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THE RELATIONSHIP BETWEEN

Fracture Toughness

AND

Shear Lip Size

Thesis submitted for the degree of

Doctor of Philosophy

At the

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By

Man-On Lai

1977
ABSTRACT

The plane strain fracture toughness of a high strength low alloy steel, En 25, tempered from the as-quenched state to 650°C, and a high strength aluminium alloy, 7075-T6, have been determined using compact tension specimens of varying thicknesses from 4 to 25mm. The size of the shear lips, $B_{SL}$, measured at the free surfaces of the specimens was found to be independent of the specimen thickness and to be related to the true plane strain fracture toughness, $K_{IC}$, through the expression

$$B_{SL} = 0.41 \left( \frac{K_{IC}}{\sigma_{YS}} \right)^{2.02},$$

where $\sigma_{YS}$ is the 0.2% proof stress of the material. A rationale for this behaviour is that $B_{SL}$ is approximately equal to the size of the plane stress plastic zone, $r_y$, at the surfaces of a plate specimen, which from theoretical analysis, has been shown to be

$$r_y = \frac{\pi}{8} \left( \frac{K_{IC}}{\sigma_{YS}} \right)^2.$$

The ASTM standard plane strain fracture toughness test method has been proven to be insensitive to detect excess yielding in the specimens tempered at temperatures higher than 450°C. At high tempers, yielding fracture mechanics approaches were used to determine the true $K_{IC}$ values.

The fracture toughness versus shear lip size relationship is believed to have considerable importance in the analysis of service failures and in the preliminary study of specimen size effect in plane strain fracture toughness determination.
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INTRODUCTION

When a plate with a through-the-thickness crack is subjected to a tensile load, the elastic stress intensity at the vicinity of the tip of the crack can be very high. The intense stresses would be expected to produce plastic deformation when the appropriate yield criterion is satisfied. Consequently, a plastically deformed region called the plastic zone is formed around the crack tip. The size of the plastic zone, \( r_y \), is estimated to be \( (1) \),

\[
r_y = A \left( \frac{K_c}{\sigma_y} \right)^2
\]

where \( K_c \) is the critical stress intensity factor, \( \sigma_y \) is the 0.2% proof stress of the material and \( A \) may be taken as \( 1/2\pi \) for plane stress conditions and \( 1/6\pi \) for plane strain conditions. A simplified three dimensional schematic representation of the crack tip plastic zone therefore consists of a cylinder and two cones as shown in Fig.1.1. The bases of the cones are on the surfaces of the plate specimen while the vertices point to each other. The two cones which represent the plane stress plastic zone on the specimen surfaces are connected by a cylinder which represents the plane strain plastic zone at the interior of the specimen.

Consider the fracture surface of the plate after it has been fractured apart. At the centre of the plate where plane strain stress state dominates, as in the case of a thick plate, square fracture occurs while at the plate surfaces, the stress state changes to plane stress because of the diminishing effect of the stress acting in the plate thickness direction. The failure of the plane stress regions gives rise to the slant fracture generally referred to as the shear lips. The cross-sectional profile of such a mixed mode fracture is shown in Fig.1.2.

In fracture toughness testing, it has been observed that the temperature dependence of the material fracture toughness is approximately coincident with the temperature dependence of the amount of shear lip.
or the absolute size of the shear lip, $B_{SL}^{(3)}$.

This would seem to suggest that a relationship exists between the material fracture toughness and the shear lip size. Since the formation of the shear lips is a consequence of the plane stress fracture mode, comparing the formal representation of the crack tip plastic zone in Fig.1.1 with the cross-sectional profile of a mixed mode fracture shown in Fig.1.2, it is indeed tempting to suppose that the shear lip size is at least, to a first approximation, equal to the depth of material at the crack lip which is in a plane stress stress state. Referring to equation 1.1, the following relationship would therefore hold:

$$B_{SL} = A \left( \frac{K_c}{\sigma_{ys}} \right)^2$$

That is, the size of the shear lip, $B_{SL}$, is proportional to the square of the relative toughness, $(K_c/\sigma_{ys})$. The constant $A$ may be taken as $1/2\pi$ as before.
A relationship like equation 1.2 would provide a useful tool for failure analysis. Instead of estimating the fracture toughness of a failed component using the cumbersome conventional standard testing, it could by means of the above equation be obtained by measuring the size of the shear lip, $B_{SL}$, and the yield stress, $\sigma_{YS}$. The latter could perhaps be found from a hardness test.

