component, aircraft engine industry need optimized materials in order to ensure the maximum possible fatigue life in stage (b).
This paper shows da/dn-ΔK curves relevant to two fine-grain Powder Metal alloys, AP1 and AF115, at different values of $R=s_{min}/s_{max}$ and temperatures of 450 °C and 600 °C. All the tests performed were carried out at the Fatigue Laboratory of the Department of Aerospace Engineering in Pisa.
The AGARD research on the TX114 programme suggested that a suitable specimen geometry is the Corner Cracked geometry, because:
- the threedimensional crack front shapes are near to the actual crack shapes growing in the bore of the disks;
- as shown in /1/, the crack propagation curves obtained from Corner Cracked specimens

describe the behaviour of actual cracks in the bores better than in all the other kinds of specimens;
- finally, the C.C. specimens allow us to start the crack-growth tests from very small crack lengths.

2. MATERIALS AND SPECIMENS

The materials examined were the two aforementioned Powder Metal superalloys obtained by Hot Isostatic Pressing: AP1 and AF115. The composition and the main mechanical properties of such materials are given in Table 1.
The heat treatment is summarized below:

AP1:- 2h (hours) at 1160°C/1050 bar in Argon, obtaining a round bar of 250 mm diameter
- forging at 1100° C
- heat treatments:
  - 4 h at 1100°C - air cooled
  - 24 h at 650°C - " "
  - 8 h at 760°C - " "

AF115:- 2 h at 1180°C/1050 bar in Argon obtaining a round bar of 150 mm diameter.
- heat treatments:
  - 4 h at 1150°C - air cooled
  - 16 h at 760°C - " "

The average grain-sizes were 16 μm and 11 μm for AP1 and AF115, respectively. Both alloys contained ceramic inclusions of about 100 μm. The Corner Crack specimens used were cut from the central bore of engine disks in the tangent direction, for AP1, and from non-forged bars for AF115; the dimensions of the specimens built in AP1 and in AF115 are, respectively, 8x8 mm and 10x10 mm (fig.1). A micronotch was made by means of electro-discharge machining on a corner of each specimen to simulate the presence of small-size initial cracks; the sizes of the micronotches are given in Table 2. The orientation of the specimen crack planes in relation to the forgings was specified such that they would have the same crack orientations as those encountered in service. The positions of the micronotches on the specimens are not related to the positions of the cracks in the disks; however, this is not a limitation, as the experience gained in the TX114 programme suggested.

3. EXPERIMENTAL PROCEDURE

The tests were carried out on a SERVOTEST electro-hydraulic 250 KN fatigue testing machine. A MAYES resistance split furnace was used to perform high-temperature tests. The Direct Current Potential Drop technique was used to measure the crack length automatically by means of a personal computer as shown in fig.2. The arrangement of the probes on the specimens is shown in fig.3 and the arrangement of the specimens is given in fig.4, together with the positioning of a thermocouple for closed-loop control of the temperature in the furnace.The loading waveform was trapeziodal with a frequency of 0.25 Hz.

4. EXPERIMENTAL RESULTS

Costant amplitude fatigue crack-growth tests were carried out at 450°C and 600°C on AP1 and AF115, respectively. The crack-length versus number of cycle curves were obtained from the P.D. measurements by means of the calibration curves for C.C. specimens reported in /1/, with essentially the same procedure as that used at room temperature in the AGARD TX114 programme.
Because the Stress Intensity Factor (SIF) is not constant along the crack front (fig. 5), an average value between the SIF on the surface and the SIF on the diagonal of the specimen was taken in order to compute the SIF variations, ΔK, according to /3/. The relationships

between da/dn and ΔK (ΔK=Kmax-Kmin), relevant to the AP1 alloy, are shown in figs 6 to 9 for R=0.1, R=0.25, R=0.5 and R=0.7, respectively, and all the average curves are collected together in fig 10. An analogous crack-growth curve for AF115 Alloy is shown in fig.11 (R=0.1); fig.12 shows a comparison between AP1 and AF115 at the same value of R=0.1, but at different temperatures.

5. REMARKS ON THE CONSISTENCY BETWEEN THE TEST RESULTS AND THEORETICAL MODELS

The values of K, used in da/dn-ΔK curves are relevant to the following conditions:

a) the Corner Crack is circular-shaped and, consequently, it can be fully characterized by the single parameter "a" (crack length);

b) as is well known, the evaluation of K is possible only in plane stress or, alternatively, plane strain state.

But in all the tests performed, the crack-front was never circular, as can be seen in fig.13; the situation was the same in all the other specimens tested. We may note, in particular, that:

- crack length in the diagonal direction is larger than that on the free surfaces
- the cracked surface is not completely flat; in the vicinity of the free surfaces, the crack is slanting.

Indeed, similar kinds of behaviour are to be found, for example, in a Compact Tension specimen, where real crack fronts are not straigth (fig.15) and crack planes are slanting in the vicinity of the free surface. It is easy to see that these experimental results can not be explained by resorting to the Stress Intensity Factor K.

Let us first assume that the crack growth rate is a growing function of K at any point on a crack front; from fig.13 we can deduce that the circular shaped crack front is not stable and we obtain a crack shape where the crack along the 45 direction is smaller than that in the zero direction (contrary to the experimental results).

These discrepances exist in other situations. Let us consider, for example, a Compact Tension specimen (in the rest of the paper we will consider front shapes relevant to room temperature tests in order to avoid any possible dependence on oxidation phenomena); the only situation where the plane strain state is present is that shown in fig.15, where K is constant along the thickness. Usually, computation of the S.I.F in a real C.T. specimen (with finite width) is carried out assuming there is strain state and, consequently, we would expect to find a straight crack front when the specimen undergoes fatigue loading.

On the contrary, the crack front of real cracks is the same as in fig.14; therefore, it seems reasonable to deduce that the plane strain condition is not satisfied and, moreover, that the presence of the plane strain condition may play an important role in the crack growth phenomenon. In conclusion, it seems that the use of the S.I.F. as a mean of measuring of the Energy Release Rate may give rise to predictions, which do not agree with significant experimental results.

6. GENERAL MEASUREMENT OF ENERGY RELEASE RATE

In order to try to overcome the aforementioned discrepancies, let us introduce the so-called J Integral in threedimensional conditions, or, briefly, Generalized J Integral (GJ in the rest of this paper).As is known from the literature, GJ is a vector whose components are:

$$GJ_1 = \int_\Gamma (W\, n_1 - T_i\, u_{i,1})\, ds - \int_A (\sigma_{i3}\, u_{i,1})_{,3}\, dA$$

$$GJ_2 = \int_\Gamma (W\, n_2 - T_i\, u_{i,2})\, ds - \int_A (\sigma_{i3}\, u_{i,2})_{,3}\, dA$$

$$GJ_3 = \int_\Gamma (W\, n_3 - T_i\, u_{i,3})\, ds - \int_A (\sigma_{i3}\, u_{i,3})_{,3}\, dA$$

where:

W is the Elastic Energy Density

T_i is the i-th component of the Stress Vector
u is the displacement vector
σ_{ik} is the stress tensor.

When we assume a global reference system as in fig.16, the GJ_1 and GJ_3 components lie in the crack plane; GJ_2 is normal to that plane.
The integration path Γ, that includes the surface with area A and surrounds the crack tip at a certain value "s" of the curvilinear coordinate defined on the crack front, can be chosen arbitrarily; indeed it is known that GJ_1 and GJ_3 are path-independent integrals; the same has not been proved in the case of the GJ_2 integral.
Following a well-known result in Continuum Mechanics, we can relate GJ to the Total Potential Energy Release Rate as follows:

$$GJ_1 = - dU^{TOT}/da_1$$

$$GJ3 = - dU^{TOT}/da_3$$

where:

U^{TOT} is the Potential Energy of the system
da_1, da_3 are the variations in the crack-length in the x_1 and x_3 directions, respectively.

It is important to note that no restriction exists on the state of stress or strain and, therefore, the quantity GJ is a more general parameter than the S.I.F. in Fracture Mechanics; moreover, GJ is well-defined in finite elasticity and general elastic-plastic conditions.
A computer program, called GJINTE, has been developed at the Department of Aerospace Engineering in Pisa for evaluating the GJ Integral along any point of a crack front; details of the program are given in /4/. Computations on Corner Cracked (C.C.) and Compact Tension (C.T.) specimens have been carried out in order to compare the results obtained with those in the Literature; a number of results obtained with the C.C. specimens are shown later in the paper; the first example is a C.C. specimen with a circular crack-front. The distribution of the GJ Integral normal to the crack-front, at any point , shows that the maximum values of GJ are set near the free surface (fig.17); if the crack moves without any change in the shape of the front in the diagonal direction, the distribution of GJ is different and the maximum values of GJ are in the symmetry plane of the section (fig.18).
The latter distribution qualitatively agrees with the experimental results, even though it can be criticized in the vicinity of the free surface where the crack grows, more realistically, in a different direction.
So, the following problem arises: "How does a given stable crack change under repeated loads?"
If we suppose that the growing process is slow, we may also suppose that the crack is modified according to a set of equilibrium conditions of the crack front. So, it may be formulated certain interesting stability conditions for the crack front, at least as a sufficient condition.
Let us try to formulate the following principle:
"If the shape of a crack is stable, the GJ Integral along the crack front is costant."
Actual cracks are "stable" in the sense that they spring from a natural crack propagation process and are far different from initial, artificially-aged, defects.
Now, let us consider again the real crack front in fig.19. The values of GJ_1 and GJ_3 along the crack front are given in fig.20.
It is easy to see that the aforementioned principle is satisfied at all the points of the crack front with the exception of two sectors about 10 degrees near to the free surfaces.
But we have already observed that the real crack-front shapes obtained in the present research belong to a slanted plane in the zones near the free surfaces, while the Finite Element model relevant to fig.19 simulates a flat crack plane. So, in a real crack, component GJ_2 is no longer identical to zero and the contribution of GJ_2 has to be considered; we suspect that in slanted cracked surface it may be significant. Computation of component GJ_2 is now in progress; GJ in elastic-plastic conditions will be computed in near future. Anyway, a lot of research remains to be done in order to confirm the proposed criterium and formulate a necessary and sufficient condition for crack-shape stability.

7. CONCLUSIONS

The da/dn-ΔK curves have been assessed for the Powder Metals AP1 and AF115, to be used in engine disks.
Analysis of the cracked surfaces of the specimens tested indicate that possible discrepancies may exist between the experiments and the theoretical models obtained with the S.I.F.
Computations carried out on an actual fatigue crack in a C.C. specimen by means of a computer program based on the results of Finite Element analysis, seem to give preliminary indications that a crack tends to modify its shape in such a way that the GJ values are constant along the crack front.
This result, if confirmed, may be interesting even from the technical point of view; for example, materials might be classified from the point of view of their toughness by means of a single test, independently of the thickness of components.

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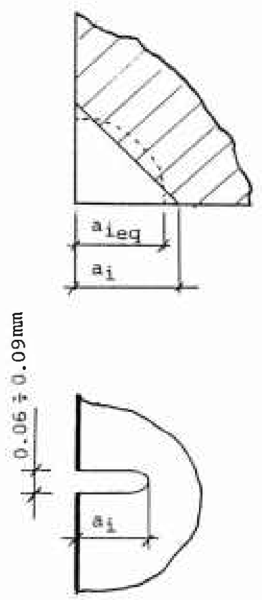
AP1 : CHEMICAL ANALYSIS													
C	Si	Cu	Fe	Mn	Mg	Cr	Ti	Al	Co	Mo	Ni	Zr	S
0.02	0.04	<.01	<.04	<.01	-	14.9	3.52	4.07	17	5.03	Bal	0.045	0.001
P	W	Ag ppm	B ppm	Bi ppm	Pb ppm	Sn ppm	Sb ppm	Zn ppm	Te ppm	Tl ppm	A ppm	N ppm	O ppm
<.005	<.05	<.1	213	0.1	0.3	5.1	<1	3	<1	<.2	1	0.002	0.006

AF115 : CHEMICAL ANALYSIS													
C	Si	Mn	P	S	Al	B	Co	Cr	Fe	Mo	Nb	Ti	W
0.02	0.1	0.11	0.012	0.012	3.7	0.02	14.85	10.03	0.05	2.75	1.9	4.15	5.88
Zr	O	Ni	Ag	Ar	Bi	N	Pb	Se	Sn	Te	Tl		
0.05	0.001	Bal	<5	<1.5	<.3	<50	<5	<3	<25	<1	<.5		

MECHANICAL PROPERTIES		
AP1 (T = 450 °C)	R = 1445 MPA ;	Sy = 1000 MPA
AF115 (T = 450 °C)	R = 1393 MPA ;	Sy = 1035 MPA

TAB.1 Chemical analysis and mechanical properties for AP1 and AF115 materials

Test	Spec.	Mat.	R	Smax (MPa)	Temp. (°C)	a_{ieq} (mm)	a_{feq} (mm)	N (cycles)
1	A2	AP1	0.1	922	450	0.08	3.65	21898
3	A3	AP1	0.1	922	450	0.22	2.67	15978
4	A6	AP1	0.1	1100	450	0.08	1.79	10420
5	A4	AP1	0.25	1100	450	0.08	1.92	18513
6	A12	AP1	0.5	1000	450	0.32	1.6	21469
7	A11	AP1	.0.5	1100	450	0.32	2.12	13952
8	A10	AP1	0.5	1000	450	0.32	2.4	20050
9	A8	AP1	0.7	1000	450	0.56	2.7	36187
10	A7	AP1	0.7	1100	450	0.56	2.6	27221
11	A9	AP1	0.7	1100	450	0.56	2.55	22430
12	A1	AP1	0.1	1000	450	0.08	1.9	13883
13	B4	AF115	0.1	1000	650	0.11	3.62	7891
14	B2	AF115	0.1	1000	650	0.11	2.83	8401
15	B1	AF115	0.1	1000	650	0.11	0.6	8721

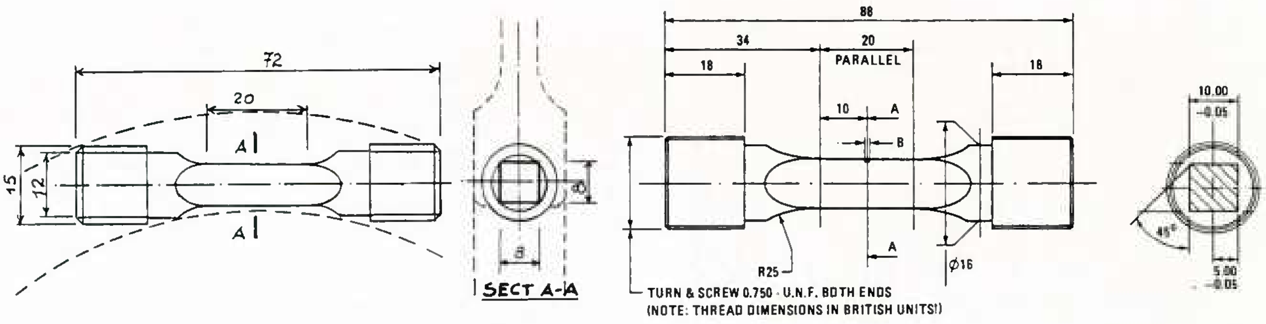


a_i = micronotch deep

a_{ieq} = equivalent initial crack length

a_{feq} = equivalent final crack length

TAB. 2 Test programme and initial artificial defect sizes



AP1 specimen

AF115 specimen

Fig.1 - Specimens geometry.

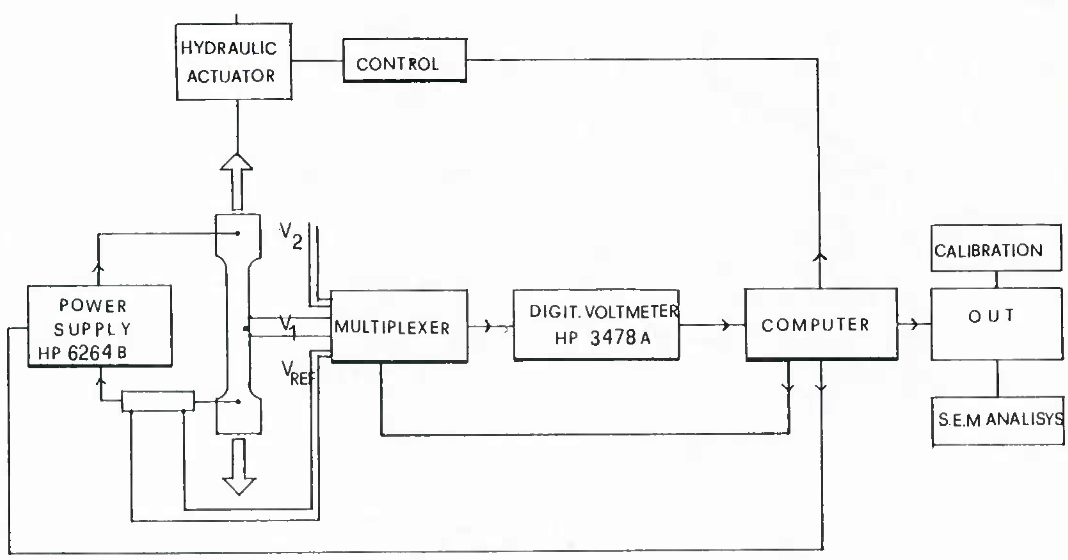
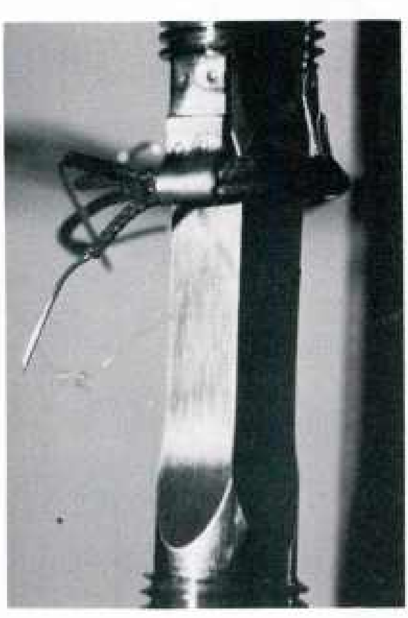


Fig.2 - Sketch of the test equipment.



