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FATIGUE CRACK CLOSURE AND CLOSURE DEVELOPMENT IN A HIGH STRENGTH ALUMINIUM ALLOY

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A dissertation submitted to the Faculty of Engineering, University of the Witwatersrand, Johannesburg, in fulfilment of the requirements for the degree of Master of Science in Engineering.

Johannesburg 1988

DECLARATION

I declare that this dissertation is my own unaided work, except where due reference is made to others. It is being submitted for the degree of Master of Science in Engineering at the University of the Witwatersrand, Johannesburg. It has not been submitted before for any degree or examination in any other University.

REINER ERNST GARZ

15th day of July 1988

ABSTRACT

Fatigue crack closure characteristics were studied in a 7017 aluminium alloy in the as received (AR) and heat treated (HT) conditions with the view of establishing the general closure trends and how well the modified closure parameter ΔK_{eff} (= K_{max} - K_{op}) was capable of characterising fatigue crack growth. The results of this work indicated that a significant amount of strain intensification occurred below K_{op} which implied that ΔK_{eff} generally underestimated the stress intensity range experienced at the crack tip during cyclic loading.

The major objective of this dissertation was to determine whether a relationship exists between the distance over which closure develops from zero to a steady state closure value, and the distance over which short crack growth behaviour occurs.

Two techniques were used to eliminate closure for a fatigue crack in a compact tension specimen. Closure development was then measured as a function crack length and a steady state closure value was approached after about 0.19 mm in the AR material and 0.28 mm in the HT material irrespective of ΔK. This distance compared well with that distance over which short cracks exhibited "anomalous" behaviour. This implies that anomalous short crack behaviour is dependent on the distance over which closure develops in this alloy, although microstructural influences may play a role.

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CONTENTS

	DECLARATION	ii
	ABSTRACT	iii
	ACKNOWLEDGEMENTS	iv
1	INTRODUCTION	1
	1.1 BACKGROUND AND AIMS	1
2	LITERATURE REVIEW	5
	2.1 INTRODUCTION	5
	2.2 FATIGUE CRACK CLOSURE	6
	2.2.1 Plasticity Induced Closure	8
	2.2.2 Roughness Induced Closure	9
	2.2.3 Oxide Induced Closure	9
	2.2.3.1 Separating out closure mechanisms	10
	2.2.4 Plane Stress vs Plane Strain	12
	2.3 CLOSURE MEASUREMENT	14
	2.3.1 Measurement Techniques	14
	2.3.2 Closure Measurement Problems	16
	2.3.2.1 Crack opening/closing point	16
	2.3.2.2 Technique resolution	17
	2.3.2.3 Compliance gauge location	19
	2.4 FATIGUE PHENOMENA EXPLICABLE IN TERMS OF CLOSURE	20
	2.4.1 Near Threshold Growth	20
	2.4.2 Mean Stress and Closure	21
	2.4.3 Fatigue Crack Retardation	22
	2.4.4 Microstructure and Environmental effects	24
	2.4.5 Crack Growth Under Compressive Loading	26
	2.4.6 The Effect of Time on Closure	28
	2.5 SHORT CRACK GROWTH	28
	2.5.1 Reasons for Faster Short Crack Growth	29
	2.5.2 Through-thickness Short Cracks	33
	2.5.3 The Extent of the Short Crack Regime	33
	2.5.4 Closure and Closure Measurement in Short Cracks	35
	2.5.5 Short Crack Characterisation	36
	2.6 CLOSURE REDUCTION AND REDEVELOPMENT	37

3	EXPERIMENTAL PROCEDURES	42
	3.1 MATERIAL CHARATERISATION	42
	3.2 GROWTH RATE CHARACTERISATION	44
	3.2.1 Long Crack Growth	44
	3.2.2 Short Crack Growth	45
	3.3 CLOSURE CHARACTERISATION	46
	3.4 CLOSURE DEVELOPMENT	47
4	PROBLEMS WITH CLOSURE MEASUREMENT AND INTERPRETATION	49
	4.1 THE PARAMETER ΔK_{eff}	49
	4.2 EXPERIMENTAL TESTING	50
	4.2.1 The R Ratio Test	50
	4.2.2 The Wedge Test	51
	4.2.3 The Particle Test	53
	4.3 SUMMARY	55
5	GROWTH RATE AND CLOSURE DATA	57
	5.1 GROWTH IN THE AS-RECEIVED MATERIAL	57
	5.2 5 mm HEAT TREATED CT SPECIMENS	58
	5.3 DISCUSSION	59
	5.3.1 Closure Effects	59
	5.3.2 The Effect of Heat Treatment on Crack Growth	59
	5.3.3 Thickness Effects	60
	5.4 SEPARATING OUT CLOSURE EFFECTS	60
	5.4.1 Roughness Induced Closure	60
	5.4.2 Plasticity Induced Closure	61
	5.5 GROWTH UNDER COMPRESSIVE LOADING	62
	5.5.1 Growth Mechanisms Under Compressive Loading	63
	5.6 SUMMARY	64
6	ELIMINATION OF CLOSURE IN LONG CRACKS AND ITS REDEVELOPMENT	65
	6.1 WAKE "REMOVAL" EXPERIMENTS	65
	6.2 CLOSURE REDEVELOPMENT	65
	6.3 TENSILE GROWTH FOLLOWING COMPRESSIVE LOADING	68
	6.4 DISCUSSION	70
	6.4.1 Closure Elimination	70
	6.4.2 Closure Traces	71
	6.4.3 Closure and the Closure Development Distance	72
	6.5 SUMMARY	73
7	COMPARISONS WITH SHORT CRACK DATA	74
	7.1 SHORT CRACK GROWTH	74
	7.2 DISCUSSION	75

8 SUMMARY OF CONCLUSIONS 78

9 REFERENCES 80

10 FIGURES 91

1 INTRODUCTION

The defect tolerant design approach now used commonly for life predictions on metal structures has ensured a reduction in catastrophic failures and an increase in component life. Yet failures still occur in instances where the materials being used have not been fully characterised for the specific conditions experienced by the components in actual structures. This thesis deals with aspects of one of the problem areas in such material characterisation, namely, fatigue crack closure, with specific emphasis on short cracks.

1.1 Background and Aims

Characterisation of fatigue crack growth was originally directed towards long cracks, which are easy to measure and detect. Much research has been done in this field over the last 25 years during which time it has become apparent that cracks below certain sizes in the range 0.05 to 1.5 mm (termed short cracks), often propagate at different rates to long cracks (> 10 mm). Determination of the causes of this difference then required detailed analysis of short crack growth, which was beset with difficulties similar to those encountered with near-threshold crack growth, namely non-continuous growth and the limited resolution of growth rate monitoring techniques, as well as the difficulties inherent in detecting and working with small cracks.