The basic aims of the present investigation were:

1) To determine the effect of tempering temperature on the size of the shear lips and the material fracture toughness of a commercial low-alloy high-strength steel.

2) To determine the effect of specimen thickness on the size of the shear lip and the material fracture toughness.

3) To relate the size of the shear lip to the relative toughness of the material.

Using compact tension specimens of thicknesses varying from 4mm to 25mm, the above aims were examined with a locally available 2.5% Ni-Cr-Mo steel (also called Comsteel En25) tempered from the as-quenched state to a temperature of 650°C and a high strength 7075-T6 aluminium alloy.
CHAPTER 2

FRACTURE TOUGHNESS - CONCEPT AND TESTING
FRACTURE TOUGHNESS - CONCEPT AND TESTING

2.1. BASIC CONCEPT OF FRACTURE TOUGHNESS

It was stated by Griffith (4) that crack propagation will take place if the energy released upon crack growth is sufficient to provide all the energy required for crack growth. Consider an infinite cracked plate of thickness B with a central transverse crack of length 2a, subjected to an applied stress of \( \sigma \) at infinity (Fig. 2.1).

![Diagram of a cracked plate under tensile loading](image)

**Fig. 2.1:** Crack subjected to tensile loading

The total stored energy, \( U_t \), of the system with the crack can be shown to be (5),

\[
U_t = \frac{(k+1)\pi a^2 B^2 \sigma^2}{8\mu} + U_o \quad \text{........................................... 2.1}
\]

where \( k \) is equal to \( (3-4\nu) \) for plane strain stress state and \( (3-\nu)/(1+\nu) \) for plane stress stress state, \( \nu \) is the Poisson's ratio, \( \mu \) is the shear modulus and \( U_o \) is the stored energy of the plate without the presence
of the crack. The term \((k+1)\pi a^2 B_0^2/8\mu\) is the decrease in stored energy as a result of the crack and is equivalent to that computed from the stress analysis of Inglis\(^6\).

The total surface energy of the crack is,

\[ U_s = 4a\gamma \]  \hspace{1cm} 2.2

where \(\gamma\) is the surface energy per unit area.

When the crack grows by increasing its length by a small amount \(2\delta a\), the stored energy will be changed by

\[ \delta U_t = \frac{(k+1)\pi a B_0^2}{4\mu} \delta a \]  \hspace{1cm} 2.3

Similarly, the change in surface energy is

\[ \delta U_s = 4\gamma \delta a \]  \hspace{1cm} 2.4

Crack growth will be unstable if

\[ \delta U_t > \delta U_s \]  \hspace{1cm} 2.5

But at the threshold of instability, \(\sigma\) becomes the fracture stress, \(\sigma_f\), which can be evaluated, from equations 2.3, 2.4 and the Griffith energy criterion for fracture\(^4\), to be

\[ \sigma_f = 4\sqrt{\mu \gamma / \pi a (k+1)} \]  \hspace{1cm} 2.6

Under plane strain conditions, equation 2.6 is reduced to

\[ \sigma_f = \sqrt{2\mu \gamma / \pi (1-\nu^2)} \]  \hspace{1cm} 2.7

and under plane stress conditions,

\[ \sigma_f = \sqrt{2\mu \gamma / \pi a} \]  \hspace{1cm} 2.8

Equation 2.6, however, is only applicable to truly brittle materials like glass in which the energy for crack growth is approximately the surface energy to form new surfaces. In the case of metals, plastic
deformation occurs at the tips of the crack and energy is expended in forming yielded zones at these locations. It is hence necessary that for the crack to grow, more energy than the surface energy is required. To accommodate this plastic deformation, it was suggested (7-9), to modify the Griffith analysis above by replacing \( \gamma \) by \( (\gamma + \gamma_p) \), where \( \gamma_p \) is the energy of plastic deformation absorbed by the fracture process. Irwin and Kies (10) considered that the modified theory involved finding the instantaneous value of the rate of supply of available energy. This energy rate takes the form (10)