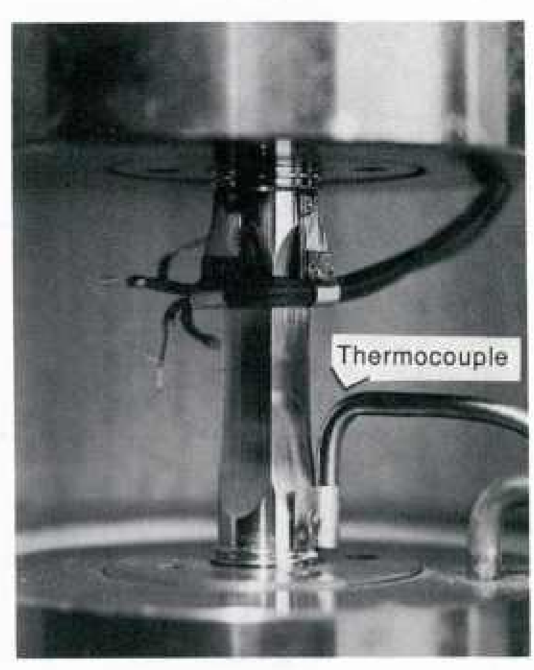


Fig.3 - Probes arrangement on the specimen.

Fig.4 - Specimen arrangement on the test rig.

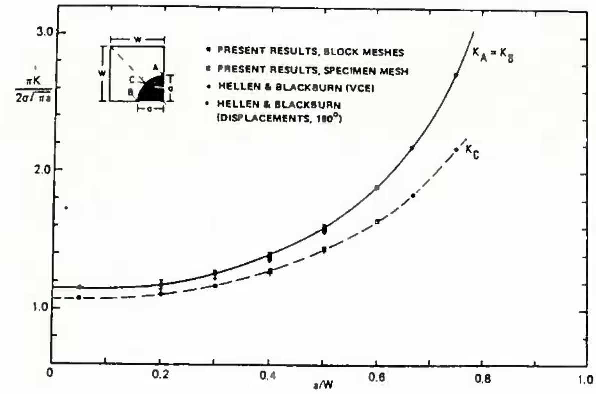


Fig.5 - K values on the diagonal direction and on the free surfaces.

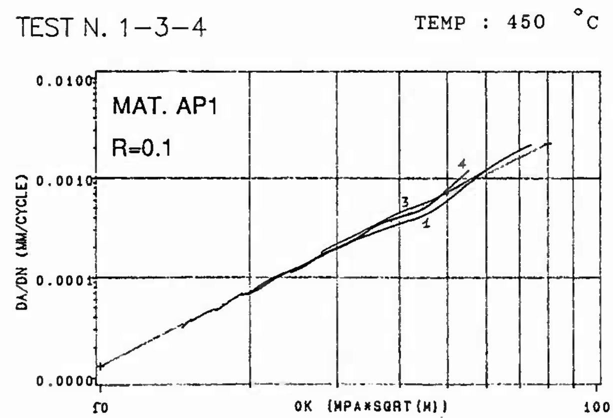


Fig.6 - Crack growth rate test results; material AP1, R = 0.1, T = 450 °C.

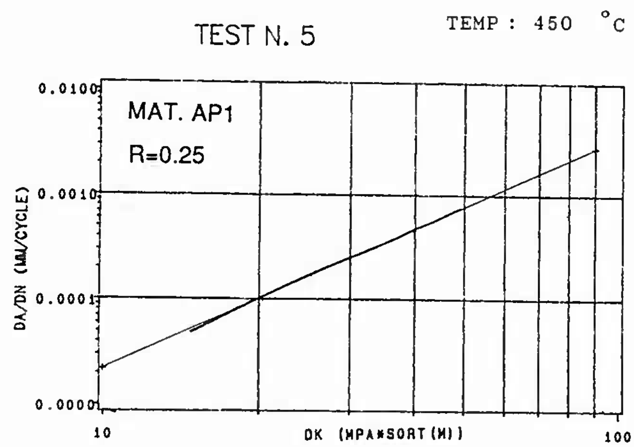


Fig.7 - Crack growth rate test results; material AP1, R = 0.25, T = 450 °C.

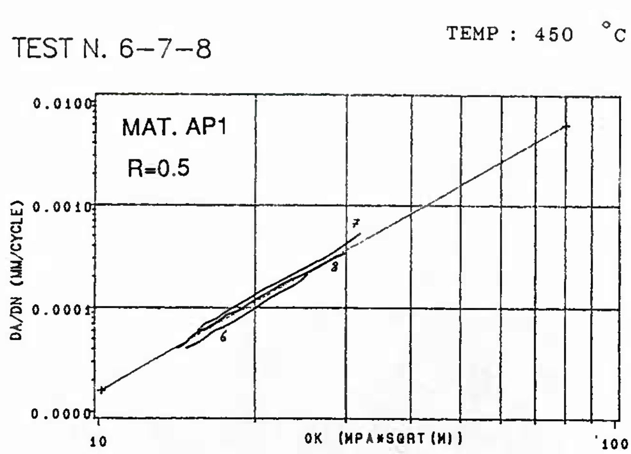


Fig.8 - Crack growth rate test results; material AP1, R = 0.5, T = 450 °C.

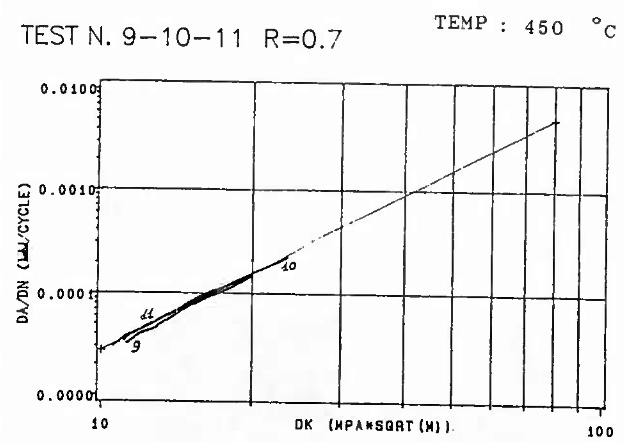


Fig.9 - Crack growth rate test results; material AP1, R = 0.7, T = 450 °C.

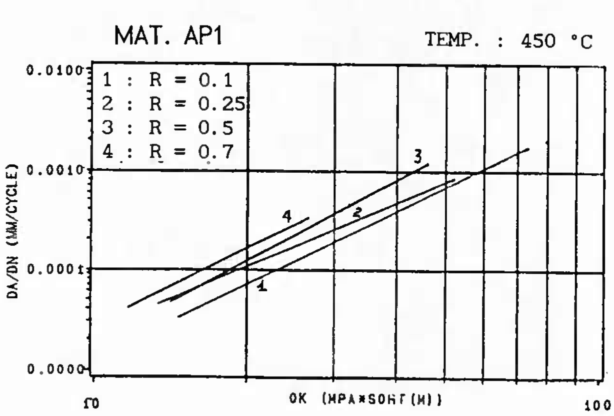


Fig.10 - Influence of R on the crack growth rate results for AP1 material.

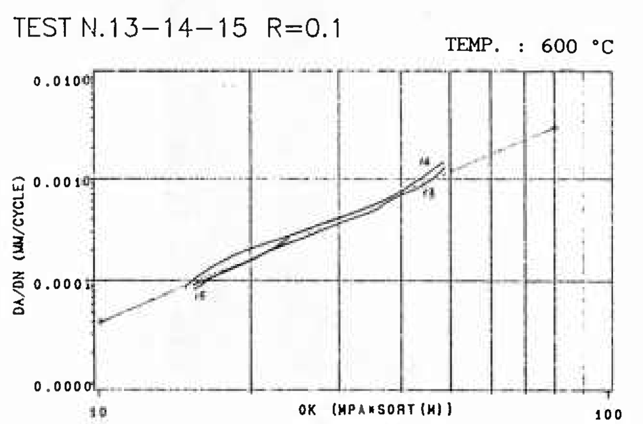


Fig.11 - Crack growth rate test results; material AF115, R = 0.1, T = 600 °C.

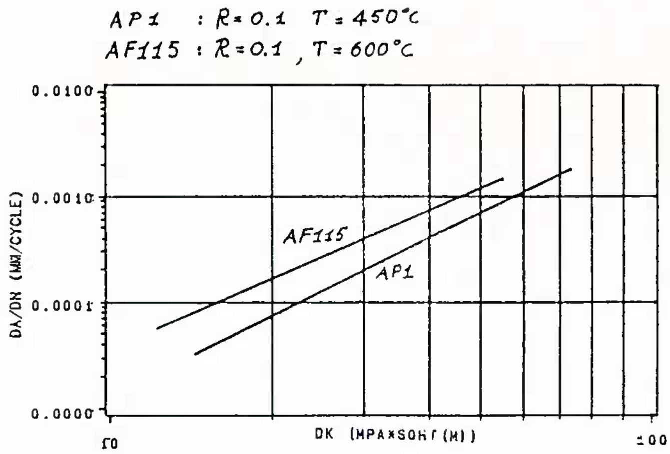


Fig.12 - Crack growth rate for AP1 and AF115 materials at different temperatures (R = 0.1).

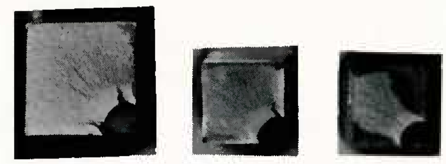


Fig.13 - Examples of crack front shapes in C.C. specimens.

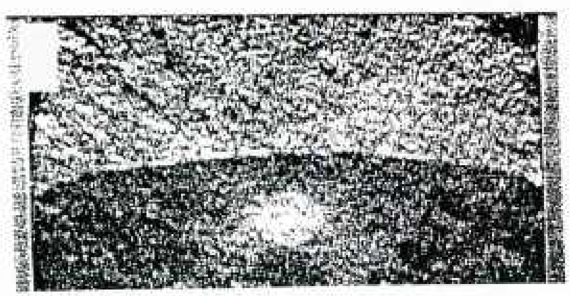


Fig.14 - Typical crack front shape in a C.T. specimen.

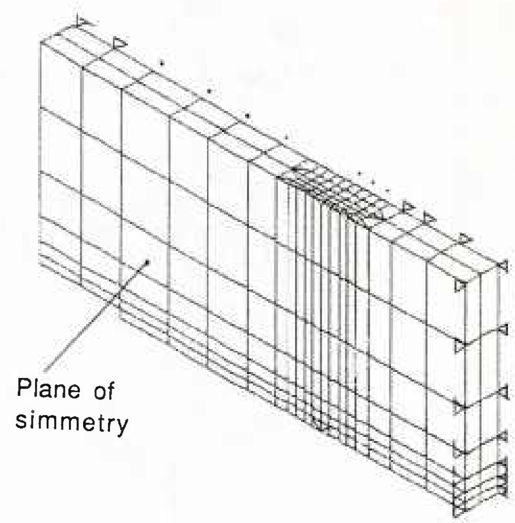


Fig.15 - F.E. mesh of a C.T. specimen with straight crack front (plane strain condition).

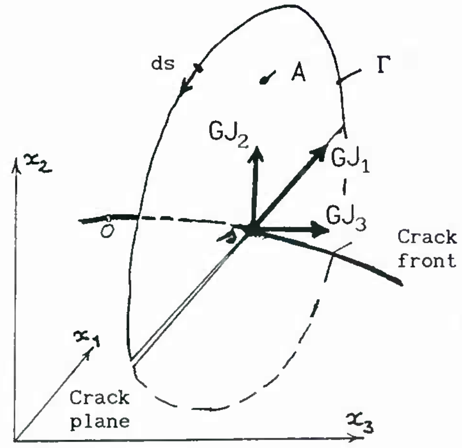


Fig.16 - Reference system and components of the GJ-integral.

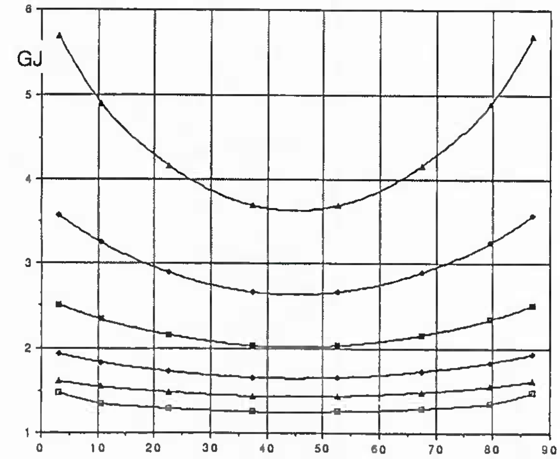


Fig.17 - GJ-integral behaviour along the crack front of a C.C. specimen for a virtual uniform crack extension according to the radial direction (circular shaped crack)

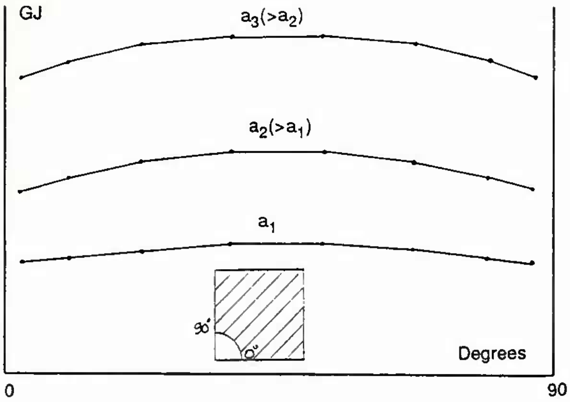


Fig. 18 - GJ-integral behaviour along the crack front of a C.C. specimen for a virtual uniform crack extension according to the 45 deg. direction (circular shaped crack)

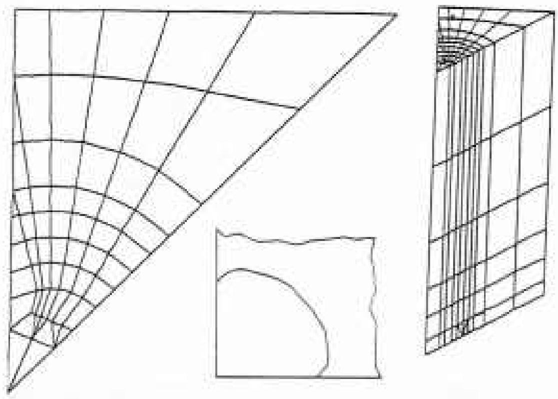


Fig. 19 - Actual crack front shape and F.E. mesh for a C.C specimen.

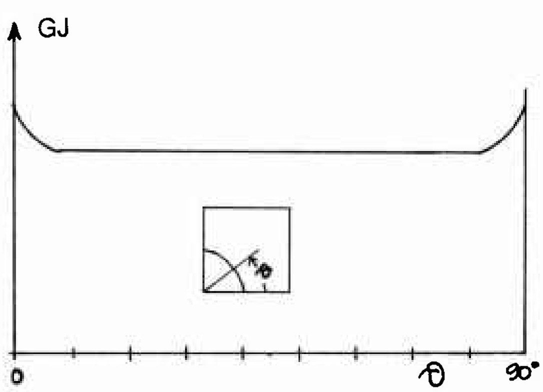


Fig.20 - GJ behaviour along the front of the crack in fig.19.

REVIEW OF MODELLING METHODS TO TAKE ACCOUNT OF MATERIAL STRUCTURE AND DEFECTS

R. N. Tadros and K. Rezai

Pratt & Whitney Canada Inc.,
1000 Marie Victorin Blvd.,
Longueuil, Quebec,
J4G 1A1,
CANADA.

ABSTRACT

Among the many requirements for the design of gas turbine engine components, understanding of the behaviour of these components and the alloys used to make them is fundamental in order to ensure structural integrity and safe operation of the engines. In this paper, the conventional Safe Life Approach is discussed together with an overview of crack growth modelling used at Pratt & Whitney Canada. The intention is not to compare the two approaches, but rather to highlight their applications and required improvement for the design and production of next generation gas turbine engines.