Because long crack growth is generally well characterised in terms of the global parameter ΔK, the tendency in small crack research has been to try to modify the K approach so as to rationalise differences between short crack and long crack growth. The greatest difficulty in making such modifications is probably related to the fact that the stress intensity factor is a continuum parameter whereas short crack growth often is noncontinuum, being microstructurally dependent. This implies that, at least for very small cracks (approximately equal to

the grain size), a global parameter such as K is an inappropriate parameter. The approach to the early stages of crack growth appears likely to remain in the realms of either probabilistic analyses or the S-N approach, rather than be amenable to characterisation by current fracture mechanics parameters.

Nevertheless with certain modifications to ΔK (allowing for fatigue crack closure), it has been possible to correlate short and long crack growth down to crack sizes of the order of several characteristic microstructural units in certain metals. Equally, by incorporating closure effects, it has proved possible to explain observed threshold behaviour for various alloys and to elucidate crack growth rate fluctuations following the application of overloads and underloads.

There are, however, certain obstacles which must be overcome when incorporating closure effects into predictions of fatigue crack growth rates.

Although extensively researched since Elber's classic paper [1] in the early 1970's, fatigue crack closure does not lend itself to easy measurement or interpretation. Consequently many results are not self consistent, and closure trends presented in the literature vary markedly with measuring technique (as will be discussed in detail in section 2.3). This can cause major difficulties when comparing and correlating information presented in the literature. In order to obtain a clear overview of the role of crack closure in fatigue crack growth it is imperative that confidence limits be established for the measuring technique. This then allows sensible judgements of validity to be made when comparing various closure results.

This thesis has as its purpose the examination and clarification of some of the unresolved questions in the field of crack closure measurement and development. In summary its objectives are:-

1. To characterise the closure properties of a high strength aluminium alloy (Al 7017 T6), and establish the degree of reliability afforded by compliance based closure measurement techniques under various conditions. This objective includes an investigation of the individual contributions of the various closure mechanisms (to be discussed in section 2.2) under high and low ΔK.
2. To estimate the distance over which closure develops to a steady state value (for a long crack which is initially closure-free)

for various stress intensity ranges and R ratios (where $R = K_{min}/K_{max}$.) These results may then be used to determine whether the distance over which closure develops (in a long crack) may be related to the extent of anomalous short crack behaviour. If it can, this technique may enable a more accurate estimate of the short crack regime to be made (the short crack regime is generally estimated by either the attainment of a specific value of ΔK, i.e. the threshold value for long cracks or a decrease in growth rate fluctuations. Considerable scatter is attendant on these techniques).

The idea of examining closure development in a long crack, rather than a short crack for which the information is required, arises from the difficulties associated with obtaining reliable and physically meaningful closure data from short cracks.

In endeavouring to become familiar with the characteristics of fatigue crack closure development in the wake of a crack, (which will affect the extent of the short crack regime) it is necessary to understand the basic mechanisms involved and the way in which closure development is affected by various parameters. To clarify such effects the following aspects will be discussed in the body of the thesis:

1. The effect of mean stress, (section 2.4.1.)
2. Fatigue crack retardation following overloads, which is of importance to this study because compressive overloads were used to eliminate closure in long cracks, (section 2.4.2.)
3. Microstructural and environmental influences, (section 2.4.3.)
4. Closure and near-threshold growth, (section 2.4.4.)
5. Crack growth under compressive loading, (section 2.4.5.) which is deemed relevant as a result of the experimental procedures used to examine closure development.
6. The effect of time on closure, (section 2.4.6.)

Following this exposition of current understanding of closure, the inherent difficulties associated with its measurement and the effectiveness of closure in explaining various fatigue phenomena will be discussed. The experimental part of the thesis centres on the technique of closure removal in order to obtain a crack effectively without a closure-generating wake, as occurs in short cracks, together with an examination of how closure redevelops. The ultimate aim of the work is to link the distance over which closure develops to the extent of short crack behaviour. Understanding of the mechanisms involved in closure build-up should throw some light on factors which influence

short crack growth, and hence provide insight into alloy design to resist enhanced growth rates for small cracks.

2 LITERATURE REVIEW

2.1 Introduction

The utility of linear elastic fracture mechanics (LEFM) in fracture or fatigue studies derives from the fact that certain material properties, e.g. resistance to further growth, in the presence of a crack can be measured in terms of a single governing parameter, the stress intensity factor K. Crack growth rates characterised by this global parameter, which more or less independent of local crack tip microstructural influences, has generally been very successful. Even when LEFM conditions no longer exist, such as for the case of bulk plasticity, ΔK may still sometimes correlate growth rates with those obtained under nominally elastic conditions [2]. In general, however, crack growth predictions under non-identical crack tip conditions are hazardous. This is clearly to be expected as K is a globally based parameter, and where similitude between crack tip conditions does not exist the use of K is limited (see section 2.5).

Before attempting to extend the use of K to non standardised situations, such as short crack growth, more precise information regarding the underlying mechanisms of fatigue crack growth are necessary. This requires greater knowledge of actual deformation processes at the crack tip including environmental, microstructural and crack wake effects.

Fatigue crack closure is a crack wake effect which occurs as a result of the wedging of the fracture surfaces during unloading, thereby reducing the local ΔK value experienced at the crack tip below the nominal applied value. Closure plays an important role in fatigue crack growth especially at low R ratios. Because closure is associated with the crack wake, short cracks, which by definition have a very limited wake, will not experience closure effects. Short crack growth rates are therefore expected to be different. The crack length over which closure develops to some steady state value for a certain

material is, therefore, anticipated to bear some relation to the length below which short cracks behave differently to long cracks, (discussed in section 2.5).

2.2 Fatigue Crack Closure

Use of the parameter ΔK in fracture mechanics is based on the premise that, at least under tension-tension loading, the crack tip will experience the whole range of stress intensity (i.e., ΔK) from K_{min} to K_{max} and that the crack tip will therefore be open during this time. The concept of closure, on the other hand, requires that the crack tip is closed during part of the fatigue cycle (and is therefore subject to compressive load transfer across the crack faces) and there is hence no contribution to crack propagation while the crack tip does not experience tensile loading. It follows that crack tip conditions may no longer be able to be characterised by the global loading conditions used to define K. The crack driving force may then be given by subtracting K_{op}, which is the stress intensity value corresponding to the elimination of crack closure during loading [1], from the maximum value in the loading cycle (K_{max}). Hence the original driving force ΔK is reduced to an effective driving force ΔK_{eff}, defined as

$$\Delta K_{eff} = K_{max} - K_{op}.$$

The existence of closure was first noted from load vs displacement traces [1] which showed the trace to be non-linear below a certain load. Effectively, the situation in which the crack is closed corresponds to a specimen with a lower compliance (ie. stiffer) than when the crack is open. This is essentially because the crack in the specimen is now effectively shorter due to crack wake contact.

