# MECHANICS AND PHYSICS OF CRACK GROWTH: APPLICATION TO LIFE PREDICTION

Editors: R.B. Thompson, R.O. Ritchie
J.L. Bassani and R.H. Jones

# MECHANICS AND PHYSICS OF CRACK GROWTH

## APPLICATION TO LIFE PREDICTION

*Proceedings of a Workshop sponsored by the Division of Materials Science, Office of Basic Energy Sciences, U.S. Department of Energy and held in Keystone, CO, August 4–7 1987*

EDITORS: R. B. THOMPSON, R. O. RITCHIE, J. L. BASSANI and R. H. JONES

Reprinted from the journal
*Materials Science and Engineering*
Vol. A103, No. 1

ELSEVIER APPLIED SCIENCE
LONDON and NEW YORK

ELSEVIER APPLIED SCIENCE PUBLISHERS LTD
Crown House, Linton Road, Barking, Essex IG11 8JU, England

*Sole Distributor in the USA and Canada*
ELSEVIER SCIENCE PUBLISHING CO., INC.
52 Vanderbilt Avenue, New York, NY 10017, USA

**British Library Cataloguing in Publication Data**

Mechanics and physics of crack growth.
1. Materials. Fracture. Mechanics
I. Thompson, R. B. (Ronald Bolton)
620. 1′ 126
ISBN 1-85166-978-7

**Library of Congress CIP Data Applied For**

ISBN 1-85166-978-7



Printed in The Netherlands by Krips Repro B.V.

# MECHANICS AND PHYSICS OF CRACK GROWTH
## APPLICATION TO LIFE PREDICTION

# Preface

Attempts to predict the life of structural components are generally based on empirical laws which are not supported by a complete understanding of the underlying physical mechanisms. Consequently, the extrapolation of these engineering relations to new situations cannot be done with confidence. Moreover, deficiencies in the link between failure mechanisms and life prediction impede the use of physical measurements to assess the expended or the remaining life or both. The papers in this issue describe the results of a workshop held to discuss the present status of life prediction and to define those areas of basic materials research needed to form a foundation for future improvements.

This is the second of two meetings directed at failure phenomena under the sponsorship of the Division of Materials Science, Office of Basic Energy Sciences, U.S. Department of Energy. The first was the panel, *Micromechanics and Physics of Fracture*, organized by the *ad hoc* Materials Science Council and chaired by V. Vitek. It was held on July 29–August 2, 1985, in Bretton Woods, NH, U.S.A. Attention focused on the microscopic processes controlling the fracture behavior of materials rather than on the macroscopic aspects of fracture. It was noted that recent advances in atomistic calculations of the propensity to fracture are significantly advancing understanding of the strength of lattices and interfaces, and that these studies must be closely coupled with experimental studies of interfacial properties, dislocation phenomena and crystal-defect behavior in model systems. Micromechanics was identified as an emerging tool in coupling the understanding of these atomic-level processes with microscopic fracture phenomenon. It was also noted that a number of non-traditional experimental techniques (*e.g.* using acoustics, synchrotron radiation and neutron energy) have reached a level of sophistication that allows them to be applied successfully in the study of the micromechanisms of fracture.

As a follow-up, the workshop, *Mechanics and Physics of Crack Growth: Application to Life Prediction*, was held on August 4–7, 1987 in Keystone, CO, U.S.A. It complemented the Bretton Woods panel by emphasizing the link between the microscopic aspects of materials failure and macroscopic structural performance. In particular, it adopted as a goal the definition of those areas of basic materials research needed to form the foundation for a reliable technology of life prediction of structural components. Consideration was given both to identifying the scientific elements that were needed to form this foundation, and to projecting the areas of technology from which the next major advances would be expected to emerge.

Even in areas where various pieces of the puzzle are reasonably well understood, it was noted that efforts to integrate concepts from the microscopic to the macroscopic scale are still needed. To focus the discussions, three processes of materials failure spanning a range of technologies were selected. These processes share the common features of (a) widely used, empirical life prediction rules, (b) partially developed but incomplete models for the evolution of damage and (c) candidate experimental methods for the non-destructive assessment of the damage state. The following three chapters present results of the discussion on fatigue, high temperature creep and environmentally assisted crack growth. In each chapter a brief introduction is followed by a series of technical papers describing the present state of damage models, life prediction methodologies and advanced measurement techniques. The final chapter presents recommendations for the direction of future research in each of the areas based on the detailed discussions that were held at the workshop.

Participants in the workshop included representatives from the academic, national laboratory, industrial and government communities.

R. B. THOMPSON
*Chairman*
*May 1988*

---

## Organizing Committee


| | |
|---|---|
| Professor J. L. Bassani<br>High Temperature Creep | Mechanical Engineering Department<br>University of Pennsylvania<br>111A Towne Boulevard/D3<br>Philadelphia, PA 19104, U.S.A. |
| Dr. R. H. Jones<br>Environmentally Assisted Crack Growth | Battelle-Pacific Northwest Laboratories<br>P.O. Box 999<br>Richland, WA 99352, U.S.A. |
| Professor R. O. Ritchie<br>Fatigue | Lawrence Berkeley Laboratory and Mineral Engineering Department<br>University of California<br>Berkeley, CA 94720, U.S.A. |
| Professor R. B. Thompson<br>Chairman | Ames Laboratory and Materials Science and Engineering Department<br>Iowa State University<br>126 Met. Dev. Bldg.<br>Ames, IA 50011, U.S.A. |


## Advisors


Dr. R. W. Heckel
ER-131, GTN
U.S. Department of Energy
Washington, DC 20545, U.S.A.