1. Introduction

The current method of commercial aircraft engine design is based on Safe Life Approach (SLA), with its design criterion based on the probability of a certain crack size formation (usually 1/32" or 0.794mm) during the lifetime of each engine component. The reliability and integrity of the engines designed in this way have been gained through proven analytical modelling and an experience, which has been accumulated over years from engine tests and component rig tests. This has been demonstrated for example by Pratt & Whitney Canada, who has accumulated over 150 million hours of field experience without a single tri-hub disc failure which leads to a probability of failure of less than 10^{-9} per flight hour for any of the high energy rotors. However, issues have been arisen during the past decade leading to search for another design alternative. These issues are related to the assumption that parts entering service are crack free, majority of parts retiring from service are crack free and still have significant residual life remaining unused in them [1] and the issue of technology readiness for high thrust/weight ratio advanced engines [2]. Various other manufacturing factors such as mismachining, tool breakage or material defects not detected should also be considered in the above alternative method. In most of the military engine applications the consensus now is that Damage Tolerance Design (DTD) system will be able to provide means to account for material defects, manufacturing, handling and assembly defects and yet exploit the useful life of each engine component and boost their performance to higher levels without jeopardizing safety of the aircraft. Although this might be true for engine manufacturers with one customer (e.g. military airforce), the situation is not the same for commercial engine applications with hundreds of customers scarcely scattered around the continent.

The ever increasing interest in DTD approach has led to a tremendous worldwide effort on various aspects of the methodology, from analytical fracture mechanics tool development and crack growth modelling to characterizing engine alloys in terms of their crack growth behaviour under complex loading and environmental conditions and to NDI techniques as well as development of materials and components qualified for DTD. However, much work remains to be done before DTD can be successfully and routinely be applied to day-to-day critical engine component design, maintaining the same level of reliability and economical inspection of engine parts.

The present paper is dealing with an overview of the issues in SLA and potential applications of DTD. The issue is not aimed at comparing the two approaches, it is rather to address the state of the design and the associated problems from both SLA and DTD points of view. In this manner, first the life prediction system is reviewed involving both crack initiation and crack propagation aspects. Then introductory remarks on SLA are given followed by brief discussion on fracture mechanics analysis methods that have been used at Pratt and Whitney. In conclusion analytical results and experimental data are presented and discussed.

2. Life Prediction Methodologies

In general, the crack growth behaviour of critical engine components are as depicted in Fig. 1. The cracks found in most of the tested components so far, have been attributed to originate from surface finish. In some other cases such as fan disc rimslots and compressor blade fixing fillet radius (Figs. 2 & 3), the cracks were only originated at highly stressed locations. Almost all these components have been examined at their critical locations and non showed microstructural defects or anomalities.

Typical crack growth curves of critical parts as shown in Fig. 1 consist of two distinct parts namely, crack incubation period, during which energy accumulates at the crack origin site prior to the onset of crack growth. Once the crack growth is initiated, it then propagates through the material body and its growth is essentially dominated by the stress/deformation fields surrounding the crack tip. During the second part of the curve, the crack grows at a stable pace until instability occurs and very rapid crack growth leads to material failure within few cycles. The onset of instable crack growth is called the critical crack condition as shown in the above figure. Generally speaking, the usable life of a component is upto the point of instable crack growth. Reliable prediction of the usable life is very difficult due to the complexities involved in structural features of critical components, loading history and the material behaviour itself. Experience with most of critical rotors at P&WC has shown that a significant portion of the usable life lies in the incubation period and/or short crack formation (e.g. crack below 0.005" or 0.127 mm), where the effect of random microstructure on the fatigue life is more pronounced and is manifested in the form of a large scatter band in the fatigue data. This matter will be discussed in more detail in the following section.

Thus far, none of the existing lifing methodologies has been able to predict accurately the usable life of a component, especially when the reliability of predictions are addressed. The conventional SLA does not consider life benefits due to stable crack growth, whereas DTD may not be able to effectively take advantage of long incubation lives. Before further discussion, let us first review some relevant issues on SLA and DTD.

2.1 Safe Life Approach

The current safe life design system is dealing with a probabilistic life, which in North America it reflects one in a thousand chance of 1/32" (0.794mm) crack formation. The design criterion is established and supported by extensive engine tests, specimen and component tests. Test results as well as field data and residual test data of parts retired from service are continuously used to improve and update the system. The design system used at PWC is also supported by proven analytical tools and material databases. A long established set of explicit company procedures (see Appendix) is strongly enforced to ensure proper control and documentation of all critical components.

As mentioned above, the outcome of SLA is a probabilistic life as schematically shown in Fig. 4. Probabilistic lives are obtained from stress-life (S-N) curves. The curves, in turn, are obtained from specimen test and are calibrated with respect to component test data. Each point on these curves has been adjusted to match a minimum safe life derived from Weibull analysis of tested components (Fig. 5). The data points used in the Weibull analysis are actual lives to 1/32" (0.794mm) crack. During the spin pit testing of rotors, very often cracks grow to lengths well beyond 1/32" limit. This occurs as a result of too long inspection intervals, difficult to inspect locations, missing the crack or simply previous intention of identifying more cracks and other possible critical locations. In situations like this, the life to 1/32" (0.794mm) is obtained by subtracting the number of cycles associated with crack growth beyond the above limit from the total cycles accumulated during the test. This is done by opening crack faces and counting number of fatigue striations on the crack faces. Striations are counted along the crack depth on discrete observation areas. The crack growth rate is then calculated for each area and linearly interpolated between each pair of observation areas in order to give a full picture of crack growth along the crack depth. The same interpolation scheme is also used between the first observation area closest to the crack origin site and the apparent crack origin on the surface of the material.

In the case of compressor blade fixing fillet radius, it was possible to observe and count striations at a crack depth of 0.006" (0.152mm) from the free surface. In this manner, it is assumed that the crack propagation process right after the incubation period is associated with the mechanism of striation formation. Based on this hypothesis, the difference between the crack growth life obtained from striation counting (i.e. number of cycles between points A and B in Fig. 6) and the total number of cycles accumulated during the spin pit test is regarded as the crack incubation period. Once crack growth curves are established for individual tested components, then lives to 1/32" crack length are extracted from the curves using final crack aspect ratios and beach marks, etc. Such results are then subjected to Weibull statistical analysis to determine a single safe life data point for calibration of S-N design curves. The test is repeated at different stress levels to get points on the LCF curve.

In this context, a good portion of the scatter of the spin pit fatigue data falls on to the incubation period. The crack growth part, particularly the part associated with crack depth containing striations (i.e. segment B-C in Fig. 6) bears very little scatter as compared to the scatter in the total spin pit lives. An example of such observation is shown in Fig. 7, which indicates striation count data from five fan disc rimslot cracks, where all the five discs were identical in forging and shape specifications and they were all tested under identical spin pit test conditions. The trend in the slope of all the curves indicate crack growth with little scatter, whereas the horizontal shifts among the curves highlights scatter related to incubation periods and/or short crack growth. The above trend has been observed in most of test results of various critical components.

2.2 Damage Tolerance Design Approach

The DTD philosophy has already been discussed in various references [3, 4]. Therefore only its conceptual issues are briefly discussed here. More emphasis will be made on methods of assessing safety limits using fracture mechanics tools to get a better understanding of the crack behaviour.

Application of fracture mechanics concepts to gas turbine engine components is basically to predict residual life of these components in presence of material and structural defects induced during material fabrication and production or assembly of engine parts. The objective of DTD is to ensure that these flaws do not reach levels that impair the safety of the aircraft during the expected lifetime of its components. In this manner, crack length is a measure of damage and crack growth rate defines the rate of damage accumulation. The crack growth rate at the early stages is slow, whereas the fracture process occurs at the critical damage level and is almost instantaneous. The critical damage level or the crack size at the onset of crack instability is directly related to the level of the applied stress at this point. Thus, for a given material, this point will vary from component to component or locations within a component depending on the loading conditions and history. A material residual strength (Fig. 8) then indicates the onset of crack instability, which drops under cyclic load as the crack grows [4].

In general, parameters that affect crack growth or damage accumulation are initial crack size, load history, material behaviour (properties) and structural properties. This is briefly shown in Fig. 9, where any point below the curve corresponds to stable crack growth and any point on the curve indicates the onset of crack instability and instantaneous fracture. Points on this curve have values of stress intensity factor equal to the fracture toughness of the material. Under cyclic loads, the material's rate of crack growth is a function of the crack stress intensity factor, thus the accumulation of damage becomes a function of the latter factor. This is a foundation for Linear Elastic Fracture Mechanics (LEFM), which forms a basis for predicting residual fatigue life of a cracked structural element. The element material is characterized in terms of its crack growth rate (da/dN) versus the cyclic change in the crack driving force, i.e. the stress or strain intensity factor, which in general embodies the effect of stresses (strains) surrounding the crack, size and shape of the crack and local structural geometry. Considerable effort in analytical fracture mechanics has been devoted to the formulation and/or computation of stress (strain) intensity factor for complex stress/geometry combinations that model the cracked structure.

Accurate assessment of residual lives requires the interaction of many engineering disciplines. Elastic, elastoplastic and thermal analysis are needed to define the cyclic stress-strain history. Levels of dynamic loading has to be determined through analysis of vibrations under engine operating conditions. Engine alloys must be characterized in terms of their microstructural features and crack growth properties under isothermal and thermomechanical loading conditions reflecting engine environments. Non-Destructive Inspection (NDI) and Testing (NDT) are required to assess multitude of initial damage (material defects, surface finish, etc) and other factors that may influence the residual life of the end product.

Crack growth life prediction of gas turbine engine hot section components is difficult not only from the analytical, but also from the material behaviour point of view. The simple Paris law relationship, relating the instantaneous crack growth rate to the range of stress intensity factor (ΔK) is hardly ever sufficient to describe the behaviour of the crack when first order effects of cyclic frequency or dwell are present. Temperature operations introduce yet additional complexity due to phase relationship between the applied mechanical loads (external and centrifugal) and the thermal load (restricted expansions and thermal gradients). In addition, creep phenomenon may also interact and become a dominant factor. In general, cyclic strength of materials is significantly reduced at elevated temperatures. At lower temperature range of about 30% of melting temperature, the material strength is basically affected by cyclic fatigue process, whereas at elevated temperatures, the creep process becomes dominant. At the intermediate temperature, however, both fatigue and creep phenomena come into picture and interact.

In all divisions of Pratt & Whitney, extensive testing programs have been conducted on various engine alloys, which cover material characterization with respect to various engine operating condition parameters and their combinations for the assessment of both the crack initiation lives and crack propagation lives of these alloys. Analytical/numerical work involves development and calibration of fracture mechanics codes using the state of the art methods of computing stress intensity factors for all types of cracks that may appear in critical engine components. These tools are briefly discussed in the following section.

2.2.1 Fracture Mechanics Analysis

Several fracture mechanics codes have been used during our investigation. All the codes are equipped with flaw libraries, containing all the common flaws that may occur in critical engine components. The above codes are also equipped with a common material data base and elastic stress shakedown routines and data banks.

Some of these codes are based on non-finite element techniques of calculating stress intensity factors for quick and accurate computation of crack growth. They are based on the following methods of stress intensity factor calcualations :

- boundary integral method (BI)
- stress intensity formulae (SIF) of Newman-Raju [5-8]
- weight functions (WF)

Other capabilities using finite element and boundary element techniques are also being studied. Some of the most common flaw types that are modelled in these programs are :

- Surface cracks with a wide range of aspect ratios
- Quarter elliptical corner cracks
- Semi-elliptical surface crack in a hole
- Quarter elliptical corner crack in a hole
- Thru-crack
- Edge notch crack
- Burried Flaw
- Generalized center crack

Crack growth is modelled along one to four degrees of freedom and the stress field around the crack tip can be uniform, univariate/bivariate variable field and may contain tensile and bending stress components. Program capabilities also include analysis of crack growth under simple load histories and mission load spectrum, transitional cracks, crack closure [9], etc.

Crack growth laws that are used in the above programs are in one or more of the following forms :

(a) Generalized Forman's model, which is a modified version of Forman's rule [10]. The modified relation is

$$\left(\frac{da}{dN}\right)_i = C_0 \left[\frac{\left(\frac{1+C_1R_i}{1-C_2R_i}\right)\Delta K_i - \left(\frac{1+C_3R_i}{1-C_4R_i}\right)\Delta K_{th}}{\left(K_{ic} - \frac{\Delta K_i}{1-C_5R_i}\right)^m} (R_i)^{C_6} \right]^n$$

where C_i, K_{th} and K_{ic} are material constants and m is a function of temperature. K_i and R_i are stress intensity factor range and R-ratio, respectively for each major or minor cycle i under consideration.

(b) Klesnil-Lukas model, which again has material constants C_i, and K_{th} as functions of temperature :

$$\left(\frac{da}{dN}\right)_i = C_0 \left[\left\{ \frac{\Delta K_i}{1-C_1R_i} - \left[\frac{(1-\varepsilon_i)\ K_i}{1-C_2R_i} + \frac{\varepsilon_i \Delta K_1}{1-C_2R_1} \right]^{\alpha} \Delta K_{th}^{1-\alpha} \right\} R_i^{C_3} \right]^n$$

where ε_i is an overload factor (zero valued for simple cycles) [12].

(c) Sine hyperbolic model, based on a four constant hyperbolic sine equation in the form of :

$$\log\left(\frac{da}{dN}\right) = C_1 \sinh\left[C_2 \log(\Delta K + C_3)\right] + C_4$$

where coefficients C_1, C_2, C_3 and C_4 contain effects of frequency, stress ratio (R-ratio), temperature and mean stress [11].

(d) Willenborg model, which accounts for crack retardation by calculating a residual stress intensity factor due to the overload plastic zone and then correcting the value of R_i[13].

(e) Wheeler model [14] also accounts for crack retardation by correcting the subcycle crack growth rate for current plastic zone size and summing these values for the entire mission :

$$\frac{da}{dN} = \sum_{i=1}^{n} \varepsilon_i \left(\frac{da}{dN}\right)_i$$

$$\varepsilon_i = \left(\frac{a_i + a_i^{pl} - a_0}{a_1^{pl}}\right)^{\alpha}$$

Where a^{pl} is the plastic zone size, a_i and a_0 are the current and initial crack sizes, respectively and α is a material constant.

(f) Gema model [15], which also account for crack retardation using a fractional derivative form of Forman's equation - it is given by :

$$\left(\frac{d^{\lambda}a}{dN^{\lambda}}\right)_i = C\left[F(\Delta K_i, R_i)\right]^n$$

where λ_i, an overload ratio defined as :

$$\lambda_i = \left(\frac{K_i^{max}}{K_1^{max}}\right)\left[1 + 2\left(\frac{K_1^{min}}{K_1^{max}}\right) \mathrm{Ln}\left(\frac{K_i^{max} - K_i^{min}}{K_1^{max} - K_1^{min}}\right)\right]$$

with K_i^{min} and K_i^{max} denoting the minimum and maximum stress intensity factors at the i-th subcycle and subscript 1 referring to the overload cycle. The actual crack growth for i-th subcycle, $(da\ dN)_i$ may be given by :

$$\left(\frac{da}{dN}\right)_i = C\left\{\frac{\left[F(\Delta K_i, R_i)\right]^{\eta}}{\Gamma(1+\lambda_i)}(\Delta N_i)^{\lambda_i} - \frac{\left[F(\Delta K_{i-1}, R_{i-1})\right]^{\eta}}{\Gamma(1+\lambda_i)}(\Delta N_{i-1})^{\lambda_{i-1}}\right\}$$

where N_i represents the subcycle count within the current mission.

(g) Newman closure model [16-18], which accounts for effects of R-ratio exhibited in crack growth rate. It collapses multiple R-ratio trends into a single crack growth rate curve, which is a function of an effective load condition given by :

$$\Delta K_{eff} = \Delta K(1 - R_{eff})(1-R)$$

where $R_{eff} = A_0 + A_1 R + A_2 R^2 + A_3 R^3 \quad , R \geqslant 0$

$$R_{eff} = A_0 + A_1 R \quad , -1 < R < 0$$

Coefficents A_0 , A_1 , A_2 and A_3 are also functions of a 'crack constraint' constant, typically associated with cracked component thickness and is related to plane stress/plane strain conditions. This method also accounts for the level of applied stress relative to the material flow stress (average of yield and ultimate stresses).

In addition to the capabilities mentioned above, material and structural nonlinearities under creep-fatigue and thermomechanical fatigue are also being investigated through extensive modelling and testing.

Residual life analysis in general consists of the following steps :

- calculation of stress field of the uncracked component at the critical location(s)
- determination of size, type and orientation of flaw in the above location(s)
- fracture mechanics analysis using the aforementioned codes

Elasto-plastic stresses (2 or 3-dimensional) are obtained using finite element stress analysis procedures. In the case of using elastic finite element programs, the elastic stresses are shaken down using pre-processor routines. As for the selection of initial flaws, several NDI techniques are being used at PWC. Application of Eddy Current is being extended to various critical components. However, the NDI limits and their reliability as required in DTD has not been established. At the present time, for particular applications Pseudo-Initial Flaws (PIF) are used as discussed below.