Direct evidence of the existence of closure was later obtained by Sunder and Dash [3] who used a loading scheme in which K_{max} was kept constant and K_{min} progressively decreased to zero. The resulting fatigue striations produced by the loading scheme, showed an increase in striation width as K_{min} decreased (and ΔK increased) to some value K. Below this value, however, the striation width remained constant indicating that this value of K equalled K_{op}, and hence there was no additional contribution to growth rate. (The value of K_{op} measured by this technique agreed with that obtained by more

conventional compliance measuring techniques to within 10 - 15% [4].)

Measurements of the plastic zone size ahead of a crack tip have also indirectly confirmed the existence of closure [5, 6]. These results showed the plastic zone size to be significantly smaller than that calculated using the Irwin equation;

$$r_p = \frac{1}{2\pi}\left(\frac{K_{max}}{\sigma_y}\right)^2$$

from which it is suggested that a modified equation may be more appropriate, ie.

$$r_p = \frac{1}{2\pi}\left(\frac{K_{max\ effective}}{\sigma_y}\right)^2$$

where $K_{max\ effective}$ is $K_{max} - K_{op}$ ($= \Delta K_{eff}$).

Three basic mechanisms have been identified as causing closure under general fatigue growth conditions. These mechanisms are plasticity induced [1], roughness induced [7] and oxide induced closure [8], and have been extensively reviewed by, amongst others, Bachman [9], Schijve [10,11] and Ritchie [12].

The magnitude of the individual closure mechanisms depends on the growth mechanisms that are operative, namely, near-threshold growth or mid growth regime. The growth rate regimes are illustrated in Fig. 2.1 [13] In the mid growth rate regime a striation formation mechanism exists which causes intense localised shear deformation in flow bands near the crack tip. This results in the creation of new crack surface by shear decohesion at the tip. In the near-threshold regime, growth is confined to one shear direction, namely, the primary slip system in the direction of growth. Such crack propagation proceeds under a combination of mode II plus mode I displacements. The above mechanisms are illustrated in Fig. 2.2 [13]. The strong mode II component at near threshold growth rates has been confirmed by various workers [14-17].

2.2.1 Plasticity induced closure

Plasticity induced closure is a result of residual displacement, (a consequence of the plastic deformation in the crack tip plastic zone), and the constraint of the elastic bulk of the material. This constraint of the remaining specimen causes the crack not only to close, but ensures that the crack faces even experience compressive forces during the early stages of globally applied tensile loads.

Plasticity induced closure occurs during all stages of fatigue crack growth, but its effect is maximised during load shedding due to the strain distribution, as depicted in Fig. 2.3. Confirmation of wake "clamping" forces occurring well behind the crack tip has been reported by James [18]. In contrast, when crack growth occurs under increasing ΔK conditions the length of wake which contributes to closure, appears (at least for a quenched and tempered steel) to be of the order of 100 μm [19], because the residual plastic deformation in the wake is always less than that at the crack tip due to the continually increasing plastic zone size.

Newman [20] successfully correlated numerical closure results from finite element analysis (FEA) of plasticity induced closure with those obtained experimentally for the following conditions:

a. constant amplitude loading
b. various R ratios and
c. two level block loading.

The fact that the above correlation was successful would suggest that plasticity induced closure was the dominant closure mechanism for the material and conditions tested, (ie. non-threshold growth rates). This mechanism is generally the dominant mechanism in the linear growth regime as the other mechanisms require mode II displacements which are generally limited to the near-threshold growth rate regime, although certain materials exhibit very tortuous growth (single slip growth see Fig. 2.2a) even at high ΔK values which implies that other closure mechanisms will also be important.

2.2.2 Roughness induced closure

Macroscopic surface roughness may be two orders of magnitude greater than the maximum crack tip opening displacement (CTOD) at K_{max} [21], from which it appears that the fracture surface roughness does not have a direct relation to closure. Some relation must, however, exist because a greater fracture surface roughness generally produces a greater closure value. The mechanism by which roughness induced closure is believed to occur involves mode II displacement, which causes crack surface mismatch. This is generally most important in the near-threshold growth regime or where growth is strongly crystallographic.

The degree to which fatigue growth conditions affect the fracture surface roughness varies. Grey et al [22] reported that roughness varies with growth rate and environment, while for constant growth rate, R ratio [22], and K_{max} [23] have been reported to have no influence. Minakawa et al [24] showed roughness to vary with R ratio and therefore with K_{max} as shown on micrographs of the fracture surface in Fig. 2.4.

Fracture surface roughness is, however, strongly dependent on grain size, so much so that the fracture surface produced on a 9021 powder metallurgy (P/M) aluminium alloy with an ultrafine grain size of 0.2 μm was so smooth so to ensure that no closure was detected [25]. (See section 2.4.3)

2.2.3 Oxide induced closure

Oxide induced closure arises through a process of fretting corrosion which is at its maximum at near-threshold growth rates because the crack tip opening displacement (CTOD) is such that contact between the crack faces may occur over significant portions of the loading cycle [13]. High fracture surface roughness, combined with a mode II (shear) component, accelerates fretting, and therefore the production of oxide debris. Oxide layers 20 to 40 times greater than the natural oxide thickness have been measured on fatigue fracture surfaces [13], and attributed to this mechanism.

Although contact is greatest at near-threshold growth rates, if closure occurs at high growth rates there will still be contact and

fretting may occur, thus the formation of oxide debris (oxide induced closure) need not be restricted to near-threshold growth. However, because of faster growth rates the time to form debris will be less, and the thickness therefore less.

There is some controversy surrounding the importance of oxide induced closure in aluminium alloys (at low growth rates). Most workers claim there can be no significant effect because the oxide layers on the fatigue fracture surfaces are very thin [25, 27], of the order of 5-10 nm [28-30]. However for a 7X75 aluminium in the overaged (OA) temper [30] oxide layers of 0.115, 0.1 and 0.09 μm have been reported for fatigue crack growth at R ratios of R = 0.1, 0.33 and 0.8 respectively. It may be noted that at R = 0.8 the CTOD at K_{min} is greater than the oxide thickness and it is, therefore, surprising that the oxide layer can be so thick, as fretting would have been very limited. On the other hand Blom and Holm [31] did find that for R = 0.5 the corrosion debris was much less than for R = 0.05, which would be expected if the mechanism of oxide thickening was as described above.