Dr. F. V. Nolfi, Jr.
ER-131, GTN
U.S. Department of Energy
Washington, DC 20545, U.S.A.

## Participants

Dr. P. L. Andresen
General Electric Research and Development
Building K-1, Rm. 3443
P.O. Box 8
Schenectady, NY 12301, U.S.A.

Professor O. Buck
205 Metals Development
Ames Laboratory
Ames, IA 50011, U.S.A.

Dr. B. A. Cowles
Pratt & Whitney
Engineering Division, MS 707-20
P.O. Box 109600
West Palm Beach, FL 33410-9600, U.S.A.

Dr. B. F. Dyson
National Physical Laboratories
D.M.A.
Teddington, Middx. TW11 0LW, U.K.

Professor J. P. Gallagher
School of Engineering
University of Dayton
300 College Park Drive
Dayton, OH 45469, U.S.A.

Professor R. P. Gangloff
Department of Materials Science
University of Virginia
Charlottesville, VA 22901, U.S.A.

Professor W. W. Gerberich
151 Chemical Engineering
University of Minnesota
Minneapolis, MN 55455, U.S.A.

Dr. L. Hagn
Allianz Zentrum für Technik G.M.b.H.
Krausstrasse 14
8045 Ismaning
Munchen, F.R.G.

Dr. M. A. Hicks
Rolls-Royce Ltd.
P.O. Box 31
Derby DE2 8BJ, U.K.

Professor F. Leckie
Department of Mechanical Engineering
University of California
Santa Barbara, CA 93106, U.S.A.

Dr. A. K. Miller
Department of Materials Science and Engineering
Stanford University
Stanford, CA 94305, U.S.A.

Professor W. D. Nix
Department of Materials Science and Engineering
Stanford University
Stanford, CA 94305, U.S.A.

Professor R. N. Parkins
Department of Metallurgy and Engineering Materials
University of Newcastle Upon Tyne
Newcastle Upon Tyne NE1 7RU, U.K.

Professor R. M. N. Pelloux
Department of Materials Science and Engineering
Massachusetts Institute of Technology
Cambridge, MA 02139, U.S.A.

Professor D. P. Pope
109 Towne Building
Department of Materials Science and Engineering
University of Pennsylvania
Philadelphia, PA 19104, U.S.A.

Dr. E. N. Pugh
Room B266, Building 223
National Bureau of Standards
Gaithersburg, MD 20899, U.S.A.

Professor A. Saxena
Georgia Institute of Technology
Department of Metallurgy
Bunger Henry Building
Atlanta, GA 30332, U.S.A.

Professor D. Socie
Department of Mechanical and Industrial Engineering
University of Illinois at Urbana-Champaign
Urbana, IL 61801, U.S.A.

Professor R. H. Van Stone
A333 — Mail Stop
General Electric Company
1 Neumann Way
Cincinnati, OH 45215, U.S.A.

Dr. R. Viswanathan
Electric Power Research Institute
3412 Hillview Avenue
Palo Alto, CA 94303, U.S.A.

# Contents

*Materials Science and Engineering, A103* (1988) 1–2



# FATIGUE

## Introduction

The papers presented in the first session of the workshop focused on the question of fatigue failure and specifically on how fundamental studies of the mechanics and mechanisms of cyclic deformation and of crack initiation and growth have been incorporated in modern life prediction procedures. Accordingly, the scope of the following eight papers is extremely broad, covering topics as wide ranging as the mechanistic modeling of fatigue crack growth and crack-tip shielding mechanisms, the quantitative measurement of such mechanisms for non-destructive testing (NDT) purposes, the respective life prediction methodologies principally based on crack initiation (stress and/or strain–life or $S$–$N$ approach) but also on crack growth (defect-tolerant approach), and the application of these approaches specifically to the aircraft, automotive and especially gas turbine industries.

The consensus of these papers is that significant progress has been made in recent years in the understanding of the salient mechanisms affecting both fatigue crack initiation and growth in metallic materials, although precise mechanisms describing exactly how a fatigue crack advances are still uncertain. Such information has been extremely useful in understanding what microstructural features contribute to superior fatigue resistance and therefore to the alloy design of materials with improved crack-initiation or crack-growth properties. However, with rare exception, such mechanistic understanding has not been incorporated in the most widely used life prediction models which still rely principally on empirical concepts. An exception to this is the approach of Miller and coworkers at Stanford who have tried to incorporate the relevant physics of fatigue and creep mechanisms into their life-prediction models through the use of numerous adjustable parameters, although their analyses have not to date been utilized in industry.

With regard to the choice of life prediction procedures, there is now a realization that the microstructural factors governing crack initiation in fatigue may be very different from those governing crack growth; accordingly the $S$–$N$ life prediction approaches, which predict the total life (for both crack initiation and growth) are fundamentally different from the fracture-mechanics-based defect-tolerant approaches, which predict the crack-propagation life. Mechanistically, this has been attributed to crack-tip shielding mechanisms, principally involving crack closure, which can have a marked influence on crack growth by locally reducing the crack driving force actually experienced at the crack tip. However, as these mechanisms primarily act on the crack wake, they are largely insignificant in influencing crack initiation processes. Therefore, the $S$–$N$ approach is focused on smaller components where the crack-initiation life is often appreciable, and so this approach has found widespread application in the automotive industry, for example. The approach is still mainly empirical but, owing principally to the efforts of Morrow and coworkers at Illinois, it has been developed in recent years into a sophisticated and consistent package of equations to predict fatigue life under a wide range of conditions, incorporating mean-stress effects, stress concentration, variable-amplitude loading, high temperature and environmental effects, and mixed-mode conditions.