2.2.2 Pseudo-Initial Flaws

Pseudo-Initial Flaws (PIF) were back calculated using fracture mechanics analysis so that the predicted lives based on them correlate well with observed fatigue lives [19,20]. PIF's are not actual flaws and so they do not bear any physical significance as far as the engine component material is concerned. The value of a PIF for a given part depends on the type of thermal/stress analysis used and the subsequent fracture mechanics modelling and the available fatigue data of the part. Additional fatigue data or any change in the analytical procedures may reset the PIF to a new value. However, PIF's have been found to be useful to estimate the crack growth behaviour of the critical components. Moreover, their calculation procedure is being simplified and automated so that PIF's may be obtained from statistical analysis of a population of optimum initial flaws. An optimum flaw itself is obtained for each single tested part through iterative fracture mechanics analyses. An example is shown in Fig. 10, which shows a balancing rivet hole at the rim of a compressor rotor where the low cycle fatigue (LCF) crack appeared in the form of an edge crack inside the hole and propagated through the rim and travelled towards the hub of the rotor. The growth of this crack was modelled in two phases as shown in Figs. 10a and 10b. In Phase I a corner crack model was used and crack growth up to plane A-A was predicted using this crack. The crack growth beyond plane A-A was modelled by a thru crack as shown in Fig. 10b assuming that unbroken material ligament during the transition period was negligible. The above model resulted in a crack growth curve as shown in Fig. 11, where striation count data points are also shown for comparison with predicted results. The change of slope in both predicted and experimental curves indicate the crack transition in the vicinity of plane A-A.

In the process of iterative fracture mechanics all the parameters are kept constant except the assumed initial flaw size. Iterations are made on this initial flaw until the predicted crack growth curve matches the observed fatigue data (See Fig. 12). The result is one optimum initial flaw per tested part. Thus, for a batch of identical parts tested under the same conditions a population of optimum flaws can be obtained. From this population a PIF can be chosen. Typical result of the above procedure is shown in the Weibull distribution of Fig. 13. In this figure each solid point corresponds to one component. A total of nine components had been tested, which were then analyzed (by iterative fracture mechanics analysis) resulting in nine optimum initial flaws. From the above figure one may choose a PIF with a given frequency of occurance. Alternatively, for each flaw in the above population one may examine the distribution of actual to predicted life (A/P) ratios (Fig. 14) Predicted lives are obtained from fracture mechanics analysis of the component. A PIF size of 0.005" (0.127mm) from the Weibull distribution of Fig. 13 results in $1<A/P<3$ as seen in Fig. 14. This range of A/P ratios was found adequate for spin pit testing purposes. The above PIF = 0.005" is used to determine spin pit inspection intervals such that number of inspections prior to a 1/32" crack formation is reduced to one [19].

3. Concluding Remarks

It was mentioned in the earlier sections of this paper that at P&WC, the design of the critical engine components is solely based on the Safe Life Approach (SLA), the reliability of which is a product of many years of accumulated experience on a variety of engine model products. However, in support of the SLA used, crack growth analysis of critical parts has become necessary for a better understanding of the behaviour of such components in terms of their potential remaining life after the service life has expired and also their behaviour in the presence of manufacturing defects and handling/assembly damages. Past experience with Pratt & Whitney engines during the previous decade indicate the need for such studies since majority of cracks in the engine parts were found to enemate from material defects and/or fatigue loading as shown in a Pie-chart in Fig. 15.

In this context, application of DTD requires interaction of many engineering fields for the development of proper engine alloys, design of damage tolerant fracture critical components and above all development of NDI methods and their instrumentation that should offer higher resolution of manufacturing and in-service defects with proven reliability and confidence of applications.

The latter issue is of utmost importance in the design of many critical engine components at P&WC. This is evident from the fatigue test results of the above components, such as the spin pit data of fan rimslot as shown in Fig. 7. The trend of crack growth curves in this figure indicates almost the same slope at depth levels above 0.020" (0.51mm), which implies very low scatter in the crack growth behaviour. Below this depth, little data (striations) could be found. Nevertheless, it can be seen from the figure that the component life below the above crack depth is at least 2.25 times higher than life above it. Moreover, all the above parts developed cracks at identical locations, where previous finite element stress analysis combined with blade/disc attachment contact analysis had already highlighted maximum principal stress components along directions that were perpendicular to the plane of crack growth. The above observations led to the conclusion that the crack initiation in the above locations occurred at the microstructure level and it has been strongly dominated by its surrounding stress field. This type of behaviour, e.g. longer incubation lives before cracks develop to reliable detectable sizes will become even more ponounced when component surface at critical location is treated to resist LCF cracking. Some treatments such as shot peening introduce residual stresses that may delay the crack growth at the early stages of propagation. Once a crack penetrates the treated surface layer, then its propagation into the untreated material underneath the layer will accelerate rapidly.

APPENDIX

A.0 Current Design System - Overview

The current design method at P&WC is based on SLA. In this method, critical parts are designed such that only one in a thousand identical parts may develop a 1/32" crack. The design system, basically utilizes Goodman-Johnson Stress-Life (S-N) curve and other similar empirical curves and/or equations. The design curves are initially based on material specimen testing, which are then extensively calibrated against actual hardware data. Hardware data, in turn, are obtained from controlled component test (spin pit) and field report information. Design system is always kept up-to-date by a constant feedback of field data and spin pit test data into design tools. An overview of lifing system calibration is shown in Fig. 16. As it can be seen from this figure, the lifing methodology contains two major calibration loops: one to adjust preliminary design S-N curves to match actual hardware spin pit data and the other to update factors that are used to convert spin pit test data under controlled test conditions to mission load and temperature/environment conditions defined by the customer using available data which can include field experience. In both calibration loops, statistical analyses are carried out to ensure the required reliability of the engine critical components. This involves Weibull and scatter factor analysis of data to set proper reliability and confidence to the design S-N curves. Weibull analysis and certainty factors are also used for optimum conversion of spin pit cycles to mission cycles.

A.1 Critical Part LCF System

P&WC system for design of gas turbine engine critical components has been tailored for both general aviation operations and commuter operations.

Critical parts are basically rotating parts and are defined as parts whose failure result in probable loss of the aircraft as a result of non-containment or, for single engine aircraft, power loss preventing sustained flight either due to direct part failure or by causing other progressive part failures. In this context, fan hubs, boost hubs, compressor discs/blisks/drum rotors, centrifugal compressors (impellers) and turbine discs, spacers and cover plates (if analysis shows a possibility of failure non-containment) are regarded as P&WC's engine critical components.

The basic requirements for Low Cycle Fatigue (LCF) lifing system of P&WC engine components are as follows :

- Work to an "approved system"
- Established and reliable calculation methods
- Verification by test and field data
- Maintain consistency

Each of the above items are discussed in more detail in the following sections.

A.1.1 Approved System

Design of critical engine components is based on a system, which is outlined to and approved by Transport Canada. The design methodology, which has been accepted by FAA, US, UK, etc has been consistently applied to all certified engine models in worldwide services.

The general policy of the approved system is to use calibrated advanced heat transfer, stress analysis and lifing procedures. It also calls for development of forging and manufacturing processes well suited to specific designs and applications. All computations and material selections are based on statistical distribution of material properties. The entire system is continuously being calibrated and updated using test data and field data, whenever they become available. At the same time applicability of past experience to new designs are verified.

The approved system establishes operating limitations for individual critical parts by specific maximum allowable number of start-stop stress cycles on the part. Hence, for each one part, one life limit is calculated. The guaranteed life is then based on the calculated life, rig test results, margin for future deviations, confidence in mission definition, customers' requirements, marketing strategy, etc.

A.1.2 Calculation Methods

Calculations contained in the design of critical and non-critical engine components are :

- Stress analysis (including transient and vibration)
- Heat transfer analysis
- Foreign Object Damage (FOD)
- Containment
- Mission analysis
- LCF life analysis (including statistical study)

Finite element (2D and/or 3D) elastic and elasto-plastic analyses are used to evaluate local stresses at critical locations. The analyses are supported by CATIA/CAEDS grid generation softwares for accurate and representative modelling of engine components and their critical geometrical details (See Figs. 17-19). Computer codes are developed in-house and are equipped with special features such as multilevel substructure computing, grid size error indicator, contact analysis of blade/disc attachments etc. Post processing support include stress/deflection contouring, calculation of stress concentration factors, error indicators and other necessary features.

In all the above stress analyses, temperature distributions (gradients) are pre-computed using heat transfer analysis procedures. Reliable temperature gradients are obtained by considering heat transfer to or from the disc, which in the case of a turbine disc occurs through four processes:
(a) forced convection between

- hot gas flow and the blade airfoils
- cooling air flows and the sides of the disc
- disc hub and gas in cavities
- the disc and a flow through axial and radial holes in the disc

(b) metal to metal conduction from blade to disc,

(c) forced convection or conduction to air in the gaps between the blade and disc fixing.

(d) radiation across the gap between the blade fixings and disc fixings.

Heat transfer analysis is mainly carried out in the hot engine section and centrifugal compressor area. Computer codes available are capable of computing steady-state, transient and thermal shock analyses.

Another type of analysis for cold engine components (fan blades, fan hub, 1st stage compressor, nose cone, etc) is F.O.D. and D.O.D. It provides adequate structural integrity resistance for small engine compressorss under foreign object impact damage by utilizing nonlinear numerical tools, in which bird impact, blade part and multi-blade impact are modelled. The design tool/criteria are being developed to furthermore include multi-blade impact/interaction, bird trajectory effects, bird debris/ice trajectory damage to core/by-pass stators and bird impact load effects on shaft/bearing rotor dynamics and bearing housing.

Various engine casings and other static structures are also designed for containment in the event of blade-off, spacer failure or cover plate failure, etc, using finite element transient analysis codes. The above F.O.D and containment codes are verified and calibrated through extensive rig testing on actual hardware.

In order to calculate LCF life of critical components, first flight missions are analyzed. In the case of turbine discs for example, excursions of speed and temperature during a mission represent a sequence of peaks and valleys, which are converted into a system of full and partial LCF cycles using rain flow cycle counting technique. Then, each repetition of a given excursion is considered to produce the same amount of damage regardless of whether the excursion occurs at the beginning or the end of the service life. A linear damage accumulation hypothesis (Miner's rule) is then used to approximate the total damage due to one excursion. The mission represents the actual operation of a worst production engine, which has experienced 50% of the acceptable deterioration in performance.

Low cycle fatigue life is essentially based on curves. As mentioned earlier, the curves were originally based on material specimen testing, which are then calibrated using actual hardware test data. Data reduction in both specimen testing and component testing are subjected to statistical analyses in order to predict component lives with the required levels of reliability and confidence.

A.1.3 Verification by Test

The LCF lifing system is supported by four types of verification tests :

- proof LCF test: carried out on unmodified production sample to verify the identification of the critical location and to ensure that there are no life debits arising from particular micro-structures.
- basic LCF life test: applied to numerous samples to determine the life to a specific level of reliability. Since thermal and environmental stresses/effects are not induced during the above tests, samples may need modifications to correctly simulate the relative criticality among the different life-limiting locations in the actual engine environment.
- overspeed tests: to determine burst speed of rotors and other margins of safety.
- residual life test: carried out on uncracked parts retired from service and is done to further obtain margins of safety and certainly factors and also to update factors that are used to convert spin pit lives to mission cycles.
- In case of fracture in any of the above tests, cracks are analyzed and monitored to reveal crack depth, aspect ratio and propagation rate. Fractographic examination of the crack surfaces are conducted to define the crack origin and striation counting. Crack growth data are then stored in a central data base file for material and component identification.

In all the above tests, samples conforming to production standards are subjected to normal production inspection procedures prior to and after the test.

Effects of external forces and engine temperatures are closely simulated and test r.p.m. is also selected based on maximum local stresses in the critical area of the component. Accelerated tests may also be done provided that the subsequent stresses and/or new stress distributions do not introduce new effects in the slope of the S-N curves and the position of the critical location or its mode of cracking.

As a final verification, the striation counting on parts performed are extended far beyond the 1/32" limit. Using number of striations beyond this length, crack propagation cycles are subtracted from the total test cycles to obtain a more realistic life to 1/32" (0.794mm) for the test sample.

A.1.4 Maintaining Consistency

Consistency of material and product quality and their conformance with the minimum requirements are established through the following :

a) Quality standards are continuously monitored after certification from the source of ingots through conversion, forging and manufacturing.

b) Monitoring of material properties of each forging heats using a Material Statistical Data System (MSDS). In this system acceptance of forging heats rely on verification of microstructure, creep, stress rupture, hot tensile properties and room temperature tensile properties. Any significant deviation in properties is evaluated as to potential impact on guaranteed LCF life.

c) Statistical acceptance of lots: Levels of testing required from acceptance sampling procedure depends on the experience of the vendor relative to the critical component and the consistency of the forgings produced. The normal sampling plan is based on MIL-STD-414. Number of forgings to be cut up depends on the lot size as shown below.

Lot Size	Minimum Number Of Forgings to be Tested	Acceptability Constant
Up to 65	7	2.00
66 to 110	10	2.11
111 to 180	15	2.20
181 to 300	20	2.24

Upon testing the required number of forgings, the quality level for each location should equal or exceed the acceptability constant in order to accept the heat code.

d) LCF testing of suspect heats: deviating or suspect heat codes are verified through spin pit testing before acceptance.

e) All component sources (ingot, forging, manufacturing) have to be substantiated extensively using actual component tests before identifying a source as an approved source.

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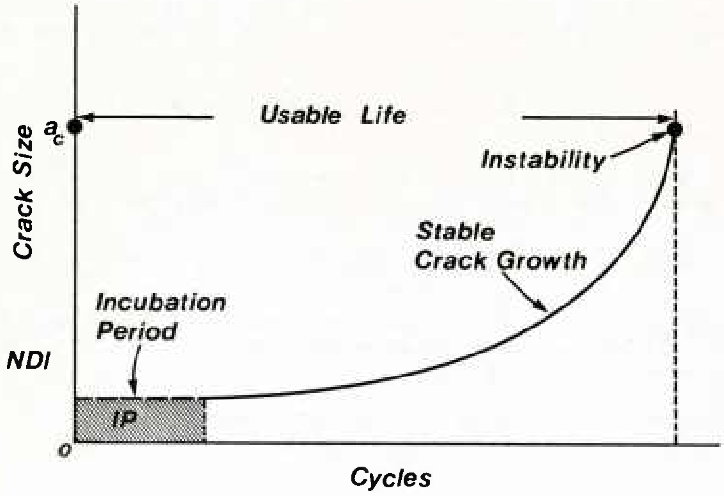

FIG. 1 GENERAL PRESENTATION OF CRACK GROWTH BEHAVIOUR OF CRITICAL ENGINE COMPONENTS.

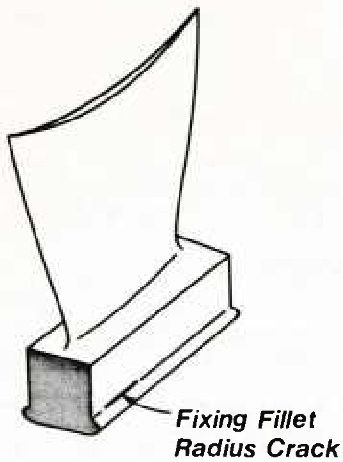

FIG. 2 AXIAL COMPRESSOR BLADE FIXING CRACK.

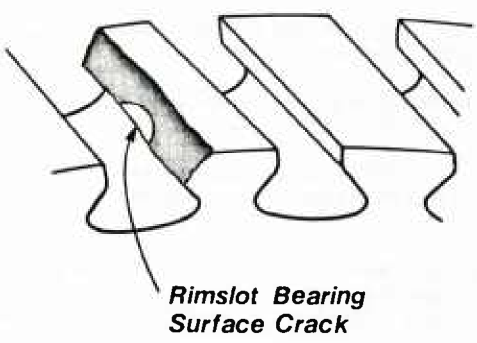

FIG. 3 FAN ROTOR RIMSLOT CRACK.

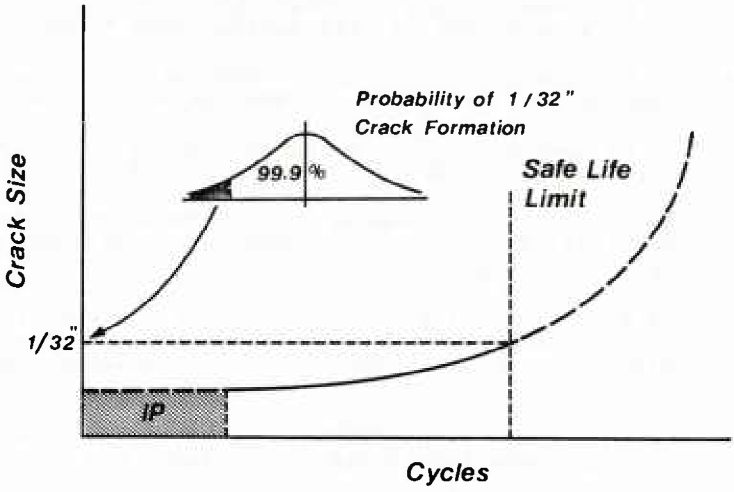

FIG. 4 PRESENTATION OF CRACK GROWTH CURVE IN THE SAFE LIFE APPROACH.