Oxide layers of about 0.1 μm, are of the order of the maximum CTOD at near-threshold growth rates in aluminium alloys [30]. Thus the OA alloy in the above investigation of Vasudevan [30] would have been expected to have the greatest ΔK_{th} because of the large closure contribution. Often, however, the greatest oxide layer in aluminium specimens corresponds to the lowest threshold value, which has been explained in terms of a pronounced embrittling effect [32]. Other factors, eg. crack path tortuosity, fracture surface roughness and an intrinsic lower resistance to crack propagation, make it difficult to separate out the exact influence of the oxide induced closure contribution. Measurement of oxide effects is also made very difficult by the fact that contact occurs very close to the crack tip (2-10 μm) [33].

2.2.4 Separating out closure mechanisms

It would be very useful from a predictive viewpoint to know the contributions that the various closure mechanisms make to the measured value at various ΔK values. This is rather difficult to determine, however, as will become clear in this section.

Various authors have attempted to clarify the actual manner in which closure occurs under various loading conditions. This has been done

with the use of infiltration techniques [7, 34, 35] (see Fig. 2.5a) or by monitoring the signal amplitude of shear waves diffracted at each wake contact point [36] The data obtained indicated that for an 2024-T3 aluminium alloy the crack did not close systematically from the crack tip backwards as the load decreased. It can then be surmised that discrete contact points behind the crack tip may thus act as wedges which prevent the stress intensity at the crack tip from falling to K_{min} as the load decreases to K_{min}, (at least for low R ratios). The crack thus "wedges" at K_{op}. If elastic and plastic deformation at the contact points occur after initial contact while the loads are being reduced, as illustrated in Fig. 2.5b, the crack tip may experience strains below K_{op}, effectively causing the strain range at the crack tip to be greater than that corresponding to the excursion from K_{op} to K_{max} [34, 37]. This phenomenon is referred to as strain intensification.

Contact positions in the crack wake may be expected to relate to the predominant closure mechanism. The plasticity and roughness induced closure act in somewhat different ways i.e., the former is likely to result in closure occurring (relatively) smoothly and sequentially from the crack tip backwards, while the latter results in closure at discrete contact points which prevents the crack tip from closing uniformly.

From the above discussion, it appears that under conditions where closure occurs predominantly b wedging at discrete points behind the crack tip there may be a contribution to crack growth from that portion of the cycle below K_{op}. The extent of the contribution will depend on the area of the contact points, which is proportional to that area of wake which would contribute to closure in the absence of discrete wedging. This obviously depends on material, microstructure and the relative contributions of roughness induced and plasticity induced closure.

For a given material and microstructure it is, however, difficult to devise experiments which will show to what extent the various closure mechanisms are responsible for closure. One possibility for measuring the magnitude of plasticity induced closure produced by load shedding involves the use of several load shedding sequences to achieve a given ΔK value, as shown in Fig. 2.6. The closure values may then be used to compare between the different load shedding schemes, as roughness induced closure contributions should not be affected unless the crack growth mechanism changes. Changes in measured closure values should

therefore reflect changes in plasticity induced closure and may allow an estimate to be made of its absolute value.

Because the fatigue fracture path in many peak-aged (PA) and underaged (UA) aluminium alloys is highly tortuous, and therefore strongly microstructurally dependent [27], (producing a very rough fracture surface), various workers have suggested that roughness induced closure is the dominant closure mechanism [7,27]. However, no concrete evidence exists to support this suggestion.

Numerous models have been proposed for the effects of roughness [38-40] and oxide [8,41] induced closure, but because these are generally based on fitting parameters to semi-emperical equations they cannot, as yet, be used in a predictive capacity for systems other than those for which they were derived. They do, however, give some understanding of the mechanisms of closure.

Generally, therefore, separation of the contributions of different closure mechanisms is not possible, except on the basis of assuming that in regime B of the growth rate curve (see Fig. 2.1) plasticity induced closure is dominant. It may then be feasible to attribute enhanced closure values near the threshold to roughness or oxide induced closure, although the load shedding schemes generally used in threshold testing can themselves increase the plasticity induced closure component [18].

2.2.5 Plane stress vs plane strain

The amount of plastically deformed material at the fatigue crack tip varies depending on the material constraint and thus on whether deformation occurs under conditions of plane stress or plane strain. Altering the amount of plastically deformed material will obviously affect the degree of plasticity induced closure. For sufficiently thick fatigue specimens plane stress conditions occur at the surface, which gradually transform to plane strain conditions as one moves to the centre of the specimen. This phenomenon leads to problems with interpretation of closure measurements, specifically whether only surface (plane stress) or bulk (combined plane stress and plane strain) closure is being measured, and whether closure does occur under conditions of plane strain.

Contrary to the suggestion of some earlier workers in the field (see

the review of reference [2]), closure generally exists under both plane stress and plane strain conditions as shown, for example, by direct measurements made by Fleck and Smith [42], using a push-rod assembly inserted into holes drilled to just above and just below the crack plane (shown in Fig. 2.7). This work showed that closure existed during 20% of the load cycle at the centre of the specimen. Although closure exists under conditions of plane strain, its magnitude seems to vary significantly with material and testing conditions [2, 43, 44].

The stress state plays an important part in understanding overload effects, where the extensive plastic deformation at the surface (plane stress) may strongly affect retardation of growth rates subsequent to overloads (due to the high magnitudes of plasticity induced closure at the surface). In this regard Tokaji et al [45] observed that retardation decreased as specimen thickness increased for specimens up to a thickness of 15 mm, implying that as the plane stress contribution (thin specimens) decreases so too does the retardation. The authors were further able to correlate the effects of the overload in terms of ΔK_{eff} as measured by surface strain gauges, (which measure surface closure) but not using back face strain (BFS) gauges (which measure bulk closure effects.) (See section 2.3 for explanations of the various types of gauges).

The effect of stress state (specimen thickness) on closure, can also be fairly pronounced even in the absence of overloads. For example, Schijve and Arkema [46] found that the closure values (K_{op}/K_{max}) increased from 0.29 to 0.4 as the specimens decreased in thickness from 6 to 1 mm for a fixed applied ΔK.

The interpretation of overload effects is complicated by the fact that, following an overload, a decrease in growth rate with a decrease in specimen thickness need not be a result of the increased plasticity induced closure contribution in plane stress. Some studies have shown that there may be a transition in growth mechanism, to a near-threshold mechanism, with a corresponding increase postulated in the roughness induced closure [47]. Only a detailed fractographic examination will show whether this factor needs to be considered in a particular case.