Corresponding damage-tolerant life prediction methodologies are more specific about the details of the crack-propagation process and therefore are based on crack-growth data characterized in terms of fracture mechanics; their application is therefore widespread for safety-critical stituations, such as in the aerospace industries where the risk of failure is far more important. The vital question here is generally the prediction of fatigue life under variable-amplitude loading conditions and, despite progress over the past 10 years in the mechanistic understanding of fatigue behavior under these conditions, many life prediction analyses in current use still rely on the largely empirical Wheeler and Willenborg models.

Another similar problem has been the realization that the crack-growth properties of micro-

0921-5093/88/$3.50

structurally small surface cracks (typically in the range 1–500 $\mu$m) may be very different from those of through-thickness long cracks (1 mm or longer) measured conventionally in standard test specimens. Although attributable to several factors, this again follows primarily from considerations of crack-tip shielding since the small flaw with its limited wake will be less influenced by closure mechanisms. However, the problem has practical significance because defect-tolerant lifetime calculations are invariably based on conventional long-crack data. As the overall life of a component is dominated by low growth-rate behavior where the crack is generally small, the accelerated (and sub-threshold) extension of small flaws can potentially lead to dangerous overpredictions of life. The small-crack problem is perhaps most acute in the life prediction methodologies used for gas-turbine components where, to date, $S$–$N$ (often termed low-cycle fatigue), conventional crack growth and small-crack approaches are all used, as described in the papers by Cowles, Hicks and Van Stone in this issue.

Clearly, areas for future research in fatigue are numerous, both for the fundamental understanding of the phenomenon and for the incorporation of this understanding into usable life prediction methodologies. However, there are two critical areas which stand out. Firstly, the modeling of fatigue mechanisms has been, and continues to be, limited by the fact that precise crack-tip fields existing ahead of a cyclically loaded crack, incorporating cyclic plasticity and non-stationary crack terms, have never been determined. By characterizing fatigue crack growth in terms of the stress intensity range, as is invariably done, we are currently utilizing a parameter based on the superposition of linear-elastic monotonic fields for a stationary cracks, whereas fatigue cracks obviously move and are influenced by cyclic plasticity. Corresponding characterizing parameters to describe the extension of small cracks are similarly unknown. Secondly, whereas most fatigue life prediction procedures are at present centered on metallic components, within the next 20 years the structural use of advanced non-metallic materials will become more and more important. Although the mechanistic understanding of fatigue processes in metallic alloys is reasonably well advanced, corresponding mechanisms associated with the cyclic fatigue of composites, and expecially ceramics, is essentially unknown. This is perhaps our most pressing need, specifically for current major developments such as reliable gas-turbine engines incorporating structurally significant ceramic components and for the design of hyperspace vehicles incorporating fiber-reinforced intermetallic alloys.

R. O. RITCHIE

# Bulk Deformation Fatigue Damage Models*

DARRELL SOCIE and JULIE BANNANTINE

*Department of Mechanical and Industrial Engineering, University of Illinois at Urbana-Champaign, 1206 West Green Street, Urbana, IL 61801 (U.S.A.)*

(Received November 23, 1987)

**Abstract**

*Bulk deformation fatigue damage models assume that the stresses and strains at the critical location of the component or structure govern the fatigue life. They allow engineering estimates of fatigue life to be made without explicitly modeling the detailed mechanisms of the fatigue process. To be successful, however, these models must reflect the physical process.*

*Observations of fatigue cracking behavior show that fatigue damage is dependent upon the loading mode, the strain amplitude and the type of material. The cracking behaviors of AISI stainless steel 304, SAE 1045 steel and Inconel 718, tested in tension and torsion, are used to identify the development of damage in these materials for various strain levels. Damage develops either on planes of maximum shear strain or on planes of maximum tensile strain. Identification of the damage in these materials then allows estimates of the lifespan to be made for complex multiaxial loading using an appropriate bulk deformation model. Multiaxial fatigue tests results, including mean stress effects, are presented and successfully correlated with the model that incorporates the dominant or controlling damage parameter.*

## 1. Introduction

Fatigue life estimates represent one aspect of a durability assessment of components and structures. Landgraf [1] recently summarized the durability assurance technology in the ground vehicle industry, which is depicted graphically in Fig. 1. A balance between testing and analysis is required for reliable designs at both the specimen and the component levels. For example, vehicle modeling is verified by service load measurements and laboratory simulation at the appropriate stage in the design and manufacturing process. The focus of this paper is on one aspect of the technology, the damage models and life prediction models for fatigue life estimates.

Fatigue lives are estimated from information on the material properties (including processing variables), the geometry of the component and the service loading history. A schematic illustration of the approach is given in Fig. 2. Bulk deformation models for fatigue life predictions include both the stress–life and the strain–life approaches, the latter being more appropriate for situations involving plastic strains. These approaches are most often used to assess the influence on fatigue life of changes in service usage, material or local geometry. They are usually employed to obtain the fatigue life to form a crack of approximately 1 mm. Because of this, the approaches are often referred to as crack initiation approaches. However, they may include a substantial portion of microcrack growth as small cracks form and grow in some complicated manner. (Often initiation and microcrack growth consume the majority of the usable life in a component or structure.)

In bulk deformation models, stresses and strains in the critical location of the structure are assumed to govern the fatigue life. Cracks are assumed to nucleate and grow in the structure in the same manner as they nucleate and grow in the smooth laboratory specimen which is used to determine the material properties. These models must reflect the physical damage process to be successful.