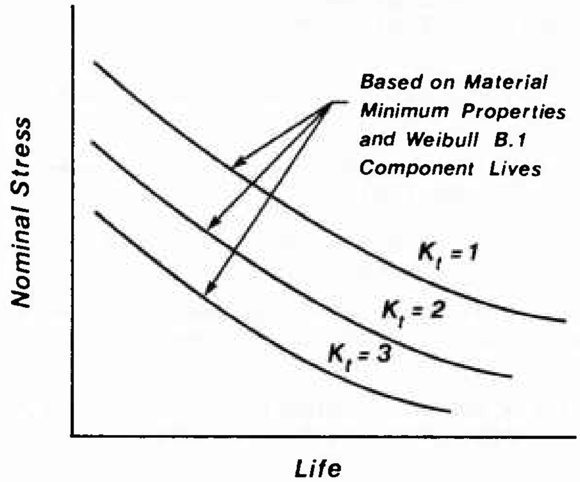

FIG. 5 TYPICAL STRESS VS LIFE (S-N) CURVE.

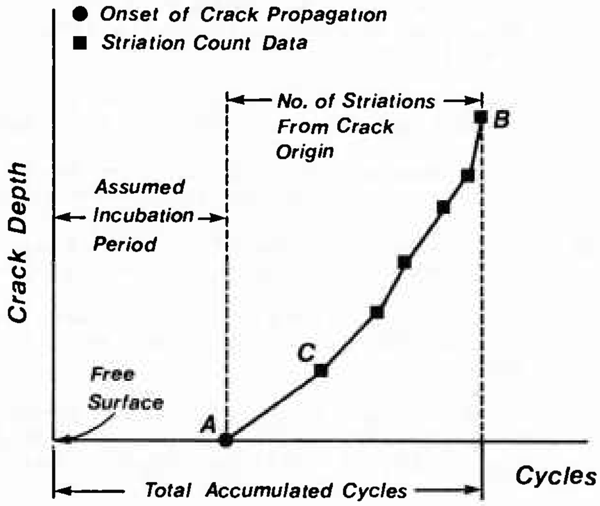

FIG. 6 STRIATION COUNT PROCEDURE.

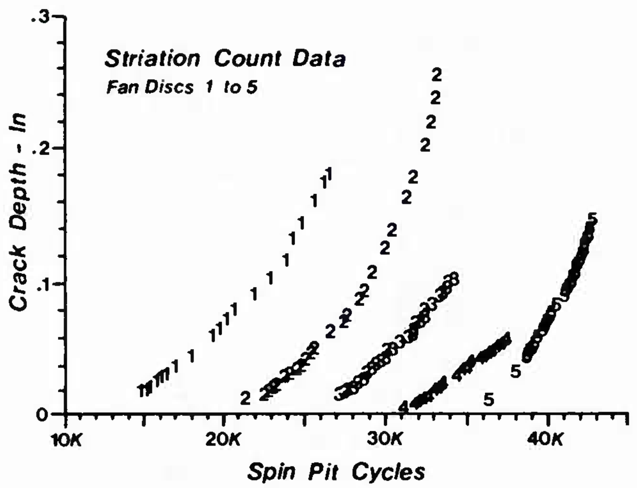

FIG. 7 TYPICAL FAN DISC RIMSLOT STRIATION COUNT DATA.

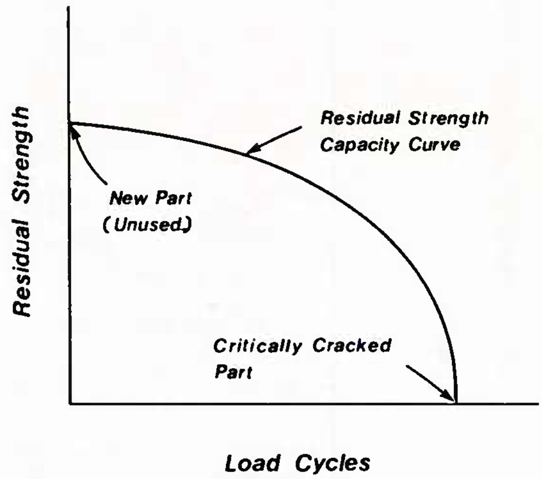

FIG. 8 RESIDUAL STRENGTH CURVE

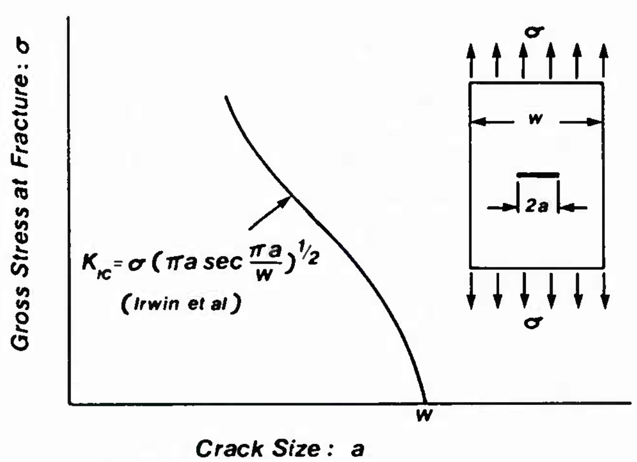

FIG. 9 FRACTURE TOUGHNESS CURVE

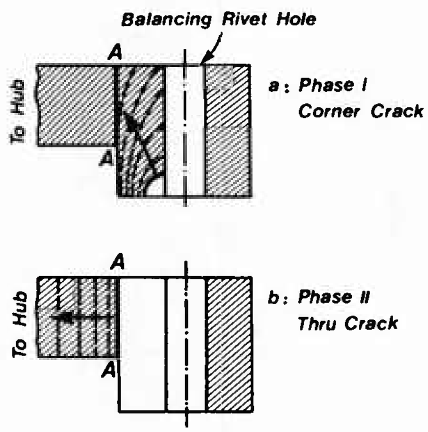

FIG. 10 CRACK GROWTH MODELLING OF A COMPRESSOR BLISK BALANCING RIVET HOLE CRACK.

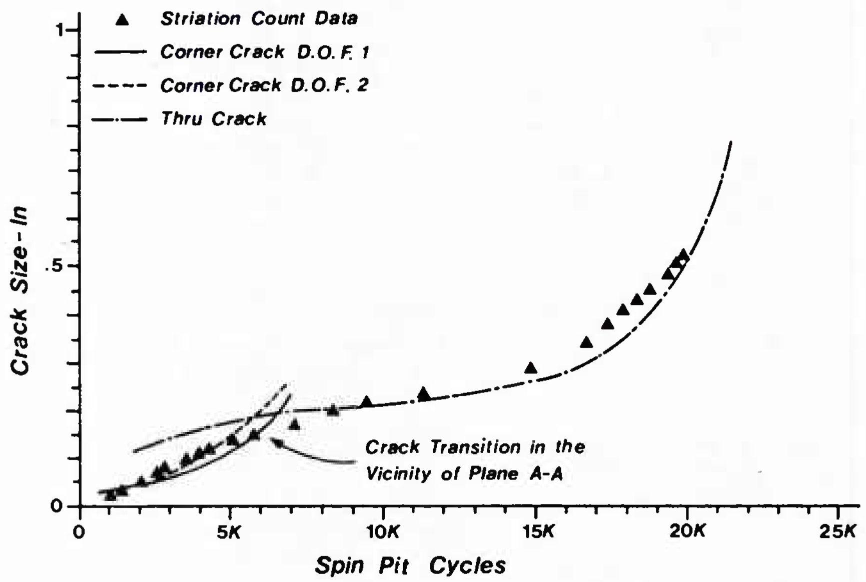

FIG. 11 STRIATION COUNT DATA FOR THE TRANSITIONAL CRACK OF FIG. 10

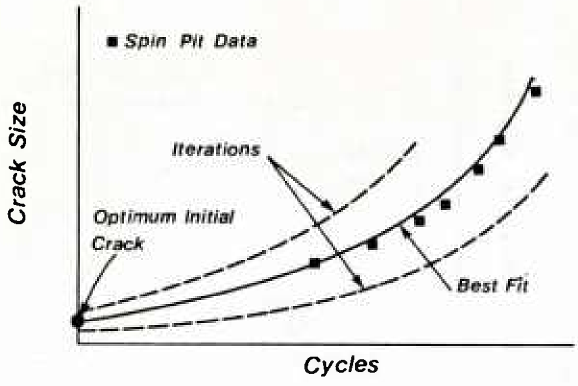

FIG. 12 ITERATIVE FRACTURE MECHANICS ANALYSIS TO DETERMINE OPTIMUM INITIAL FLAW.

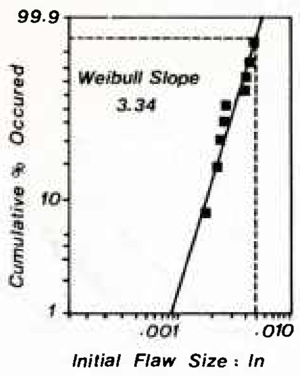

FIG. 13 WEIBULL ANALYSIS OF OPTIMUM FLAWS.

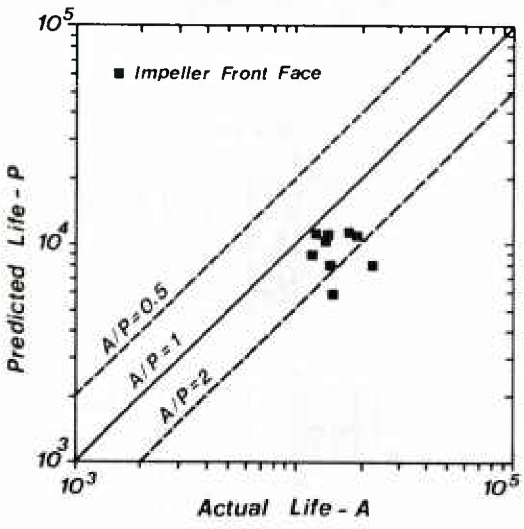

FIG. 14 ANALYSIS OF PREDICTED LIVES VS ACTUAL LIVES USING A PIF OF a=0.005" & a/c = 0.25.

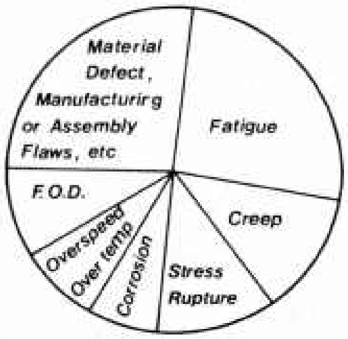

FIG. 15 PIE-CHART PRESENTATION OF LAST DECADE COMPONENT CRACK PROBLEMS.

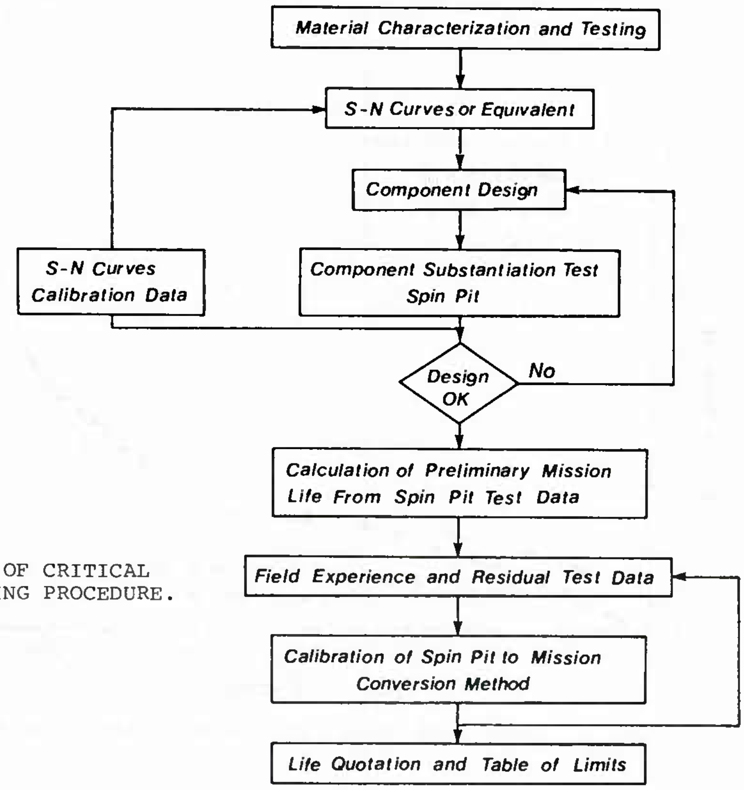

FIG. 16 OVERVIEW OF CRITICAL PART LIFING PROCEDURE.

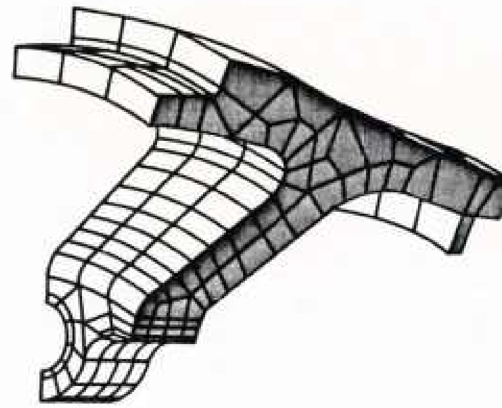

FIG. 17 3-D FINITE ELEMENT GRID OF A COMPRESSOR BLISK (AIRFOILS NOT SHOWN). PERIODIC SEGMENT

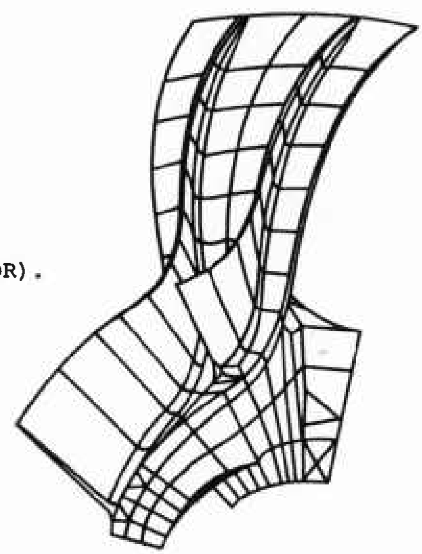

FIG. 18 3-D FINITE ELEMENT GRID OF AN IMPELLER (CENTRIFUGAL COMPRESSOR). PERIODIC SEGMENT

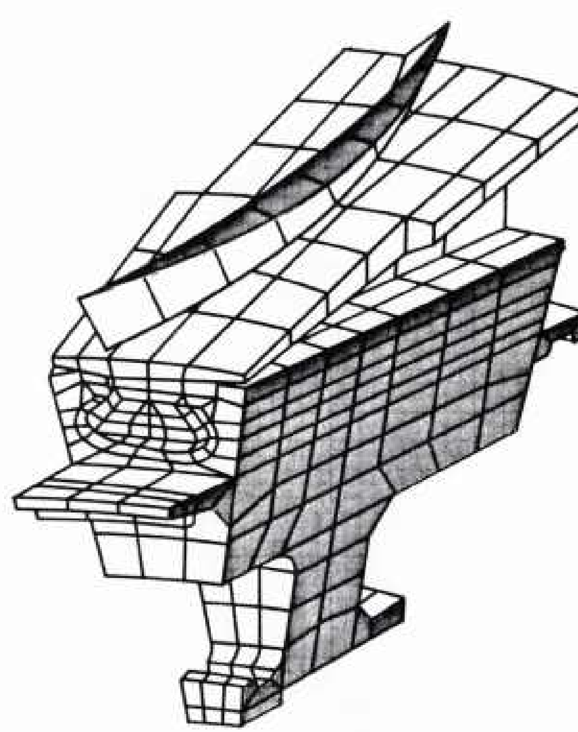

FIG. 19 COARSE 3-D FINITE ELEMENT GRID OF A FAN ROTOR (LOW PRESSURE COMPRESSOR) WITH PARTIAL AIRFOIL MODELLING. PERIODIC SEGMENT

FUTURE TECHNOLOGY REQUIREMENTS (RELATED TO DEFECTS AND QUANTITATIVE MATERIAL BEHAVIOR TO AID IN IMPLEMENTATION OF DAMAGE TOLERANCE FOR DESIGN OF ENGINE STRUCTURES).

David W. Hoeppner, P.E., Ph.D.
Chair and Professor
Department of Mechanical Engineering
University of Utah
3210 Merrill Engineering Building
Salt Lake City, Utah 84112
U.S.A.

SUMMARY

This paper presents issues that are viewed as critical to an increased rate of implementation of damage tolerance concepts to the design of critical engine components. Areas emphasized are "defect" definition, inspection, modelling of discontinuities, evaluation of subcritical crack propagation behavior, development of material properties data banks, increasing the understanding of microstructural influences on material behavior, methods of comparing materials and various processing methods for modifying "defect" tolerance, and development of lifing methods that will enable designers, operators, and maintenance personnel to understand and incorporate all the concepts necessary to assure integrity of critical engine components. Various conclusions, including suggestions for expanding knowledge, are presented.