Furnee and Ewalds [48] conducted tests in which they progressively removed material from the surfaces of a CT specimen of a 2024 T6 aluminium alloy. The results, given in Fig. 2.8, show that there was

an initial large reduction in closure which would be expected from a greater plasticity induced closure contribution at the surface due to a state of plane stress.

2.3 Closure Measurement

2.3.1 Measurement techniques

Many conflicting closure results have been attributed to problems associated with the accuracy or different sensitivities of the various measurement techniques used, and/or to the interpretation of the opening and closing points on the closure traces. (to be discussed in section 2.3.2.1) Extensive literature exists on closure measuring techniques see, for example, Fleck [49] and James [2].

The techniques used to measure crack closure divide into two categories :-

1. Compliance techniques. These include measurements using 1) back face strain (BFS) gauges, 2) strain gauges bonded across the crack line, 3) near tip strain gauges, 4) displacement transducers attached across the crack line or ahead of the crack tip (crack tip clip gauges) and, for SEN or CT type specimens, 5) clip on displacement gauges at the notch-mouth. The various types of gauges are shown in Fig. 2.9.
2. Non-compliance techniques. These include electrical potential (PD) methods, ultrasonic methods, optical interference methods and, for transparent materials, photoelastic techniques.

The techniques which can be applied to yield bulk measurements, ie. an average through thickness closure value, are BFS, notch-mouth clip gauges, potential drop, ultrasonic and photoelastic methods. The other techniques only give closure information at the specimen surface (plane stress).

Fleck [49] has made a comprehensive and systematic study of the merits of the various techniques used and has concluded that compliance techniques are the most consistent and accurate, in agreement with other authors [34].

The techniques which have been most often used in the literature are PD techniques, compliance methods and acoustic emission techniques.

However, some facts should be noted with regard to the use of PD and acoustic emission techniques.

Potential drop techniques are affected by the presence of an oxide layer on the fracture faces, which may prevent electrical contact between contacting surfaces, and thus closure will not necessarily be detected. Evidence supporting this effect comes from tests in air and vacuum, where the vacuum environment produced higher closure values using the PD technique, while ΔK and growth rates were similar. Such behaviour would be expected if an oxide layer formed in an air environment and prevents contact, whereas no oxide layer forms in a vacuum environment [50].

PD techniques may also give erroneously high closure values if current is transmitted while asperities just slide past each other without transferring any load [51]. Such an event could also cause acoustic closure measuring techniques to give higher closure values than actually exist [52]. PD measurements may also give inaccurate results due to changes in resistance as plasticity varies and to variations caused by contact of individual asperities.

Of the compliance methods, strain gauges are reported to exhibit less drift and abnormal hysteresis effects compared to clip gauges [53], and the BFS gauge is the easiest to use from the point of view of inaccuracies in closure measurement resulting from poor positioning, because this is not critical [54] for these gauges. The BFS gauge is also reported to have twice the sensitivity of notch mouth clip gauges [19] and an estimated relative error of less than 10% [28].

Besides the uncertainty associated with the accuracy of the various measurement techniques, uncertainty also exists regarding the exact position of the closure or opening point on a compliance trace. This is discussed in the following section.

2.3.2 Closure measurement problems

2.3.2.1 Crack opening /closing point

The closure magnitude is generally determined from the load on a compliance trace where the trace deviates from linearity during unloading (P_{cl}) or achieves linearity on loading (P_{op}) see Fig. 2.10. The physical interpretation of the nonlinearity during the bottom portion of the cycle is that crack wedging is occurring, and the crack appears shorter than it really is, thereby causing the specimen to appear stiffer. A linear portion at the beginning of the loading half cycle indicates that the crack requires a certain minimum load to start unwedging.

An alternative way of defining crack opening utilises the intersection of tangents from the linear elastic portions of the load displacement trace corresponding to the fully open and the totally closed crack. This is easy to do but its actual physical significance appears to be limited, whereas the point where the trace deviates from linearity does represent (as accurately as the system can) the beginning of crack face contact.

The following example illustrates the importance of the manner used to choose P_{op}. Figure 2.11 shows a set of curves obtained by Buck et al [55] for which the closure value increases as R increases if tangent intercepts are used to determine P_{op}. Alternatively the closure value remains approximately the same, if the closure value is chosen at the onset of linearity during loading. The latter trend is the one which is generally accepted [56] as correct, and is better able to explain the growth rate trends of Buck et al than the original tangent hypothesis.

Better resolution of the crack tip opening point is obtained by incorporating an offset elastic displacement circuit (see figure 2.12) into the compliance measurement system. This technique, first used by Kikukawa et al [57] generates an offset displacement $\delta' = \delta - \alpha P$ where α is chosen such that $\delta - \alpha p$ is zero if the compliance trace is linear. This allows amplification of the change in compliance associated with crack opening. The resolution of the offset circuit has been reported to be a crack length increment of the order of 50 μm by James [19] and 25 μm by Kikukawa [57] (for "long" cracks.)

Where contact between the crack faces is slight, the increased accuracy of the offset circuit may give a vastly greater P_{op} load [58, 57] than that obtained from a conventional compliance trace, which would obviously add to the difficulties when making comparisons between sets of published data.

Very few offset displacement vs load traces are presented in the literature [49, 56, 57]. Generally, however, the traces change shape as ΔK changes, with three distinct regions visible for high ΔK, see Fig. 2.13. As ΔK decreases the plastic portion disappears, and in the near-threshold region, the closure portion K_{op}-K_{min} becomes dominant, i.e. the closure value increases. Robin and Plumbridge [59] suggested that the disappearance of the plastic stage is related to a modification in the propagation mechanism caused by a transition from stage II growth (regime B of Fig. 2.1) to stage I growth (regime A of Fig. 2.1).

Closure measurement is further complicated by hysteresis in the closure transducer signal believed to be due to plastic deformation at the crack which causes P_{op} and P_{cl} to be different. See Figure 2.14. Numerical analyses indicate that this difference may be as large as 14% [60], with P_{op} always being greater than P_{cl} [20]. This is supported by experimental results [1]. In most situations the difference is within the experimental scatter of the measurement technique [11].

2.2.2.2 Technique resolution

As mentioned earlier, another difficulty encountered when measuring closure is the relatively low sensitivity of the techniques available. Typical fracture surface asperities (say less than 3 μm in height) are likely to be important only if they are close to the crack tip (of the order of 1-50 μm) [8, 38, 61]. For the mechanisms of oxide induced or roughness induced closure, significant contact probably occurs within 25 μm of the crack tip, which is at the resolution of conventional closure measuring techniques. Although Walker and Beevers [44] reported that the acoustic technique was capable of detecting a crack increment of 10 μm, the possibility of misleading closure indications, as discussed earlier, makes the use of this technique unreliable.