## 2. Background

August Wöhler, while not the first, is the most famous early fatigue researcher. He was active in

*Paper presented at the Workshop on the Mechanics and Physics of Crack Growth: Application to Life Prediction, Keystone, CO, U.S.A., August 4–7, 1987.

0921-5093/88/$3.50

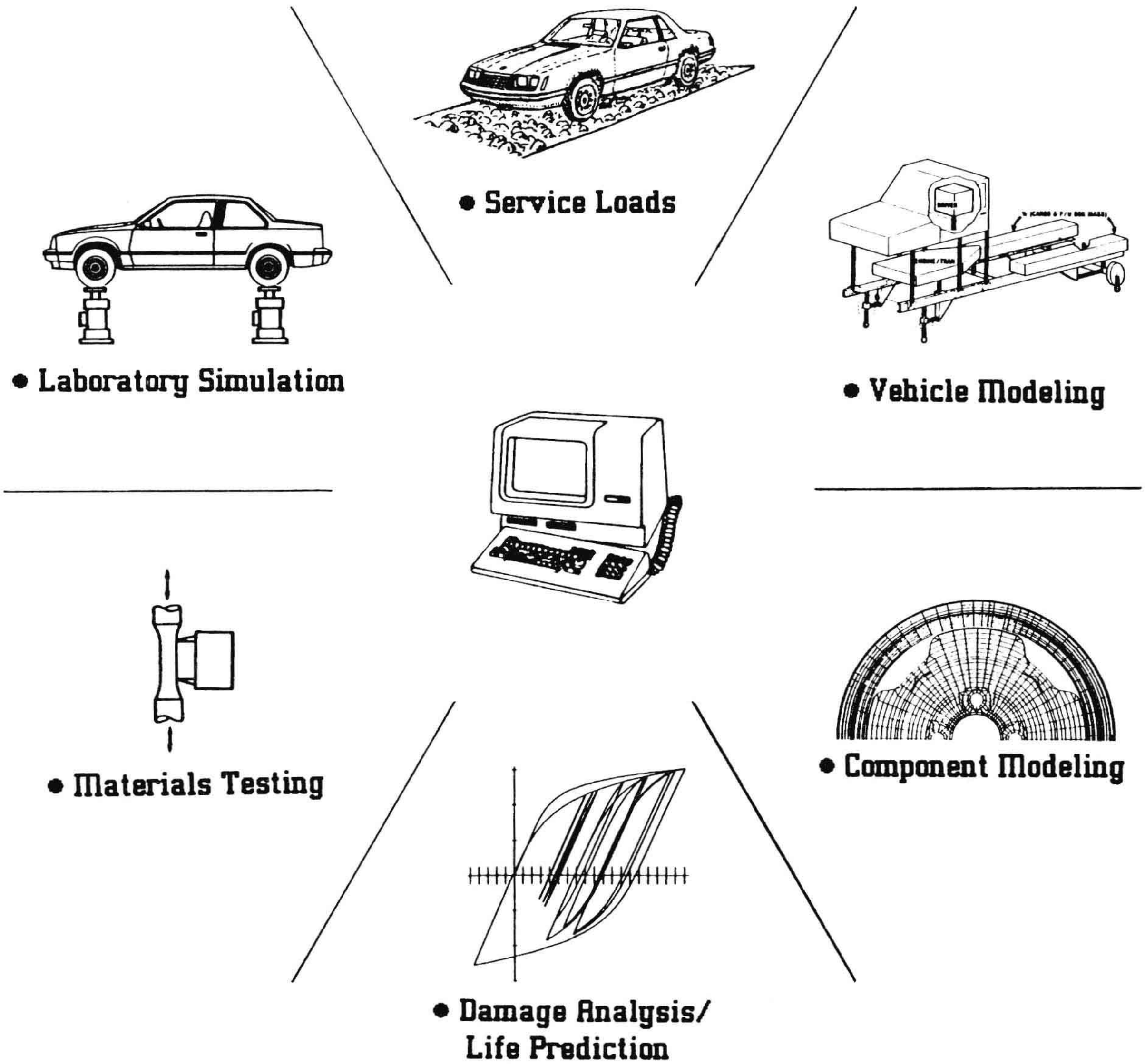



Fig. 1. Durability assurance technology in the ground vehicle industry.

Germany during the period from about 1850 to 1875, working on fatigue problems associated with railroads [2]. He conducted bending, axial and torsion fatigue tests to establish a safe stress below which failure would not occur. He used full-scale axles as well as smaller specimens to simulate axles and established the endurance limit concept for design. Research for nearly 100 years has been performed to establish experimentally the effects of the many variables that influence the long-life fatigue strength.