\[
\frac{\partial U}{\partial \alpha} = \frac{1}{2} P^2 \frac{\partial C}{\partial \alpha} \quad \text{(2.9)}
\]

where \( P \) is the applied load and \( C \) is the compliance of the specimen.

In equation 2.9, \( \frac{\partial U}{\partial \alpha} \) is denoted by the symbol \( G \) after Griffith and is known as strain energy release rate. Irwin (11) regarded \( G \) as a force, and hence \( G \) is also known as the crack extension force. At the critical value when the crack starts to propagate, the stored elastic strain energy and the energy absorbed by crack extension are equal. \( G \) at critical condition is denoted by \( G_c \), generally referred to as the critical strain energy release rate or critical crack extension force, or, simply, the fracture toughness. With this term \( G_c \), equations 2.7 and 2.8 can be rewritten as

\[
\sigma_F = \sqrt{\frac{EG_c}{\pi a}} (1-v^2) \quad \text{for plane strain and} \quad \text{(2.10)}
\]

\[
\sigma_F = \sqrt{\frac{EG_c}{\pi a}} \quad \text{for plane stress} \quad \text{(2.10)}
\]

By considering Westergaard's result (12) (see section 3.1) in which the elastic stresses at the vicinity of the tip of the crack is analysed, it can be shown that the strain energy release rate, \( G \), is related to the stress intensity factor, \( K \), through the expression (11)

\[
G = \frac{(1 + k)}{8\mu} K^2 \quad \text{(2.11)}
\]

Under plane stress and plane strain conditions, equation 2.11 becomes

\[
\begin{align*}
K^2 &= EG \\
\text{and} \quad K^2 &= \frac{EG}{(1-v^2)} \quad \text{(2.12)}
\end{align*}
\]

respectively.
Combining equations 2.10 and 2.12, therefore, \( K = \sigma \sqrt{\pi a} \) .............. 2.13

At critical condition, \( K \) is denoted by \( K_c \) and is often referred to as the fracture toughness. Since \( K \) is a parameter which depends on the applied stress, it has the advantage over \( G \) in that its evaluation in terms of the applied stress and specimen dimensions is a normal linear elastic stress analysis problem. As long as an expression for \( K \) can be determined, experimental value of \( K_c \) can be obtained from recording the applied stress and crack length at critical condition. Methods to derive the expression for \( K \) are well documented in the work of Paris and Sih\(^{(13)}\) and recently reviewed by Cartwright et.al.\(^{(14)}\). Some of these expressions for different practical specimen geometries are found in References\( (13,15-17) \).

2.2 EFFECT OF PLASTIC YIELDING AT CRACK TIP

In the singular field immediately ahead of a crack tip (Fig.2.1), the elastic stresses are \( (12,13,18) \),

\[
\begin{pmatrix}
\sigma_x \\
\sigma_y \\
\tau_{xy}
\end{pmatrix}
= \begin{pmatrix}
\frac{\sigma}{\sqrt{2r}} \cos \frac{\theta}{2} \\
1+\sin \frac{\theta}{2} \sin \frac{3\theta}{2} \\
\sin \frac{\theta}{2} \cos \frac{3\theta}{2}
\end{pmatrix}
\]

\[
\begin{align*}
\sigma_z &= 0 \ldots \text{plane stress} \\
\sigma_z &= \nu (\sigma_x + \sigma_y) \ldots \text{plane strain}
\end{align*}
\]

where \( r \) and \( \theta \) are the polar coordinates of a point in the vicinity of the crack tip and \( \nu \) is the Poisson's ratio. The distribution of the stress \( \sigma_y \) as a function of \( r \) at \( \theta = 0 \) is shown in Fig.2.2.