The limit in resolution of closure measuring systems is illustrated by

the results from Gan and Weertman [62] using crack tip strain gauges. They measured closure and growth rate after an overload was applied. Although the growth rates decreased immediately, an increase in closure was only detected after 0.05 to 0.1 mm of growth as shown in Fig. 2.15.

Another cause of discrepencies between various closure results may come from the fact that most techniques require the cyclic loading frequency to be reduced to between 0.02 and 0.5 Hz while measurements are being made, while fatigue cycling frequencies are generally greater than 20 Hz. Both Buck et al [55], using acoustic emission techniques, and Kikukawa et al [57], using compliance methods, found that a number of such cycles were required before a stable load displacement trace was produced. It has been suggested [63] that at low frequencies both reverse slip due to residual plasticity and creep of asperities may result, which would therefore require a few cycles to stabilise. Presumably, the initial closure trace would accurately represent the closure which occurred during fatigue cycling.

Erroneous closure results may also result from a highly curved crack front [63].

In view of the difficulties associated with closure interpretations alternative parameters have been suggested. Williams and Stoffer [65] have used a contact stress parameter K_c which is a measure of the crack face contact stress, with which they were able to correlate mean stress and overload effects for two aluminium alloys. However, measuring the contact stress was very tedious and measurements for growth rates below 10^{-5} mm/cycle were not considered, which is the region in which closure is most significant and where discrepencies in growth rate predictions are maximised.

Another parameter which has been suggested [45], is the crack mouth opening displacement (CMOD) which has a smaller range than that corresponding to ΔK if closure does occur, but larger than that corresponding to ΔK_{eff} shown in Fig. 2.16. The CMOD must be normalised as it changes with crack length for the same applied stress intensity. Walker and Beevers managed to eliminate a discontinuity and the R ratio effect using this parameter in a log da/dN vs log σ plot. They also observed a direct relationship between σ and ΔK which is a relationship that is not predicted by LEFM, however, it is an empirical relationship which works.

2.3.2.3 Compliance gauge locatio

Measured closure values appear to be independent of the crack length, except when the crack tip is less than 2.5 mm from the notch tip [37, 66, 67]. Fleck [37] correlated closure measurements with growth rate measurements, for various compliance gauges and found crack tip strain gauges to give the most accurate closure response for crack lengths less than 2.5 mm measured from the notch tip. The accuracy of crack tip strain gauges is, however, dependent on location and they have a disadvantage in that they can therefore only monitor closure over a small crack growth increment [37]. In addition, when positioning the near tip gauges, care must be exercised so that they are not too close to the crack plane because of possible plasticity effects [11], nor too far as they lose sensitivity, (ie. within 1 mm [68]).

Closure values do vary slightly with compliance gauge type, with surface strain gauges generally exhibiting higher closure levels than BFS or crack mouth clip gauges [57, 68]. This is not an inherent difference between the various techniques but rather occurs because closure (plasticity induced) at the surface is higher due to the existence of plane stress conditions. Confirmation of such an effect was deduced from the results of Sunder and Dash [3], which showed smaller striations in the near surface region compared to the central regions, implying a lower ΔK_{eff} at the surface, ie. greater closure.

In summary therefore, both the measurement of closure and the interpretation of the experimental data is not a trivial exercise. It is, however, a necessary one if the influences of this important effect are to be determined.

2.4 Fatigue Phenomena Explicable in Terms of Closure

The importance of closure to fatigue crack growth lies in its ability to shed significant light on many of the complex phenomena relating to both loading conditions and microstructure. Some of the effects explicable in terms of closure are detailed in the following sections.

2.4.1 Near threshold growth

The concept of closure has been widely used to rationalise the large differences observed in threshold data for similar materials under similar testing conditions. Various authors have proposed an intrinsic threshold value K_{in} which is independent of closure such that the measured threshold value $\Delta K_{th} = \Delta K_{in} + K_{op}$ [69].

In this case measured threshold data could vary with any factors that affect closure, e.g. microstructural variations and environmental sensitivity. Generally, K_{op}/K_{max} increases as ΔK approaches ΔK_{th} [31, 63, 70-73], and some authors have reported $\Delta K_{eff,th}$ (= ΔK_{eff} at threshold) to be almost zero [57,28]. The increase in closure at threshold probably occurs because of the other closure mechanisms, namely roughness and oxide induced closure which become operative when the CTOD is small, (see section 2.2.2 and 2.2.3). The relevance of these mechanisms to near-threshold growth rates was illustrated by numerical analyses performed by Newman [20] and Blom and Holm [31] which showed that plasticity induced closure, modelled for plane strain, was unable to explain the high closure obtained at the threshold.

However, high near-threshold closure values have also been attributed to increased plasticity induced closure which depends on the load shedding routine used to reach threshold [18, 74, 75]. If load shedding is rapid i.e., load decreases are large, there will be large plastic zones close behind the crack tip which will increase the closure load [76] and lead to a conservative ΔK_{th}.

If, however, roughness induced closure is large enough to equal, or exceed, the plasticity induced closure resulting from the load shedding scheme, the details of the process will have a negligible

effect on near-threshold growth, as the near-tip asperities will then dominate closure values. James [63] used this idea to explain why wake removal to within 0.5 mm of the crack tip caused a reduction in closure for a fine grained quenched and tempered steel (low surface roughness) while no reduction in closure was apparent for a coarse grained material (high surface roughness).

2.4.2 Mean stress and closure

Fatigue crack propagation rates in various materials show widely differing sensitivities to the influence of mean stress (stress ratio). In general the degree of sensitivity to mean stress effects is low for mild steels, higher for high strength steels and even greater for aluminium alloys [77]. Thus for aluminium increasing the R ratio results in a significant increase in growth rate and a corresponding reduction in ΔK_{th}.

The effects of closure are most evident at low stress ratios R (= K_{min}/K_{max}) where the absolute value of closure load is high relative to the maximum stress intensity. As R increases the closure contribution decreases and in this way the R effect is at least qualitatively explained in terms of closure.

An example illustrating the extent to which the R ratio effect depends on closure comes from work on a nickel-base superalloy IN 9021 with a very fine grain size of 0.2 μm. This alloy exhibited no closure and correspondingly no R ratio effect (the material produced a very low fracture surface roughness which was probably the reason why it exhibited no measurable closure [24, 25]). The fracture surface for the very fine grain sized material is shown in Fig. 2.17 together with a rougher fracture surface corresponding to a slightly coarser grained material which exibited significant closure.