In 1903, Ewing and Humfrey [3], motivated by the work of Wohler and Bauschinger, published their classic paper, "The fracture of metals under repeated alterations of stress". They tested flat fatigue specimens made from high quality Swedish iron in the annealed condition and microscopically examined the same region of the specimen at various stages in the fatigue life. They stated, "The course of the breakdown was as follows. The first examination, made after a few reversals of the stress, showed slip-lines on some of the crystals. ... the slip-lines were quite similar in appearance to those which are seen when a simple tensile stress exceeding the elastic limit is applied. ... After more reversals of stress additional slip-lines appeared. ... After many reversals they changed into comparatively wide bands

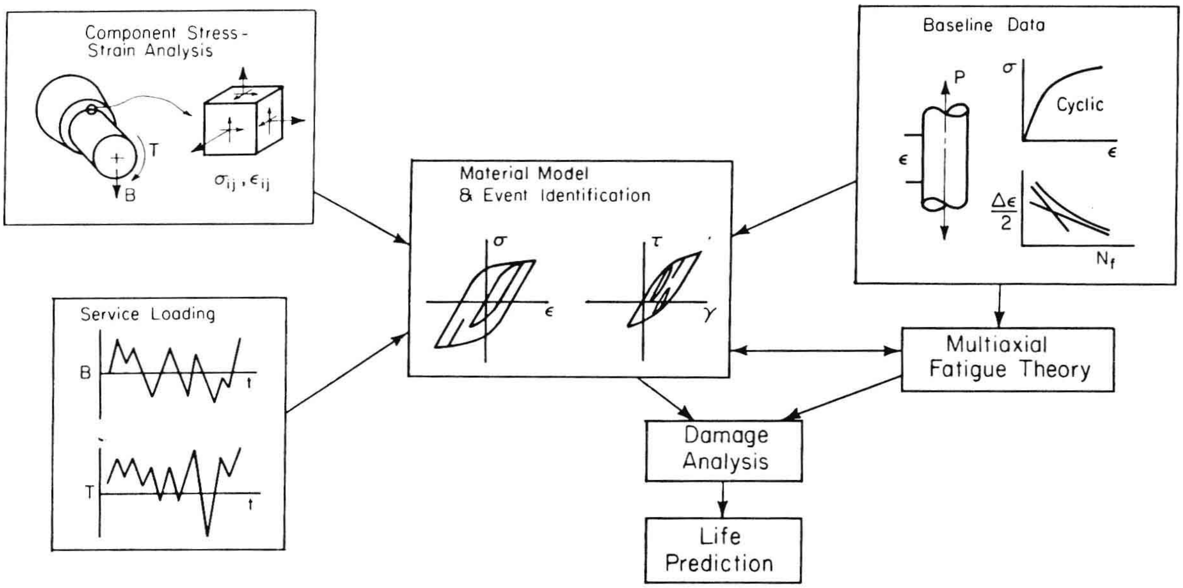



Fig. 2. Schematic illustration of a component fatigue analysis.

with rather hazily defined edges. ... As the number of reversals increased this process of broadening continued, and some parts of the surface became almost covered with dark markings. ... When this stage was reached it was found that some of the crystals had cracked. The cracks occurred along broadened slip-bands; in some instances they were first seen on a single crystal, but soon they joined up from crystal to crystal, until finally a long continuous crack was developed across the surface of the specimen. When this happened a few more reversals brought about fracture." These slip bands were observed to occur on planes of maximum shear stress. In addition, these researchers recognized that "Once an incipient crack begins to form across a certain set of crystals, the effect of further reversals is mainly confined to the neighborhood of the crack".

It is therefore generally accepted that the basic cause of fatigue crack nucleation is cyclic shear stresses and strains. Subsequent fatigue research has shown that after initiation the cracks may turn or grow on tensile planes. Forsyth [4] designated crack initiation and growth on shear planes as stage I growth. He reported that stage I growth continues until reversals of dislocation movement are prevented. The crack may then turn and propagate in a stage II direction on a plane normal to the maximum principal stress. This change in cracking direction is dependent on strain amplitude, loading mode or state of stress, and material type [5].

State-of-stress effects were studied by several investigators with the work of Gough [6] being noteworthy. He performed studies of the fatigue limit in bending and torsion and found that the ratio of the fatigue limit in torsion to that in bending varied with material tested. He proposed models that would reduce to the maximum shear stress theory for ductile materials and to the maximum principal stress theory for brittle materials such as cast iron. Guest [7] also proposed a single model for both ductile and brittle materials with an adjustable constant that could change the theory from maximum shear stress to maximum principal stress. These models were among the first to incorporate the failure mode of the material in the damage model.

This early work showed that it is not necessary to consider the entire fatigue process in order to make engineering estimates of the fatigue life of a structure or component. Rather, a model which describes the process during the majority of the fatigue life is sufficient.

## 3. Fatigue damage

Detailed crack observations have been made on three materials, AISI 304 stainless steel, Inconel 718 and SAE 1045 steel. These materials exhibit different regions of cracking behavior and represent extremes in the behavior observed in isotropic metals during tensile and torsional fatigue testing. Experimental data and observations can be found in earlier papers by Hua and Socie [8], by Socie *et al.* [9] and by Bannantine and Socie [5].

The behavior of the three materials subjected to tension and torsion is summarized in Figs. 3–5. In these figures, the vertical axis is in terms of the life fraction $N/N_f$ and the horizontal scale is presented in terms of fatigue life $N_f$ in cycles. The full line represents the first observation of a surface crack 100 $\mu$m and serves as a demarcation between crack nucleation and growth. It could be argued that nucleation occurs much earlier, say for example 10 $\mu$m. This would simply shift the