INTRODUCTION

The increased implementation of damage tolerance for safety critical components of gas turbines will require improvement of knowledge, expansion of data bases, and generally improved methods to incorporate the concepts. The following sections will present those areas that are viewed as critical to such activity. Furthermore, those areas where additional research or an expanded data base is needed are highlighted. Previous AGARD meetings (1-3) have provided the background for much of the information.

As well, the information from the first workshop in the series of four scheduled related to this subject (4) has been helpful in providing background as well.

Defects*

It is essential that a clear definiton of "defect" be established related to the materials in question. This is essential for all materials typically used in gas turbines. As well, it is necessary to formulate an exact description of a "defect" particularly related to its source. "Defects" must be described for:

- Titanium alloys,
- Nickel base alloys,
- Advanced materials.

Subsequently, it must be determined whether a "defect" can be either eliminated or controlled in distribution, type, size and location.

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* I prefer use of the terminology discontinuity, heterogeneity until it has been established whether a "defect" actually exists related to the component material, manufacturing, geometry specification.

It is also essential that the probable locations of "defect" be established with respect to the applied stress or strength field and also with respect to the inspection that will be used in an attempt to locate it or crack it during growth.

Many of the critical components for gas turbines have been designed using the "safe-life" or "crack initiation" concept. The basic concept is to utilize the mathetmatical theories of elasticity and plasticity with the inherent assumptions that the materials used are ideal continua (homogeneous, continuous, isotropic, free of "defects"). The inherent assumptions make the analysis tractable but also are simplistic. The safe-life fatigue design concept as traditionally used is simplistically stated as follows:

> Analysis and testing is required to show that the probability of failure is extremely remote for the assumed life of the structure. Either finite life or "infinite life" concepts are used. Since the materials are usually assumed "defect-free" they are also believed to be crack free. This led to the development of the so-called "crack initiation" fatigue design concept. This can be represented by the following quote:
>
> "Cyclic crack initiation concepts in fatigue have been used to advantage in engineering design ever since the phenomenon of fatigue was first recognized over a century ago. *Crack initiation is considered to be a singular event. Details of how the event evolves are considered to be irrelevant. A unified precise definition of the event has evaded researchers over the years with the result that a broad range of definitions have been used and considerable confusion has resulted.*** For example, in the past in the area of isothermal low-cycle fatigue testing, failure into two pieces (or strong indications of impending failure, such as load drop-off in strain controlled tests) of the relatively small (1 cm) diameter specimens was usually considered as the initiation event. With the advent of in situ SEM equipment and surface replication techniques, the crack initiation event can be pegged at a much smaller crack size. The lower the cyclic lifetime of a given material, the smaller the fraction of total life consumed in generating a crack which is smaller than the specimen dimension. Stated alternately, cracks emerge sooner in low-cycle fatigue than in high cycle fatigue. In the case of recent high temperature low-cycle fatigue results reported by Moreno, et. al. (Ref. 67) for a cast nickel-base superalloy, surface crack lengths of 0.030 in. "initiated" as early as 0.1 to 0.6 of the total separation life for specimens with surface crack lengths at separation of about 0.40 to 0.50 in." (5) (Ref. 67 is in Ref. 5).

It is clear from this quote that the exact definition of the end of life in the safe-life design concept can get pretty fuzzy. An extreme case of this would be to define a crack as having initiated when it was of a certain length with respect to the grain size, let us say that you would define crack initiation to equivalent to three grain diameters and you were dealing with single crystals. It should be clear to the reader that by this definition the cracked component would never have "initiated" cracking even though the crack could have propagated to cause the component to fail in practice. So, it is imperative that the concept of end of life be specified related to either a crack detection criteria or some other end of life criteria. This author has suggested on many occasions (see e.g. D.W. Hoeppner in Ref. 3) that the initiation concept could be described more appropriately as a crack detection concept. Furthermore, if components were assumed to be made of initially crack free or "defect-free" material then there may be a crack formation (nucleation, generation) stage that would preceed crack propagation. Currently, there is no known method to establish where a crack formation stage ends (given it occurs) and initiation of crack propagation begins. Thus, the end-of-life criteria usually employed in the safe-life concept should be referred to as a crack detection criteria rather than an "initiation" criteria. Surely one of the greatest needs we have is to more clearly state our terms and their intended meaning.

Having previously defined the "defect" and discussed it in relation to its type, location, size, and inspectibility, it is imperative that mathematical models be formulated to describe the "defect". It is here that the various aspects of fracture mechanics (damage mechanics) become extremely useful. However, it is necessary at this point to decide what model should be used. As one begins to

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** Italics were added by DWH.

model discontinuities and heterogeneities one must immediately ask what role if any the local material structure will play in the formulation of a realistic (rational) model. Furthermore, if a model can be formulated, we must ascertain if data can be collected related to the potential propagation of a discontinuity that could be input to design considerations. Some of the data that we need are listed in Table 1.

Table 1. Data Needed for Damage Tolerant Design

- K_{IC} (or other instability parameter).
- ΔK_I vs. da/dN or K_{Imax} vs. da/dt or other crack propagation data.
- Thresholds, ΔK_I , or other mode of propagation.
- Spectrum effects (cold turb, mini turb, hot turb, AMT, etc.).
- Mixed mode propagation.
- Multiaxial stress effects.
- Temperature effects.
- Environment effects.

It is quite clear that standard test procedures and .standard data representation procedures need to exist in order to allow generation of data. In addition, it is necessary that data be capable of storage in computers and this would allow the insertion of data into computer assisted materials engineering and computer assisted design algorithms. These issues have been addressed to a certain extent and are currently being addressed under both NATO/AGARD and Turbistan auspices, but additional information is needed; particularly related to thresholds and the five last items in Table 1.

Data Needed

In order to generate the data necessary for Table 1, it is essential that final fracture or critical instability data be available for the relevant model that exists and is to be used. Whenever possible, as previously indicated, a test standard is desired. It is essential that an adequate, statistically based sample of data be available to assure that designs can be formulated in a recognition that there is a statistical distribution of the final fracture data as well.

Also, as indicated, it is necessary that subcritical crack growth rate data be available over the range of propagation of interest for the particular conditions of interest. In this regard, data must be available that will account for various extraneous effects such as the following:

- Spectrum,
- Temperature,
- Time,
- Environment,
- Complex Stress Fields,
- Multiple "Defects".

The issues listed above are extremely complex and require a great deal more information than that which we currently have available.

In the utilization of a damage tolerant design philosophy, the items listed below must be included.

STRUCTURAL FATIGUE DESIGN PHILOSOPHIES

DAMAGE TOLERANT OR
FAIL-SAFE DESIGN

REQUIRE A DAMAGED STRUCTURE TO CONTINUE SATISFACTORY PERFORMANCE UNTIL "DAMAGE" IS DISCOVERED AND REMEDIAL ACTION COMPLETE.

- CRITICAL COMPONENTS (AREA OF COMPONENTS)
- MULTIPLE LOAD PATHS (REDUNDANCY)
- CRACK STOPPERS
- PROOF TESTING
- CRACK GROWTH PREDICTION TECHNIQUES
- INSPECTION (TYPES)

The overall definition of damage must be addressed in a serious way as is indicated in the above table and furthermore the issues shown in figure one must be addressed. Certainly, we will need a great deal more information related to the items listed in Figure 1 as individual mechanisms of producing damage, but we will also need a great deal more information on the various synergisms that can occur related to the production of damage.

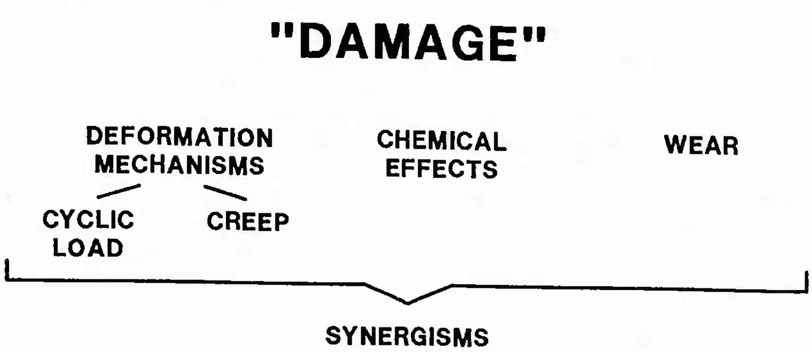

Fig. 1: "Damage" Mechanicms to be Considered in Damage Tolerant Design.

Damage can be produced that is either external or internal to the component and an example of these types of considerations is shown in Figure 2. Although we have attempted to characterize the various "defects", that can exist in a component initially, and we have made attempts to characterize the damage that can develop during storage, use, abuse, and the various other issues previously mentioned, it is essential that we develop increased insight into formulating a more complete understanding of the various damage producing mechanisms. It is in this area that future technology requirements will be extremely hard pressed to come up with answers in the short term. There is little doubt that we are making significant progress in this area, but certainly need to accelerate our progress a great deal to increase our confidence in the design rationale.

D

DAMAGE

EXTERNAL
(TO COMPONENT)

SOME EXAMPLES:

- Slip/slip bands
- Twins
- Scars
- Pits
- Scratches
- Cracks

Mechancial Loads,
Environments, T, t
Influence dD/dt

INTERNAL
(TO COMPONENT)

SOME EXAMPLES:

- Pores
- Cavities
- Dislocation structures
- HIDS
- LADS
- Other particles

Mechancial Loads,
Environments, T, t
Influence dD/dt

Thus,

1. The Damage and Damage Process Must be Defined, e.g., crack growth, pore growth.

2. The Rate of Damage Growth Must be Known for the Specific Conditions of Interest.

3. The Damage Must be Observable and Trackable.

Fig. 2: Examples of Damage.

Subsequent to an attempt to recognize the type of damage and characterize the initial discontinuities, approaches that utilize conceptual diagrams such as shown in Figure 3 and formulate the methods to deal with these stages of nucleation and damage growth as shown in Figure 4 have previously been suggested (see e.g. R. Jeal and D. Hoeppner articles in Ref. 1 and 3). It is in numerous of these areas that additional technology is required in order to allow us to implement damage tolerant concepts into critical engine components with greater speed and reliability. In the remaining sections of this paper I will discuss areas where it is viewed that we need better and improved techniques to handle the job of bringing these things together in various methods.

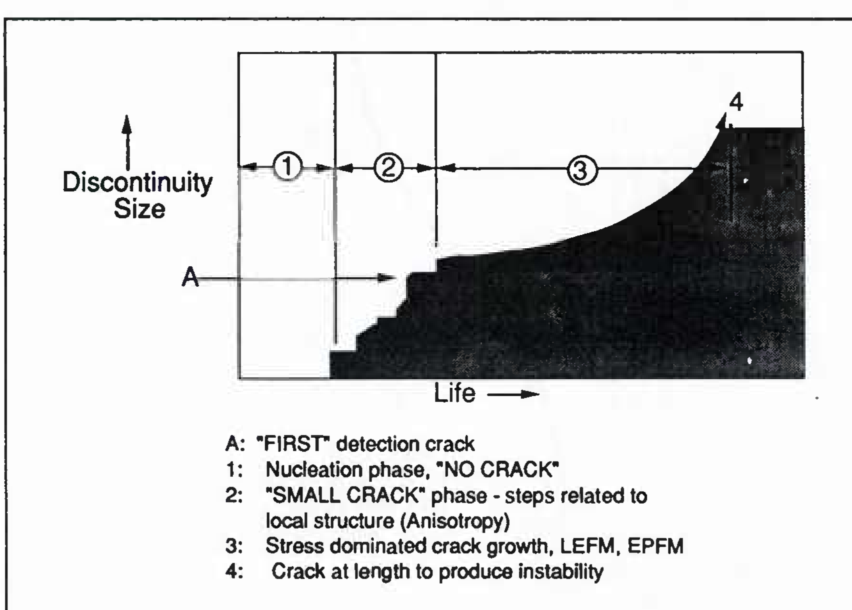

Fig. 3: Damage Growth Diagram Previously Proposed
(Reference 3 e.g.)

METHODS FOR EACH LIFE PHASE

NUCLEATION	"SMALL CRACK" GROWTH	STRESS DOMINATED CRACK GROWTH	FAILURE (FRACTURE)
Material failure mechanism with appropriate stress/strain life data	Crack Prop. threshold related to structure (micro)	Fracture mechanics •similitude •boundary cond. LEFM / EPFM ?	K_{Ic} etc
Nucleated discontinuity (not inherent) type, size, location	Structure dominated crack growth	Data base**	C.O.D.
Presence of malignant D*, H*	Mechanisms, rate	Appropriate stress intensity factor	Tensile/ compressive buckling
Possibility of extraneous effects •Corrosion •Fretting •Creep •Mechanical damage	Onset of stress dominated crack growth	Initial D*, H* size, location, type	
	Effects of •R ratio •Stress state •Environment (chem, T) •Spectrum -waveform	Effects of •R ratio •Stress state •Environment (chem, T) •Spectrum -waveform	

Fig. 4: Methods for Each Life Phase to Implement Damage Tolerant Design.

As we deal with the issues covered in Figure 4, it is important to formulate methods that are based on physical models formulated from rational and probabilistic based principles that allow our methods to be used in a predictive fashion. However, in many cases, many of the models that we have been utilizing are correlative at best. Even though these models thus are correlative and can be used within the bounds of the data, there is a very high degree of risk when we utilize these data outside the bounds by methods of extrapolation. It is obvious that this is the case when empirical relationships are used even though we are developing data based on fatigue crack propagation concepts or other more complex damage propagation modes.

The generalized curve for fatigue crack growth is shown in Figure 5 and it illustrates the type of data that are needed in order to proceed with making a fatigue crack propagation life prediction. In a following section I will briefly discuss the necessity to fit these data. However, before I do, I would like to briefly

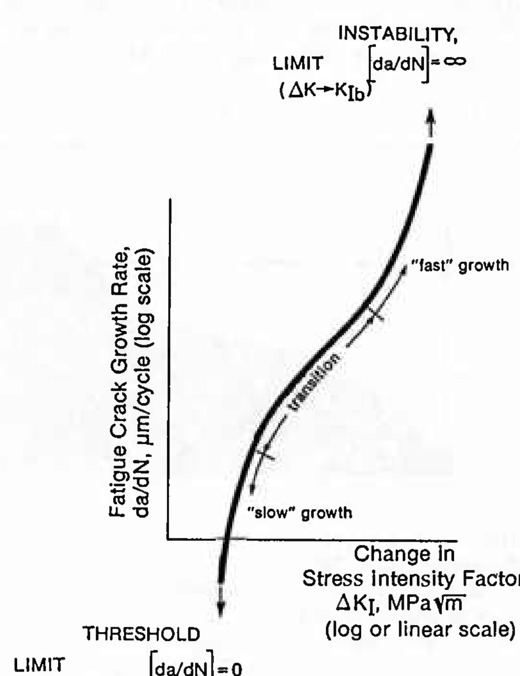

Fig. 5: Typical Fatigue Crack Propagation Curve.

touch on the concept of sustained loading. In the event that we have sustained loading, it is essential that we develop data that are applicable, as illustrated in Figure 6, to handle the mechanisms that could produce the respective mechanisms of damage production and propagation listed in Figure 6.

SUSTAINED LOAD-GENERAL (SLG)

- **Creep**
- **Environmentally assisted (SLG)**
  - **Stress corrosion cracking**
  - **Oxidation**
  - **Hydrogen embrittlement**
    - **External**
    - **Internal**

Fig. 6: Sustained Load Potential Damage Mechanisms

Figure 7 provides a summary of a method for estimating crack growth under creep or other similar sustained loading conditions. This type of methodology is strongly dependent upon the inputs of data that are obtained for the materials under consideration for the conditions of interest. Thus, the development of increased data banks for creep crack growth and other forms of sustained load crack growth is absolutely essential to the expansion of the development of improved damage tolerance capability for engine structure.

SUMMARY OF METHOD FOR ESTIMATING CREEP CRACK GROWTH

1. Failure criterion decision - creep crack growth.
2. Obtain data on material for conditions of interest for the structure:
  a) Consider; crack sizes, locations in product form,
  b) Consider; thermal and chemical environment,
  c) Consider; load and environmental spectra.
3. Represent data.
4. Integrate the expression in (3) above to obtain:
  a) Life (time) for crack size load/environment conditions.
  Could be total life or time between inspection.
or b) Crack size for given life and load/environment conditions.
or c) Load/environment for given life and crack size (inspection) conditions.
5. Establish trade-offs.
6. Write specifications and purchase documents to obtain design goals.

Fig. 7: Summary of a Method for Estimating Crack Growth

CYCLIC LOADING

The problem of loading under cyclic load forms (often called, of course, "fatigue") is a continuing problem, but, generally speaking, we have utilized various linear elastic fracture mechanics techniques that recognize the numerous parameters that influence fatigue crack growth. Equation (1) of Figure 8 illustrates the often referred to Paris expression for fatigue crack propagation, but is applicable for only an extremely limited portion of the fatigue life and, consequently, must be used with care. Nonetheless, it is useful in that it does emphasize the correlative capability between fatigue crack growth rate (da/dN) and ΔK_I, the Mode I stress intensity factor. It is important to emphasize here that

FATIGUE CRACK GROWTH

$$\frac{da}{dN} = C\,(\Delta K)^n$$

$$\Delta N = \int_{a_i}^{a_c} \frac{da}{f(\Delta K, K_{max})}$$

INTEGRATE FOR LIFE

Figure 8. Paris Relation for Use in Fatigue Crack Growth Life Estimation

there are numerous parameters that influence fatigue crack propagation. These are depicted in the equation shown in Figure 9.