The significance of the stress ratio and closure effect on fatigue life predictions is well illustrated by cracks in welded material, where residual (tensile/compressive) stresses of the order of the yield stress may exist. Measurements of residual stress distribution in centre cracked tension (CCT) transverse butt-weld joint specimens, revealed that high tensile residual stresses are always induced at the crack tip as the stresses redistribute themselves as the crack extends [75, 78]. Thus closure may be absent throughout the whole fatigue life for certain component shapes and growth may proceed at identical

rates for R = 0 and R = 0.8.

Where closure is responsible for R ratio effects on growth rate, the growth rate data can often be rationalised by the use of ΔK_{eff} [2, 31, 73, 79, 80]. Typical data showing how ΔK_{eff} reduces growth rates onto a single curve at various R ratios is shown in Fig. 2.18.

However, many cases have been reported where closure is only capable of partially explaining R ratio effects, [26, 37, 59, 71, 81] as seen in Fig. 2.19. This discrepency could arise if:

1. ΔK_{eff} is different to the real effective stress intensity range because either the measurement system has insufficient resolution to measure closure very near the crack tip (real ΔK_{eff} is lower), or because strain intensification occurs below K_{op} (real ΔK_{eff} is greater).
2. Environmental effects occur, such as hydrogen embrittlement (due to moisture in the atmosphere).
3. The imposition of static fracture modes onto the baseline fatigue crack growth rates, caused by the high K_{max} attendant on high R ratios [77].

In most cases, however, the use of a closure-based ΔK_{eff} affords a significant reduction in the R ratio effect on growth rates.

2.4.3 Fatigue crack retardation

The reduction in growth rate immediately following a tensile overload has frequently been attributed to closure effects [6, 68, 82]. Although other mechanisms causing crack growth retardation have been proposed, only closure is able to explain the fact that growth rate reaches a minimum only <u>after</u> some crack extension. This delayed retardation arises because some growth is required before the overload deformation zone becomes part of the crack wake and then causes the increase in closure [11, 68, 83]. Closure is also the only mechanism which can explain why retardation is not limited to the plastic overload zone, but occurs over a distance several times the size of the overload plastic zone [82]. The effectiveness of the overload plastic zone in causing a drop in growth rate (and an increase in closure,) is clearly illustrated in work by Trebules [84]. He performed a constant ΔK fatigue propagation test, then applied an overload following which growth rate was measured to have decreased significantly. The crack wake including the overload zone was then

machined off after which growth rates commenced at the same rate as that immediately prior to the application of the overload [84].

Surface effects in the form of plasticity induced closure appear to play a very important role in retardation following tensile overloads. For example Shin and Wei [85] observed crack growth to reinitiate at the mid-thickness region of a specimen after the application of a tensile overload. Further, Davidson [83] observed that only the surface regime in a 6061 T6 aluminium alloy, showed zones of intense plastic deformation, following a 100% overload = 21 MPa√m. Under such conditions it is probable that crack growth in the interior is favoured. This idea is also endorsed by tests using side grooves, which ensure plane strain conditions at the specimen surface (triaxial stresses), and consequently results in a much smaller retardation effect than for specimens without side grooves.

Despite the success of closure based arguments in explaining crack retardation following tensile overloads, various results cannot be explained in terms of closure [86]. Alternative mechanistic arguments which have been suggested [87] to account for transient effects following overloads involve concepts based on:

1. Crack tip blunting.
2. Residual compressive stresses close to the crack tip (e.g., reverse cyclic plastic zones.)
3. Crack tip strain hardening.
4. Non planar crack growth and crack branching.

However, immediate retardation as opposed to delayed retardation are predicted by all the above mechanisms, which does not generally equate to the experimental observations.

Equally, although one mechanism may dominate in reducing growth rate, generally a combination of the above mechanisms (including closure) will be responsible for the retardation. Interaction of the above mechanisms was observed by Corlby and Packman [88] and Shin and Fleck [89], who found crack growth rate actually accelerated following a tensile overload before retardation to below the pre-overload level. This was possible because of crack tip blunting effects which separate the crack faces and therefore reduce closure.

The role of non planar crack growth and crack branching noted as point 4 above is significant and its effects clarified by the following:

1) Crack deflections of the order of 30°C, may reduce the effective driving force by 14% [87].
2) Crack branching reduces the driving force as two cracks require more energy to propagate than one.

Thus, for situations where crack deflections, crack branching, or significant mode II displacement exists, closure may only be a contributary mechanism responsible for crack growth retardation.

Compressive loads following tensil verloads have been shown to reduce the effects of tensile overloads significantly [23, 85]. The mechanism by which such compressive overloads reduce retardation is not known exactly, but presumably relates to 1) a reduction in crack tip blunting, and 2) a reduction in closure, possibly by crushing and compacting the asperities.

2.4.4 Microstructural effects on closure

Near threshold growth is generally sensitive to microstructural influences because of the small crack tip plastic zone size relative to the dominant microstructural dimensions. Thus crack propagation occurs along the most suitably oriented slip planes which results in out of plane growth [86], and leads to extensive facet formation at low growth rates [86].

The dominant microstructural influences in aluminium alloys are the grain size, the grain orientation with respect to the crack growth direction (for pancake shaped grains), and the ageing condition [90]. For steels the dominant microstructural dimensions might include the prior austenite grain size, the pearlite colony size and the inter lamellar spacing [21], and bainite/martensite packet size.

The influence of grain size on ΔK_{th} has even been expressed in the form of a Hall Petch relationship:

$$\Delta K_{th} = \Delta K_{th(0)} + kd^{1/2} \qquad [61]$$

where $\Delta K_{th(0)}$ is the threshold stress intensity without closure, k is a constant and d is the diameter of the grain size.

Microstructrual effects are generally manifested in the fracture surface roughness which in turn alters the closure response and

therefore the growth rate. Closure has indeed been found to increase with grain size for low carbon steels [91] an α/β titanium alloy [92] and nickel based superalloys IN901 and APK -1 [93]. Closure is thus often able to account for microstructural variations as evidenced in results where the effects of grain size variations are eliminated at high R ratios [21].

Closure is not always able to correct for grain size effects, as was illustrated by the results of [94]. They measured a 100 fold increase in grain size which caused a 2 fold increase in ΔK_{th} for both R = 0.1 and 0.8. Although closure effects could explain the increase in ΔK_{th} at R = 0.1, closure is not present for R = 0.8, thus the intrinsic threshold must also have increased.