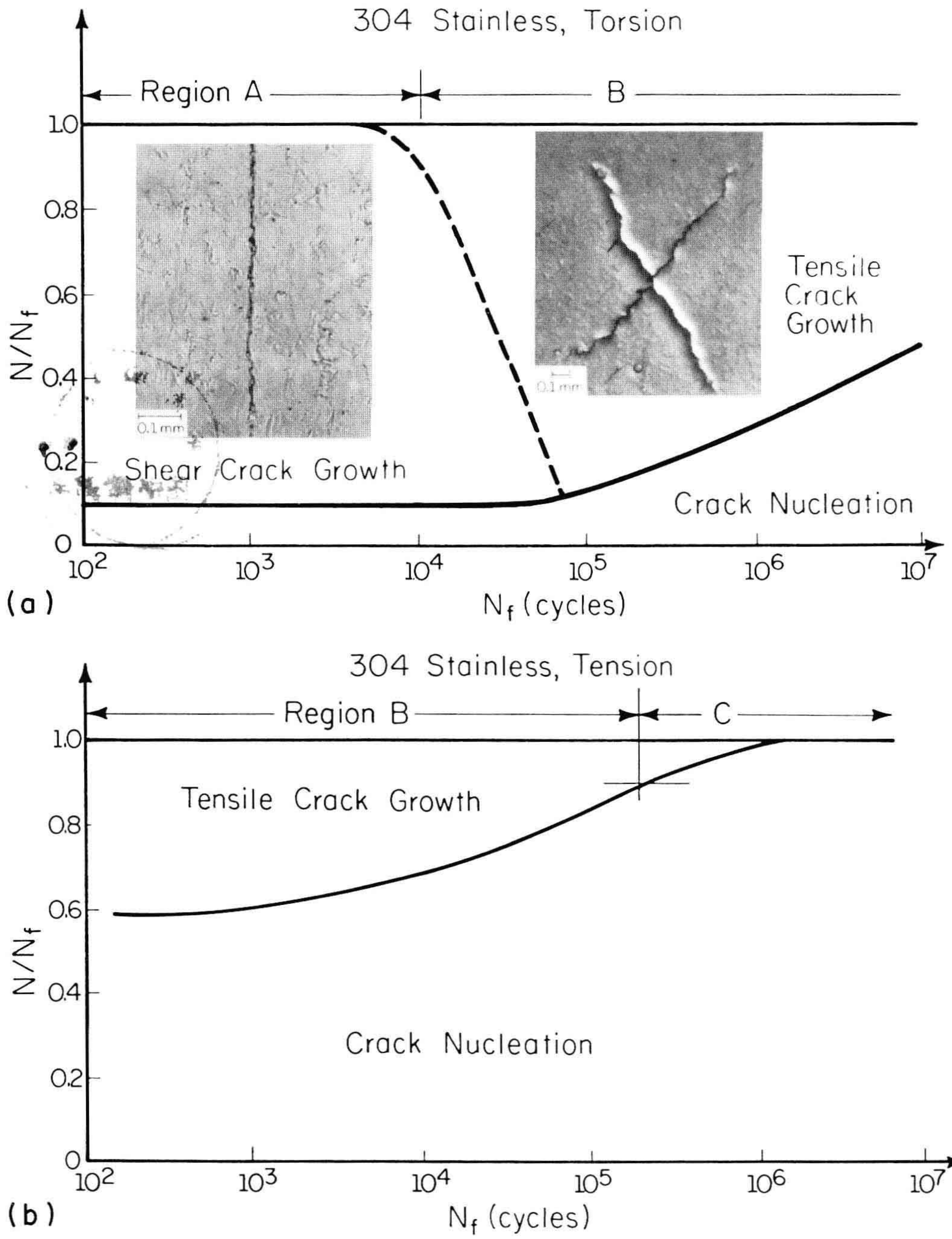



Fig. 3. Cracking behavior observed in AISI 304 stainless steel tested in (a) torsion and (b) tension.