FATIGUE CRACK GROWTH

$$\frac{da}{dN} = f(\underbrace{a, \sigma}_{K}\{\varepsilon\}, \text{env.}, T, \omega, P\{t\})$$

a - Crack Size	env - Environment
σ - Stress	T - Temperature
ε - Strain	ω - Frequency
	P{t} - Carrier Function

Fig. 9: Parameters that are Known to Influence Fatigue Crack Propagation

It must be emphasized that environment, temperature, frequency, and carrier function or type of spectrum are known to influence the fatigue crack propagation behavior of numerous materials. Thus, the development of a data bank on fatigue crack propagation data to be used to correlate to full scale structural behavior is extremely important for expanding damage tolerance capability. The complexity of the fatigue process in some gas turbine materials, such as titanium and nickel base superalloys, makes this an extremely great challenge.

Shortly after Paris introduced his equation in around 1962, numerous persons, including this author, suggested that the fatigue crack propagation behavior of materials should be represented more completely by recognizing that fatigue crack propagation behavioral curves would have the shape of a general sigmoid type behavior with modifications dependent upon the influences of temperature, environment, frequency, and wave form. As well, it was early on recognized by such persons as Royce Foreman, Ken Walker, and other investigators that R ratio could play a significant effect. Consequently, in 1965, Royce Foreman, et. al. introduced the modified Paris equation which is the first equation shown in Figure 10. Shortly thereafter, my colleagues and I, then at Battelle Memorial Institute, in work being done there, (continued later by Charles Fedderson and Carl Jasky for NASA), and in work also done at Pratt and Whitney-Government Products Division, started using the hyperbolic sine

- $$\frac{da}{dN} = \frac{C_2 (\Delta K)^n}{(1 - R) K_c - \Delta K}$$ [Foreman, Shape, Location, and Upper Instability Parameter (K_c) Introducted]

- $$\frac{da}{dN} = f(\Delta K,\ t,\ w,\ \text{Spectrum, environment, microstructure})$$

  t = time
  w = frequency

CORRELATION AND PREDICTION WITHIN BOUNDS OF KNOWN APPLICABILITY, EXTRAPOLATION TO CONDITIONS OUTSIDE BOUNDS INVOLVES UNKNOWN RISK.

- **HYPERBOLIC SINE EQUATION (SINH) (TRANSCENDENTAL FUNCTIONS)**

$$\text{LOG}\ (da/dN) = C_3\ \text{SINH}\,[C_4\ (\text{LOG}\ \Delta K + C_4)] + C_5$$

Battelle Memorial Institute
Hoeppner, Federsen, Jaske, et. al.
(1968), (1973)
Pratt and Whitney - GPD
(1976)
Model (?)

INTRODUCED SIGMOIDAL SHAPE
INTERPOLATIVE

- **MODIFIED SIGMOIDAL EQUATION**

e.g. $$\frac{da}{dN} = \exp.\,(B)\left(\frac{\Delta K}{\Delta K^*}\right)^P \left(\ln\left(\frac{\Delta K}{\Delta K^*}\right)\right)^Q \left(\ln\left(\frac{\Delta K_c}{\Delta K}\right)\right)^D$$

G.E.
Other forms of
above
Duggan (Portsmouth)
Others

- **WEIBULL EQUATION**

$$\Delta K/Kb = [1 - \exp[-[(da/dN - e)/(v - e)]^\wedge k]]$$

Hoeppner, et. al.
(1971 - current)

Fig . 10: Various Fitting Functions for Use in Representing Fatigue Crack Propagation Data.

equation to represent fatigue crack propagation data. During the 1970's various investigators, such as those at General Electric in Evandale and Terry Dougan at Portsmouth Polytechnic Institute proposed a modified sigmoidal type equation as indicated in Figure 10 as well. In 1971, I suggested the utilization of the Weibull equation as also indicated in the last equation in Figure 10, and various investigators, including myself, have been continuing the study of the Weibull equation since 1971. The importance is that we attempt to find fitting functions that adequately describe the full range of behavior of fatigue crack propagation data in the event we want to use these equations to help us in life estimation procedures, and, in comparing materials and manufacturing processes, and in evaluating effects (e.g. environment etc.).

Today, with the advent of personal computers and computing machines that are more readily available to people, it is not as important that fitting functions be obtained for all cases because we can actually take the raw fatigue crack propagation data versus ΔK data and use that data from the raw form in the computer to perform a step-wise integration on the data itself. Nonetheless, there are many cases where we still are required to use fitting functions, and the assessment of the appropriate fitting function becomes an important part of damage tolerant design procedure. To date, various companies and governmental agencies have different approaches that they use with respect to this problem. It is extremely tempting to continue the utilization of the Paris equation for many cases, but nonetheless, I suggest that we try to get away from this and move toward the sigmoidal forms in virtually all cases to improve reliability.

It is extremely important at this point that we recognize the microstructurally dependent phase of crack growth and in the crack formation phase are, without a doubt, two of the most extensive phases of fatigue behavior. They are the two phases that we need to know a great deal more about in order to adequately represent them to develop accurate life prediction procedures. Much of the work within the AGARD community on short cracks and work outside of AGARD on "short cracks" has been helpful in this regard. But, nonetheless, we still need more rationally formulated and probabilistic models to deal in the range of behavior where cracks are significantly affected by the microstructural influences. There has been a great deal of work that has gone on in this field but we need to accelerate this work in order that our fatigue design procedures in this regime of behavior become even more accurate. At this point I would like to quote what Dr. Wood had mentioned on numerous occasions and one of my favorites is in a quote from his work in 1974 which follows:

> "What the cyclic stress does do in practice depends so much on experimental conditions that it is not easy to distinguish general principles. Nor is it easy to resist the temptation of drawing general conclusions from observations that may hold only for limited experimental conditions. For example, much has been made by some authors, including this one, of fatigue cracks that originate in slip bands; and by others of cracks that grow according to "Stage I" and "Stage II" processes. But not all cracks originate or grow in those ways. Evidently a much needed first step in the study of fatigue, one that would clear away contradictions of the kind just noted, is to find why merely the way in which fatigue cracks form should vary. Then it might be possible to recognize specific types of fatigue, and thus to deal with the subject consistently." (6).

The important point to be noted here is that fatigue behavior takes on different forms and the form that the fatigue formation and propagation phases take on are related very much to the material. We must continue to explore ways in which we can incorporate the strong role of the material influence in the early stage of crack formation and structurally dependent crack propagation. Some work is underway in this area but much remains to be done. The frustration that occurs in this regime is best indicated by reviewing another quote from one of the great metallographers in our history, as follows:

> "Also, having spent many years seeking quantitative formulations of the structure of metals and trying to understand the ways in which the structures change with composition and with treatment, and the ways in which structure relates to useful properties, I have slowly come to realize that the analytical quantitative approach that I had been taught to regard as the only respectable one for a scientist is insufficient." (7).

Conclusions

Implementation of damage tolerance to engine structural components requires:

1. An improved data base on engine materials for the following conditions:

   - Realistic mechanical load and temperature spectra (hot turbistan, etc.).

   - Realistic micro-structure.

2. Development of "standard" procedures of evaluation of microstructural influences.

3. Characterization of discontinuities (D) and heterogeneities (H) that will propagate under engine loading conditions. More complete characterization of materials and manufacturing processes.

4. Evaluation of the influence of fretting, corrosion, temperature, and wear (other extraneous effects) as processes leading to generation of crack-like D's and H's under engine operating conditions.

5. Acceleration of fundamental research to improve understanding of metallic, ceramic, and other engine materials under realistic engine load and environmental spectra to improve predictive models.

6. Physically and rationally based models for damage growth prediction.

7. Extensive research to understand temperature and dwell effects in damage growth.

8. Formulation of "damage mechanics" models that incorporate physical concepts (thermodynamics may offer more here).

9. Additional evaluation and understanding of environmental effects.

10. Incorporation of probabilistic concepts into model development.

11. Further evaluation of damage growth data generation procedures (techniques, errors, etc.).

ACKNOWLEDGMENT

I would like to express my appreciation to Mr. Robert Jeal and other Rolls Royce personnel for continued encouragement and inspiration in relation to pursuing various issues connected with damage tolerance of materials. As well, I would like to extend many thanks to Dr. John DeLuccia and Dr. William Wallace for continued encouragement over the years related to numerous research and engineering endeavors and also for stimulating interaction within the AGARD community. I would also like to extend my thanks to the many persons within the AGARD community who continue to direct their wholehearted attention to the questions of damage tolerance applied to engine structures as well as other airframe structures.

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LE BESOIN D'ACTIONS CONCERTEES AU SEIN DE L'AGARD

par
Th. Pardessus
Ingénieur Principal de l'Armement
Chef du Groupe Matériaux et Technologies
C E A T
23, avenue Henri Guillaumet
31056 - TOULOUSE CEDEX, FRANCE

RESUME

En s'appuyant sur le schéma général (fig. 1) de la boucle d'optimisation des matériaux (spécification, essais, modélisation du comportement, "résistance utilisable", gestion de la durée de vie), cet article passe en revue un certain nombre d'aspects de l'influence des défauts sur le comportement mécanique.

Les domaines d'investigation concernés sont variés (procédés, mécanique, physique, modélisation et traitement de données), et on essaye pour chaque rubrique de dresser le bilan des actions de standardisation en cours et de tracer les axes d'effort.

1 - QU'EST CE QU'UN DEFAUT ?

A y bien réfléchir, la notion de "défaut" est très vague. Si dans un premier temps, on ne considère que le critère dimension, on peut situer un "défaut" sur un segment dont les extrémités sont respectivement représentatives des défauts cristallins et de la fissure visible à l'oeil nu (fig. 2). Entre les deux, le seuil de détectabilité par les techniques non destructives est un critère dont l'importance aux plans industriel et opérationnel n'est plus à démontrer.

Si maintenant on s'intéresse à la nature de ces défauts, on pourra trouver :

- des hétérogénéités structurales (taille de grains, amas de phases précipitées, défauts cristallins, .. .)
- des discontinuités (fissures, porosités, soufflures, . . .)
- des inclusions exogènes que l'on saurait différencier du point de vue chimique (inclusions réactives, non réactives) ou mécanique (inclusions "dures").

Vers une définition synthétique : on pourrait appeler "défaut" la zone très ponctuelle d'une éprouvette ou d'une pièce dont on aura pu démontrer qu'elle a été ou qu'elle aurait pu être à l'origine d'un processus mécanique local de dégradation voire de ruine. Une formulation plus rigoureuse sous-entendrait une description de la structure des pièces en termes mathématiques, qui ferait intervenir une norme et une statistique des écarts à la spécification. Les écarts admissibles devraient tenir compte de la nocivité, pour autant que l'on puisse l'évaluer.

Il serait donc intéressant de procéder à l'établissement d'une classification standard des types de défauts et de passer au crible les résultats d'essais les plus récents ; les trois techniques principales d'élaboration de pièces critiques (métallurgie du lingot, métallurgie des poudres, fonderie sous vide) méritent d'être étudiées.

Par nécessité, de telles classifications existent mais sont le fruit d'initiatives individuelles, il y aurait sans doute à gagner en confrontant ces travaux parallèles pour définir un standard.

2 - METHODES D'ESSAIS MECANIQUES.

2.1. Essais de base.

On ne saurait trop rappeler l'importance de pouvoir disposer de procédures d'essais validées. Si la standardisation rigoureuse de ces procédures peut paraître un objectif séduisant, elle se heurte cependant à des difficultés d'ordre pratique ; on s'attachera donc à réaliser les essais suivant des procédures au moins comparables. Ceci suppose qu'un certain nombre d'essais croisés aient été réalisés et que l'on définisse une "instance de décision", à caractère statistique.

Au sein de l'AGARD, le programme "Engine disc material" (TX 114) a montré, lors de sa phase initiale, un bon accord entre laboratoires pour 4 essais différents, mais tous du type force imposée et à l'ambiante.

Le programme "High temperature cyclic behaviour" (TX 134) se propose, lui, d'aborder l'inter-comparabilité d'essais réalisés à une température typique de fonctionnement des disques et d'inclure les essais à déformation imposée.

La comparaison de résultats d'essais de fissuration met en évidence la nécessité. de procéder à des traitements numériques standard, ou du moins d'expliciter au maximum les formules ou algorithmes employés.

2.2. Séquences standard de fatigue oligocyclique.

S'inspirant des travaux réalisés au milieu des années 70 par des groupes tels que FALSTAFF et TWIST, TURBISTAN a défini deux séquences de base, représentatives des charges de fatigue oligocyclique subies par les disques : l'une pour les pièces à faibles gradients thermiques (COLD TURBISTAN), l'autre au contraire, pour les pièces où les effets de la température et du temps sont significatifs (HOT TURBISTAN).

En ce qui concerne les matériaux pour aubes de turbine la question reste ouverte.

La caractérisation des matériaux pour disques sous chargement réaliste est nécessaire si l'on veut appréhender le plus tôt possible les effets de séquence et travailler parallèlement à la modélisation (cf. § 3). On pourra, pour plus d'information, se référer à la littérature du groupe TURBISTAN.

2.3. Plans standard de caractérisation mécanique.

Face aux coûts des caractérisations et qualifications de matériaux pour disques de turbomachines, il importe de tenter de concentrer ses propres forces : une solution possible est de constituer des plans standard de caractérisation. Fédérer au sein d'un groupe de travail les différentes tentatives que l'on peut recenser dans différents pays, pourrait être un rôle de l'AGARD. Bien sûr, cette proposition d'action n'a fondamentalement de sens que si, au préalable, ou au moins en parallèle, on conduit assidûment l'effort sur les méthodes d'essais standard. Plus généralement, c'est à la question : "quelle est le plan de caractérisation optimal des matériaux pour disques ," qu'il faut s'efforcer de répondre. De nouvelles voies, comme les essais de fatigue à déformation imposée sur éprouvettes pré-endommagées, méritent certainement d'être explorées.

3 - LE ROLE DE LA MODELISATION.

La modélisation des comportements mécaniques est un souci permanent, à divers degrés, du Bureau de Calcul : le contact avec le Laboratoire Matériaux devient alors très interactif. Si les efforts ont porté essentiellement sur la prévision de l'amorçage, de la fissuration, en fatigue, et du comportement cyclique, c'est sans doute ce dernier volet qui présente le plus de maturité. La prévision de l'amorçage s'expose à une difficulté supplémentaire, qui est que seul, à la limite, un modèle capable de prendre en compte certaines données locales (c'est-à-dire rôle de la microstructure locale) peut prétendre à une description correcte des phénomènes. Il n'en ressort pas moins que la littérature révèle l'existence de différents modèles

La modélisation de la propagation de fissure dans les matériaux moteur (disques surtout) devrait bénéficier des développements récents sur le comportement des alliages pour cellule d'avion.

Un autre aspect, beaucoup moins souvent abordé, est celui de la modélisation de la réponse des "défauts" aux différentes techniques d'analyse non destructive. Pour l'instant, la démarche s'appuie davantage et c'est logique, sur une pratique coordonnée et rationalisée. L'époque des systèmes experts de reconnaissance de défauts est déjà partiellement ouverte et l'apport de la modélisation peut s'avérer déterminant, en tant qu'elle facilitera les phases d'apprentissage et d'identification.

Dans le cadre d'une caractérisation graduée, de l'éprouvette élémentaire au composant réel, en passant par les disques-éprouvettes, le rôle de la modélisation est important : chaque niveau permet de valider les prévisions établies par application du modèle au niveau précédent (fig. 3).

N'omettons pas de souligner l'importance des traitements statistiques, notamment ceux qui traitent de la statistique des coefficients des modèles : sensibilité aux conditions d'identification, répercussion sur les calculs en conditions réalistes, . . . La forme mathématique même de ces modèles complique sérieusement la question.

4 - LA DUREE DE VIE.

Pour passer à un stade industriel, il est indispensable de maîtriser le difficile problème de la relation microstructure/propriétés mécaniques, du moins dans son acception classique, qui traite de microstructures moyennes et de propriétés moyennes.

Cela n'exclut pas, bien au contraire, l'approfondissement de la micromécanique des défauts qui peut d'ailleurs englober l'étude de conditions locales anistropes. Une approche statistique de type "ingénieur" est bien sûr nécessaire.

Les effets de l'environnement (par exemple, l'amorçage sur inclusion est évidemment différent suivant que celle-ci est interne ou en surface) doivent être étudiés : la définition de la matrice d'essai optimale doit donc inclure cet aspect.

Mais la complexité de la prévision des durées de vie des disques, qui tient non seulement à une mécanique intrinsèquement difficile, mais aussi au traitement de statistiques de défauts dans les très faibles probabilités, doit faire réfléchir à la limite des concepts qui sous-tendent caractérisation de matériaux et prévision.