For aluminium alloys closure increases with grain size and with a reduction in the ageing condition, both of which result in an increase in the crack path tortuosity and fracture surface roughness [95, 96]. This is explained by the fact that larger grains encourage planar slip over longer crack lengths [97]. Equally the underaged microstructure has no or few non-coherent particles (which are present in the peak-aged and overaged structures) to interrupt slip, and thus encourages planar slip and slip reversibility as little interference exists.

Differences in crack path tortuosity may not necessarily be reflected in a higher closure contribution, but it may nevertheless cause a reduction in the crack tip stress intensity because a combined mode I and mode II stress intensity distribution is effectively less than that due to mode I loading alone (see section 2.2). Thus in some cases where tortuosity differs between microstructures, the materials show intrinsic differences in growth rate, as shown by results at R = 0.8 where no closure contribution was found to exist [13, 26].

In limiting cases the overaged structure may exhibit a slower growth rate and greater crack path tortuosity than for the underaged alloy, this may be explained by the occurance of cross slip which again encourages planarity of slip [27, 86].

2.4.5 Crack growth under compressive loading

In section 2.4.2 it was mentioned that compressive loads immediately after tensile overloads reduce the delay effect caused by the tensile overload. Generally compressive loading acts in two ways:

1. The asperities are crushed to reduce closure, and
2. The crack tip develops residual tensile stresses.

The latter mechanism is effective enough to produce crack growth under fully compressive cyclic loading provided there is a suitable initial notch [98] to allow residual tensile stresses to develop [23, 99]. Such growth is limited to approximately the compressive plastic zone size indicated by the Irwin equation $r_p = 1/(2\pi)\left(K_{max}/\sigma_y\right)^2$ (which is not strictly valid, because it is applied at the notch and because loading is in compression, but it is useful). The fact that growth occurs only within the confines of the initial notch plastic zone under compressive loading is shown by the concave crack front obtained by Reid et al [100] shown in Fig. 2.20. Growth at the surface was about 9 mm, but only 2 mm in the interior of a 25 mm thick mild steel compact tension (CT) specimen. These values correlate well with the plastic zone sizes which were about 8 mm in plane stress and 3 mm in plane strain.

The sequence of events causing growth under compressive loading may be made clearer by considering the results of Yu et al [98] on a 2024 T351 aluminium alloy. Cyclic loading from -207 MPa to -25 MPa produced 2.04 mm of growth before arrest. Increasing the maximum stress from -25 to -12 MPa added a further 1.41 mm of growth, and a further increase in the maximum stress to -1 MPa caused an additional 3.9 mm of growth, (total crack length = 7.36 mm). Essentially the situation can be simplified by considering that the crack tip is open above a certain compressive load, because of the residual tensile stresses. (These tensile stresses develop by a similar mechanism to that of the reverse plastic zone under conventional tensile loading, as shown in Fig. 2.21a.)

Thus as the globally applied load increases from K_{op} to K_{max}, the crack tip experiences a range of residual tensile stresses which can cause crack growth. As the crack moves away from the notch tip the range of tensile stresses decreases (because K_{op} tends to K_{max} as the crack length increases) and the crack arrests somewhere near the

elastic/plastic interface. Hence, it may be expected that growth under compressive loading is limited to a length lying somewhere between the sizes of the reverse plastic zone and the monotonic plastic zone. Although it may be expected that growth under compression would occur only until the end of the reverse plastic zone (as that is the limit of the residual tensile stresses) the crack propagates beyond this value probably because the reverse plastic zone always moves along with the crack tip, and eventually the monotonic plastic zone size will limit the growth, as shown in Fig. 2.21b.

Growth under compressive loading is useful for obtaining values of ΔK_{th} independent of load shedding effects and for precracking of fracture toughness specimens in brittle material. With respect to threshold testing, Suresh et al [23] suggested that there was no plastic damage ahead of the tip of the crack which had arrested during compressive cycling. This suggestion was based on the fact that subsequent growth at an R ratio of 0.75 produced a threshold value (3.0 MPa$\sqrt{m}$) and growth rates identical to those from conventional load shedding sequences. Hence the technique can yield information regarding the intrinsic, or closure-free, threshold at high R. However, their data for R = 0.1 indicated that growth initiated at ΔK = 5.5 MPa$\sqrt{m}$ while ΔK_{th} obtained under conventional load shedding was 7.7 MPa$\sqrt{m}$. In this case, however, growth did not commence at ΔK = 3 MPa$\sqrt{m}$ which was the intrinsic material threshold for the alloy condition tested. This implies that unknown effects have an influence at low stress ratios.

A few authors [23, 98] have made closure measurements in an endeavour to determine at what loads the crack tip is open under compressive loading. It had been expected that the open load range in compression would produce a growth rate similar to that obtained under the corresponding open load range in tensile loading, however the results of Suresh et al and Yu et al [98] did not show this. Possible reasons for this discrepency may include :-

1. Inaccurate closure measurements, during compressive or tensile loading, due possibly to plasticity effects (very high under compressive loading), or the sensitivity of the closure measurement technique.
2. Growth under compressive loading does not appear to be controlled entirely by the "open" load range, and there appears to be strain intensification below K_{op}, (see section 4.3).

However, in one case, where opening loads were calculated [23]

numerically the numerical values compared well with those obtained experimentally, suggesting that the closure measurements were accurate (based on plane strain measurements), as shown in Fig. 2.22. Further calculations by Blom et al [101] using finite element analysis (F.E.A.) based on opening loads in compression, enabled accurate predictions of experimentally obtained crack growth, which suggests that strain intensification may only occur under certain conditions.

2.4.6 The effect of time on closure

It is well known that the reverse slip of dislocations and creep of asperities are time dependent processes [63]. It is not surprising, therefore, that closure which originates from plastically deformed material and asperity tip contact has been found to vary with time [63]. In this study, Lei et al found that closure values first decreased and then increased to a level above the original K_{op} value, eventually achieving a steady state value. For a lower yield strength material, the variation of K_{op} with time was observed to be more rapid than for a higher strength material, which supported the premise that dislocation reverse slip and creep of asperities was in fact the mechanism whereby K_{op} changed, as slip and creep are expected to be more rapid in a lower strength material (at a given temperature and applied stress).

Thus a certain amount of uncertainty associated with closure may be attributed to changes in closure due to time effects.

2.5 Short Crack Growth

Short crack growth has attracted much interest in the last decade since Pearson [102] and subsequently many other authors [103-105] observed that short cracks could grow significantly faster than long cracks under nominally identical stress intensity driving forces and, at values of stress intensity range well below the threshold. Such growth is illustrated in Fig. 2.23. Indeed, a number of service failures, particularly in the aircraft and power generation industries have involved the growth of short cracks [106], where life predictions based on long crack growth rate data have resulted in erroneously high life estimates.

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