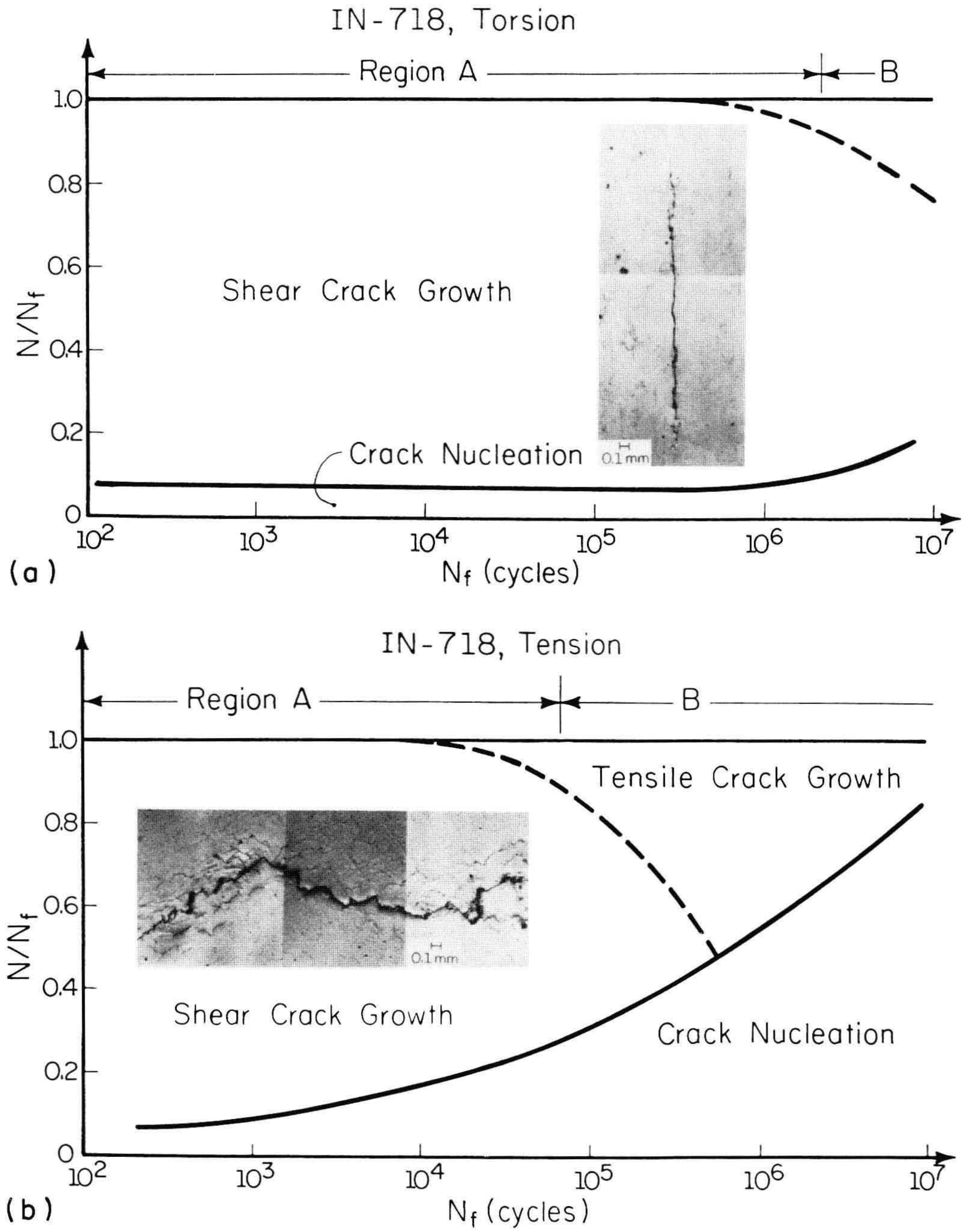
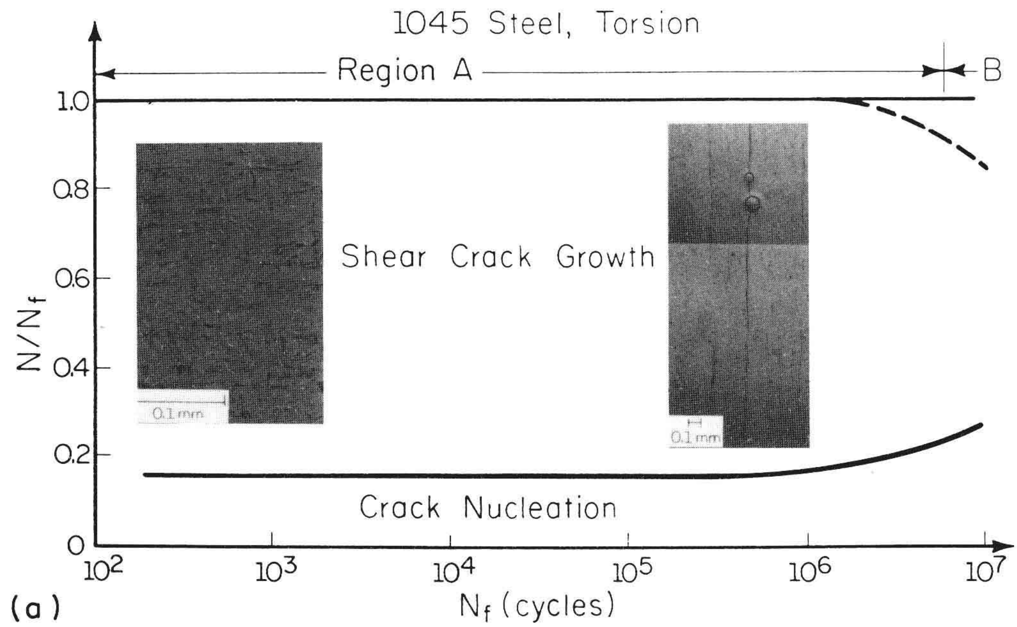



Fig. 4. Cracking behavior observed in Inconel 718 tested in (a) torsion and (b) tension.

line downward without changing the qualitative phenomena represented by the plots. The broken line represents the demarcation between crack growth on planes of maximum shear strain amplitude and crack growth on planes of maximum principal strain amplitude. Cracking behavior is categorized into three general regions: regions A, B and C. Region A denotes a failure mode which is dominated by shear crack growth. In region B, shear crack nucleation is followed by crack growth on planes of maximum principal strain (stage II planes). The fatigue life represented in region C is dominated by crack nucleation. Materials may exhibit cracking behavior which is