![Fig.2.2](image)

**Fig.2.2:** Elastic stress distribution at the crack tip.
It is observed that when $r$ is small, the stress tends to infinity. In practice, this is not possible because plastic yielding will occur at the crack tip to keep the stress finite. A plastic zone will therefore be created. Provided that the size of this yield zone is small in comparison to the length of the crack, as well as the net remaining cross section, the linear elastic analysis above will be invalidated.

It was suggested that the presence of the plastic zone gave rise to an increase in the effective length of the crack. Therefore, to accommodate to some extent the yielding ahead of the crack, Irwin (19) postulated the use of an effective crack length of $(a + r_y)$, where $r_y$ is the radius of the plastic zone given by

$$r_y = \frac{1}{2\pi} \left( \frac{K_c}{\sigma_{ys}} \right)^2 \quad \cdots \quad 2.15$$

with $\sigma_{ys}$ being the yield stress of the material.

From equation 2.13, the modified stress intensity factor under plane stress conditions becomes

$$K^* = \sigma_{app} \sqrt{\pi (a + r_y)} \quad \cdots \quad 2.16$$

where $\sigma_{app}$ is the applied stress.

Equation 2.15 can be obtained by combining equations 2.13 and 2.14 with $\sigma_y = \sigma_{ys}$. However, $r_y = \frac{1}{2\pi} \left( \frac{K_c}{\sigma_{ys}} \right)^2$ is only applicable to condition under plane stress. Because of the effect of constraint in restricting plastic flow, the plane strain plastic zone size is smaller than that in plane stress. A factor of one-third reduction has been suggested (1). A more detailed description of the size and shape of the crack tip plastic zone will be taken up again in Chapter 3. Nevertheless, it is convenient to give here the formal representation of the plastic zone at the tip of a through-the-thickness crack in a plate specimen as shown in Fig.2.3 (20).
2.3 EFFECT OF SPECIMEN THICKNESS

In measuring the fracture toughness of a plate specimen of a given high-strength material at a given temperature, it has been found that the measured values of $K_c$ or $G_c$ are strongly dependent on the specimen thickness (21-25). This thickness dependency is related to the change in the appearance of the fracture surface. In somewhat thin specimens, $45^\circ$ slant fractures are observed along the free surfaces while along the central region, the fracture is square or flat. Fully slant fracture results as the thickness is reduced. On the other hand, the proportion of square fracture increases when the plate thickness is increased. This is followed by a decrease in the value of $K_c$ as shown in Fig.2.4. Although the exact shape of the curve is influenced by work hardening exponent, yield strength and the toughness level of the material (26), similar curves have been observed (21-24, 27-29) for other engineering materials. A limiting value of $K_c$ is observed when the specimen thickness is increased to a state of complete plane strain. $K_c$ value measured under this condition is denoted by $K_{IC}$ (or $G_{IC}$) where the subscript I denotes the opening mode (see Fig.2.5). $K_{IC}$, generally referred to as the plane strain fracture toughness, is considered a material constant. The value of $K_{IC}$ is of special importance
because it represents the lower limit to the fracture toughness of a material at a given condition.

Fig. 2.4: Dependence of $G_c$ and fracture appearance on specimen thickness ($30\%$).

Fig. 2.4 shows that there is an optimum thickness $B_o$ where $K_c$ has a maximum value generally regarded as the true plane stress fracture toughness.

Fig. 2.5: The three modes of cracking

Fig. 2.6: Types of load/displacement records
To examine the variation of fracture toughness with plate thickness, Fig.2.4 is divided into three regions for convenience. The corresponding schematic load/displacement records are shown in Fig.2.6. When the plate thickness is small, as in region A, fracture toughness increases linearly with the increase in plate thickness \((24,31,32)\), although roughly constant fracture toughness values have also been observed \((33)\). Because of the lack of constraint, the stress in the thickness direction tends to zero, giving rise to a stress-state that is virtually plane stress. This stress-state is illustrated in Fig.2.7(a) using Mohr's Circle analysis. It is noted that the maximum shear stress, \(\tau_{\text{max}}\), acts on the planes rotated \(45^\circ\) from the directions of \(\sigma_1\) and \(\sigma_3\).