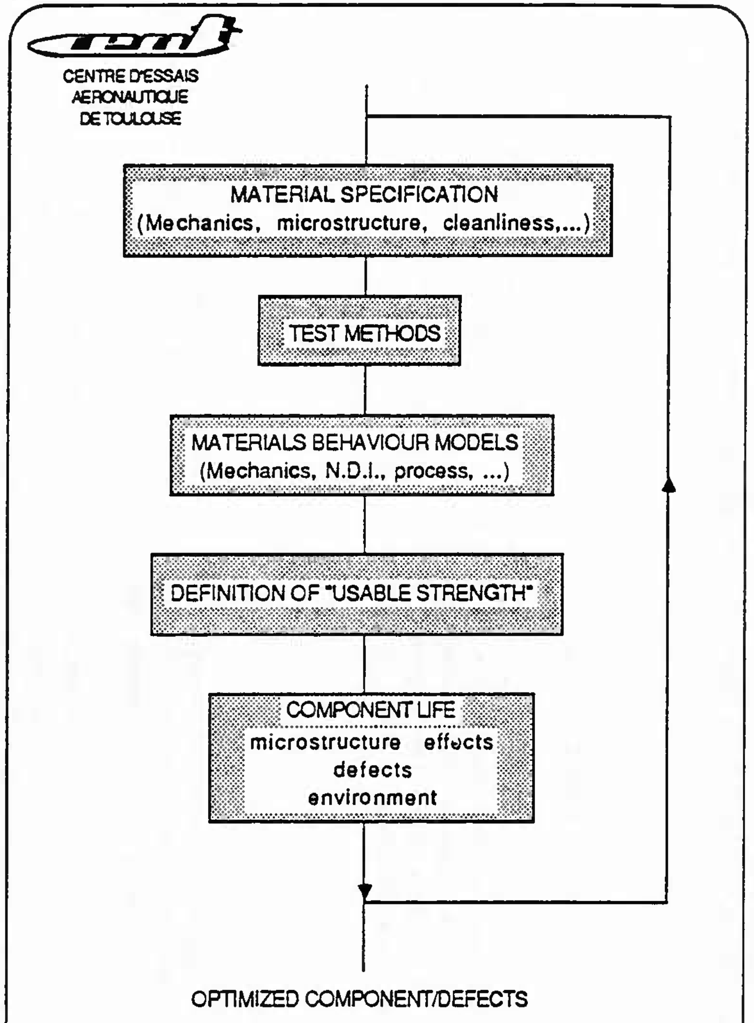
5 - VERS UNE NOTION DE "RESISTANCE UTILISABLE"

Il est maintenant clair, à la lumière de l'expérience et de ce qui a été exposé précédemment, que l'expression du potentiel d'un matériau n'a de sens que si l'on sait lui attacher le traitement des défauts (dans le cadre de gammes d'élaboration/transformation précises) et la géométrie (plus exactement la mécanique) de la pièce finale. Les propriétés statiques sont certainement moins sensibles que la tenue en fatigue, pour laquelle l'approche proposée est capitale.

6 - CONCLUSION.

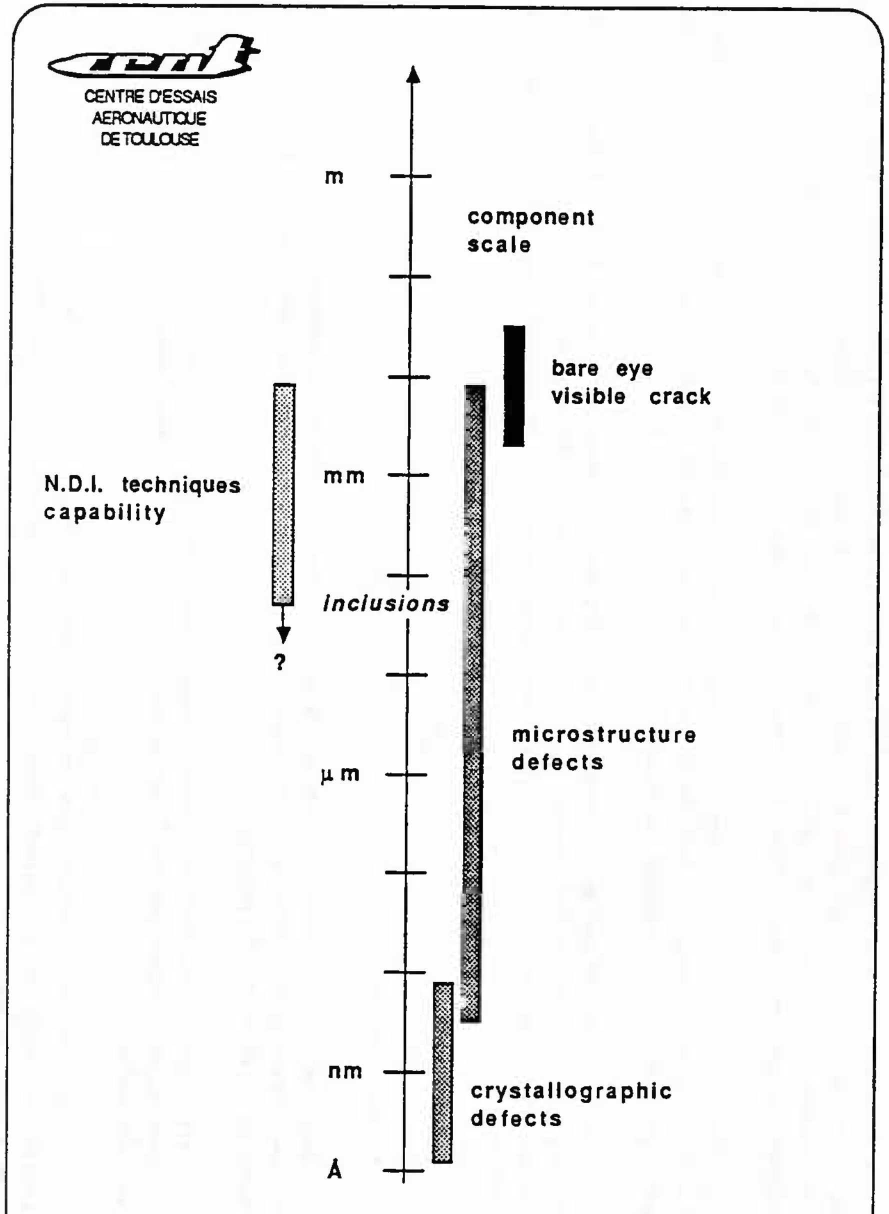
Tout au long de cet article, on a essayé de faire ressortir les problèmes concrets : un certain nombre d'entre eux ont déjà trouvé dans le cadre de l'AGARD le moyen d'être abordés en commun. D'autres pourraient faire l'objet de travaux coopératifs enrichissants.

Quoi qu'il en soit, l'étendue des questions esquissées dans ce papier montre à l'évidence la nécessité d'unir les forces disponibles et de se livrer à des réflexions de fond sur la caractérisation des défauts et de leur influence sur les propriétés finales des pièces critiques, voire sur leur conception.



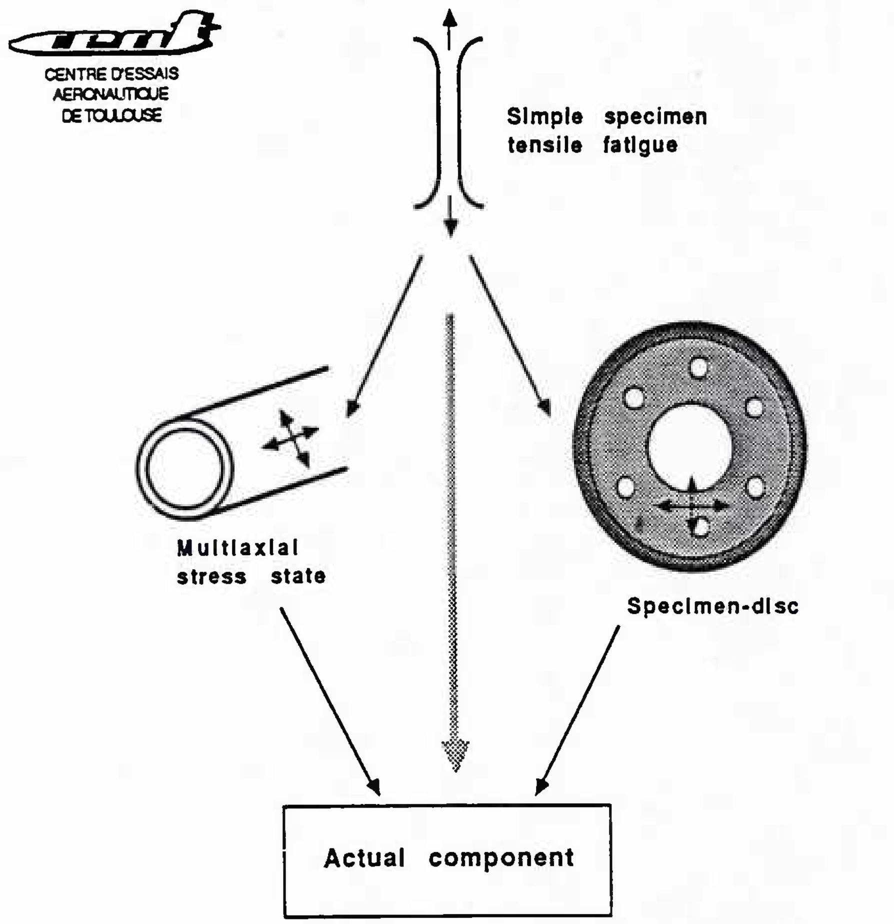
A GENERAL "CLOSED LOOP" APPROACH

Fig. 1



SCALE OF DEFECTS

Fig. 2



A STEP BY STEP APPROACH OF MATERIALS BEHAVIOUR

AND COMPONENT LIFE PREDICTION

Fig. 3

RECORDERS REPORT
DAMAGE TOLERANCE CONCEPTS FOR ENGINE CONSTITUENTS WORKSHOP II: DEFECTS AND QUANTITATIVE MATERIALS BEHAVIOUR

Mr M J Weaver
Mr C R Gostelow

Propulsion Department
Royal Aerospace Establishment
Pyestock, Farnborough
Hants, GU14 0LS, UK

1 Introduction - Jeal

In opening the workshop Jeal noted that five years ago a meeting on defects would have attracted a small number of metallurgists. Today the significance of defects are appreciated equally by the metallurgist, the designer and the lifing engineer.

Beevers reminded the meeting that simple assumptions concerning crack growth could be misleading, citing the order of magnitude difference in the growth rates between large and small grained titanium alloys. This he said could be attributed to crack closure. Labourdette mentioned the difficulty of knowing, or measuring, residual stress at the microstructural level and suggested this could be a possible subject for future research. Jeal concurred that a knowledge of the stress at the crack tip was most important, and that this would require the development of predictive tools and techniques.

2 Structure of Engineering Materials - Bachelet

Bachelet's paper emphasised the role of microstructure and process control in the development of modern damage tolerant materials, citing titanium IMI 834 and nickel N18 as prime examples. Both alloy chemistry and microstructure effect crack propagation. He suggested that to optimise properties the grain size would need to vary across the disc.

In discussion Hoeppner queried the validity of using the mode I stress intensity factor, particularly for microstructurally small cracks and cracks growing on other than the plane normal to the applied stress. This was particularly relevant for single crystals. Although this concern found widespread acceptance, the problem of reaching agreement on a viable alternative was felt to be insurmountable at the current state of knowledge. Indeed, there was not even an agreed definition of crack length for a crack growing with a varying orientation to the stress axis. Jeal reminded the meeting that, for the high strength nickel discs at least, cracks of engineering significance were still likely to be large compared to grain size, and K_1 still appeared to be an acceptable approach.

3 Component Defects - Pickard

Pickard explained the background to the various component lifing methods used by Rolls-Royce, paying particular attention to fracture mechanics based schemes. Both sub-surface and surface defects (scratches and indentations) were considered.

A number of questions were raised concerning the analysis of surface defects, and the need to understand when these behaved as cracks from the first cycle. In reply to a question by Hoeppner, Pickard said that he modelled cracks in multiaxial stress fields using the stress normal to the plane of the crack, and that this gave good results. Although in his paper Pickard claimed he was able to model a small crack or indentation, his analysis did not allow for any local residual compressive stress put into the component as a result of an impact type indentation. The need for more work to understand and model the damage effect of surface blemishes was highlighted.

4 Component Defects - Fournier

Fournier described the results of an extensive study of cast nickel blisks. He emphasised the need for the best possible process control, supported by reliable NDE. Over 5000 blisks had been tested and 250 cut up in evaluating the manufacturing route.

When questioned Fournier confirmed that all failures had been from surface or near surface origins and that sub surface defects did not appear to be a problem. This he explained as being due to good process control and a design that took account of residual defects. A further point raised concerned the cleanliness of the cast nickel superalloy MM004 used in the blisk manufacture. From the number of defects cited the material was of average to good cleanliness and of a level similar to that found in wrought material. In reply to a question by Tadros, Fournier said that whilst defects in forged blisks were physically smaller, they appeared more harmful than in the cast material.

5 Defect Behaviour and Component Design - Frediani

The paper described a three dimentional J Integral approach based on an energy parameter GJ. Frediani said this modelled the behaviour of growing cracks more precisely than the normal K approach. In particular it could predict the crack front geometries observed by Rolls-Royce in their corner crack and CT specimens.

In response to questioning, Frediani said that the Finite Element mesh was recomputed to position the Gauss points onto the crack front. Pickard noted the approach involved an elastic-plastic analysis and that any result would be specific for a given material, geometry and loading condition. He was reasonably happy that, by using corner crack specimen data to predict the behaviour of corner cracked components, any errors in the shape function would be cancelled out and the simpler K approach used. The art was to compare like with like.

6 Modelling Methods - Tadros

Tadros summarised the Pratt and Whitney (Canada) approach to damage tolerance lifing. The approach is very similar to that being adopted by Rolls-Royce in that a pseudo crack starter size is determined by back calculation using fracture mechanics and striation counting, but the end point of the crack size calculation is 1/32 inch (0.8 mm) rather than burst.

In answer to a question on the validity of striation counting, Tadros said this was acceptable for most titanium alloys and for Waspaloy, but less successful when applied to IN718. Jeal made the point that whereas using 1/32 inch as the failure criterion was satisfactory for older, less strong alloys, this may not be conservative for the more modern high strength materials. For such materials a better and safer approach was to use a fixed percentage of burst in determining component life.

7 Future Technology Requirements - Hoeppner

Hoeppner gave a detailed presentation on the pitfalls and problems of damage tolerance lifing. In particular he emphasised the need for a greater understanding of the physical principles behind the damage tolerance approach.

Gostelow questioned the use of the Sinh Function approach, arguing that a series of straight line fits through the data gave a better description of the competing physical processes involved in crack growth. Whilst agreeing in principle, Hoeppner said that, using the Sinh Function, data was easier to store and use in computer predictions. It was nevertheless important to realise that extrapolation beyond the limit of the actual test data involved a high degree of risk.

8 Need for Common AGARD Approach - Pardessus

In a wide ranging presentation Pardessus suggested that within AGARD there was a need to agree on what constituted a fully integrated material specification for damage tolerance. A novel idea put forward in the paper was the need to create a new test procedure that would help define more precisely defect sizes and defect populations.

In discussion Labourdette again raised the problem of biaxial and triaxial testing. Although it is possible to undertake biaxial testing it is still difficult to undertake a laboratory 3D test. He questioned if such tests were necessary. Pickard said that full scale component testing will give 3D stresses, but accepted that these would be difficult to quantify in absolute terms.

Hoeppner questioned what was meant by modelling and if models should be correlative or predictive in nature. Labourdette replied that models must be correlative, as to become predictive would involve going back to very basic physics. Bachelet said physical mechanisms should however be studied to help define the validity of the various models. It was agreed that an eventual aim must be to develop predictive models, and that this would require a major effort through AGARD and other collaborative exercises.

The need to agree standard techniques for testing, data analysis and data presentation was again highlighted. This was identified as an area where AGARD SMP could make a major contribution through existing and possible future programmes.

CONCLUSIONS

Jeal said that for the first time in his recollection a meeting on damage tolerance had been held when there had been no dissenting voices from the view that defects are important. He then summarised the four main points that had arisen out of the meeting:

1. The closed loop design/manufacturing process was essential for a damage tolerance approach.
2. That compatible test methods/data collection/presentation systems should be developed between Nations.
3. There was still much to be learnt about the nature/origin of defects.
4. There were still many unanswered questions regarding the best analytical methods/models to be used in describing crack growth behaviour.

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a. The move towards designs allowing higher operational stresses, and the use of advanced high-strength alloys make it likely that a disc burst could happen (following a rapid crack growth) well before the

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b. It is potentially wasteful of expensive components, since it has been estimated that over 80% of engine discs have ten or more cycle fatigue lives remaining when discarded under "safe life" rules.

Damage tolerance being an alternative lifeing philosoply, the Sub-Committe on "Damage Tolerance Concepts for the Design of Engine Constituents" has therefore decided to conduct a series of four Workshops addressing the areas critical to Damage Tolerant design of engine parts.

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