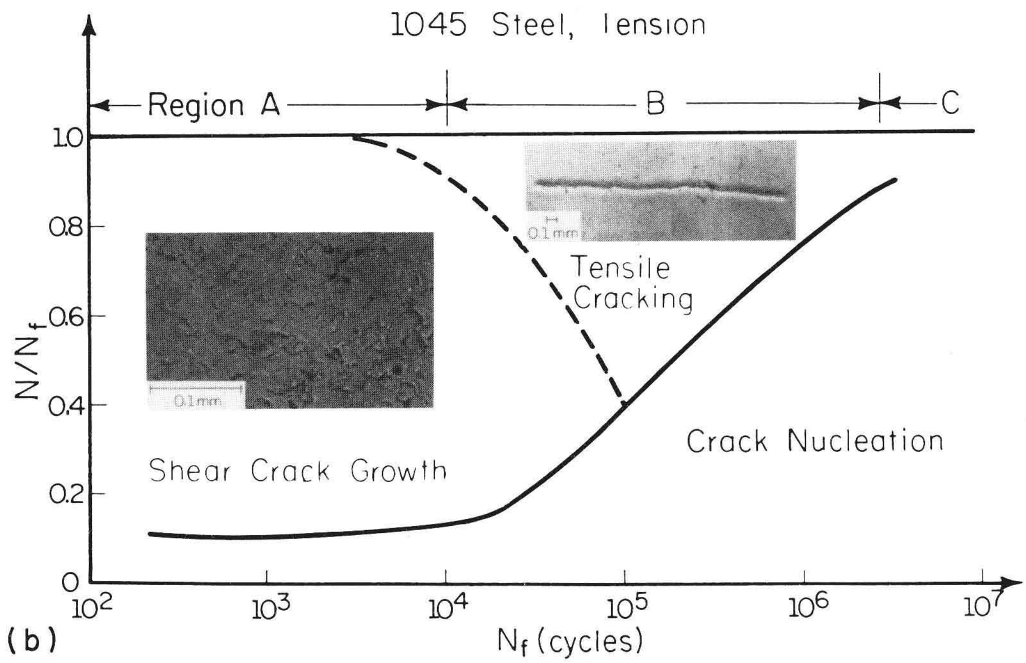




Fig. 5. Cracking behavior observed in SAE 1045 steel tested in (a) torsion and (b) tension.

representative of one, two or all three of these regions. The cracking behavior of each of the three materials is discussed in detail below.

AISI 304 stainless steel (yield strength, 325 MPa) was tested in tension and torsion. The type of cracking behavior exhibited is summarized in Fig. 3(a) for the stainless steel tested in torsion. Cracking behavior could be categorized into two regions: regions A and B. Region A behavior was observed at short lives. Microcracks initiated on shear planes. Once initiated, the cracks became more distinct but showed no significant increase in length. At failure, a large density of small coarse cracks dominated the surface of the specimen. A small amount of branching onto tensile planes (stage II planes) was observed. Failure cracks grew on either shear planes (stage I planes) or tensile planes (stage II planes) by a slow linking of previously initiated shear cracks. Region B is characterized by shear crack nucleation followed by crack growth on planes of maximum principal strain amplitude (stage II planes). Shear crack growth consumes a small fraction of the fatigue life. Region C behavior was observed at the longest lives in torsion. The fraction of life spent growing the crack on shear planes was reduced, as was the crack density. A small number of cracks initiated on shear planes but quickly branched to stage II planes. Growth on these planes occurred by the propagation of the main crack rather than by a linking process.

Surface replicas and scanning electron examination of fracture surfaces of AISI 304 stainless steel specimens tested in tension showed no perceptible evidence of stage I growth. As a result, no region A behavior is shown in Fig. 3(b). The fracture surfaces appeared to be almost entirely dominated by stage II growth. Plumbridge [10] also reported that, at low strain amplitudes, up to 90% of the fatigue life may be taken up in initiation and stage I growth while at high strain amplitudes a similar fraction may be spent in stage II crack growth.

The behavior of Inconel 718 (yield strength, 1110 MPa) is summarized in Fig. 4. The behavior of Inconel 718 tested in torsion is presented in Fig. 4(a). Unlike the stainless steel which displayed a mixed behavior, the results of the Inconel 718 torsion tests showed that cracks initiated and remained on the maximum shear planes (region A behavior) at all values of the shear strain investigated. Even at the lowest strain amplitude, in which the normal stress–strain response was essentially elastic, cracks initiated and remained on shear planes throughout the life. The crack density decreased with increasing fatigue life as it did in AISI 304 stainless steel but no branching onto tensile planes was observed.

Under tensile loading, cracks remained on shear planes for the majority of the fatigue life and a large zone of region A behavior was observed (see Fig. 4(b)). Final failure in all tension tests was in a macroscopic tensile direction consisting of large portions of microscopic shear growth. Large amounts of shear growth were observed at failure for short and intermediate fatigue lives. Growth on stage II planes occurred only late in life.

Damage accumulation in Inconel 718 appears to be shear dominated. This is attributed to localized shear deformation bands developed during cyclic loading. Reversed movement of dislocations progressively shears precipitates in these bands. Crack propagation then occurs along the bands with extensive shear crack growth exhibited throughout the fatigue life.

Two types of cracking system have been observed in the hot-rolled and normalized SAE 1045 (yield strength, 380 MPa). A large density of microcracks was observed at high strain amplitudes, with the final failure occurring by a very rapid linking of these cracks. This type of damage has been termed the R system by Marco and Starkey [11]. Alternatively, the S system, which dominated crack behavior at low strain amplitudes, exhibited one dominant crack which grew until failure.

In torsion, at high amplitudes, the R system crack behavior was characteristic of region A shown in Fig. 5(a). Two common features were observed. First, the number of microcracks increased with increasing number of loading cycles. Second, the surface length of microcracks which appeared in the early stages remained almost unchanged during the fatigue life. Darkness and clarity of the microcracks substantially increased with progress of fatigue cycles. These observations indicate that the crack opening and hence the crack depth increased. Cracks initiated on the surface and propagated into the surface, while the surface crack length remained nearly constant. Also, crack orientations were developed equally on both planes of maximum shear. These multimicrocracks were almost uniformly distributed over the entire gauge length. The failure was similar to that observed in the stainless steels at high amplitudes except that the linking of microcracks and final failure in SAE 1045 steel occurred over a very few cycles, while the growth of the region A failure crack in stainless steels occurred progressively throughout the fatigue life. At lower amplitudes, progressive growth of a single crack occurred by a linking process on the shear plane.

Region B behavior was observed only at long lives. At the lowest strain amplitude, 0.26%, the crack branched and growth occurred on the tensile plane by a linking of previously initiated shear