These planes subtend an angle of \(45^\circ\) with both the plate surfaces and the loading direction. Since plastic deformation is a result of shear stresses, the plate fails by shear rupture on a \(45^\circ\) slip plane. This mode of separation generally involves both the opening mode and the tearing (or antiplane strain) mode \((31,35)\). The first stage of crack propagation from the starter notch is confined to the mid-thickness of
the plate. This cracking, usually triangular in shape, is a square type fracture with its plane normal to the loading direction (Fig. 2.8(a)). The 45° slant fracture along the plate surfaces progressively grow until at some point the square fracture disappears completely. Thereby, the entire fracture is of the slant type.

![Diagram](image)

(a) Fully developed slant fracture

(b) Predominantly square fracture

![Diagram](image)

**Fig. 2.8:** Schematic drawing of stages of crack propagation.

The occurrence of the initial square fracture does not give rise to instability in the plate because the developing slant fractures require increasing strain and load before separation can take place. However, it produces a gradual decrease in slope in the load/displacement record as shown in Fig. 2.6(a). (The non-linearity is also partly due to crack tip plasticity). As the energy per unit area required to form slant fracture increases at the same rate as the plate thickness, $K_c$ or $G_c$, in region $A$ of Fig. 2.4, increases approximately linearly from very thin sheet to a maximum level at thickness $B = B_0$.

In the case of very thick specimen, consider a notched plate being loaded in tension. The high longitudinal stress at the root of the notch causes the material there to contract. Due to the constraint of the elastic material at the notch, this contraction sets up a trans-
verse tensile stress, \( \sigma_z' \) in the thickness direction \(^{(39)}\). \( \sigma_z \) is maximum at the centre of the plate while at the surfaces, it drops to zero as the plate surfaces are not loaded externally. Hence the centre of the plate is under triaxial or plane strain stress state and the surfaces, biaxial or plane stress stress state. With specimen thickness in region C of Fig.2.4, the surface effect is small and can be neglected.

It can be noted from Fig.2.7(b) that the planes of maximum shear stress under plane strain situation are different from those of plane stress \(^{(40)}\). The plane strain stress state, as illustrated by the Mohr's circle analysis, produces shear planes that are perpendicular to the specimen surface but make an angle of about \( 45^\circ \) with the direction of crack propagation \(^{(35,41)}\). Such a slip system is known as hinge-type deformation \(^{(32)}\). However, this process may not be favourable if the shear stress in plane strain condition is low \(^{(34)}\).

If the condition of stress approaches idealized plane strain in very thick plate, the initial instability can cause the entire specimen to fracture apart. The load displacement record then will look like that shown in Fig.2.6(c). The fracture surface of such a specimen will consist of so great a proportion of square fracture that the contribution of the \( 45^\circ \) slant fracture at the surfaces will not be detectable.

However, it is observed that many materials do not fail catastrophically at initial instability, instead macroscopic slow crack growth occurs first before failure \(^{(42,43)}\). In these specimens, initial instability in the form of square type fracture is initiated at the notch root in the central low toughness plane strain region. This sudden initial extension of the crack is often referred to as pop-in \(^{(43)}\). The fracture moves forward so that the leading crack front is U or semi-circular in shape as shown in Fig.2.8(b). The crack front is bounded by thin plane stress regions at the plate surfaces, which are no longer negligible in specimens of thicknesses in region B of Fig.2.4. The initial burst of crack growth is thus arrested by the high energy plane stress boundary regions. With the occurrence of pop-in, the maximum longitudinal stress is transferred from the central plane strain region onto the plane stress regions at the surfaces. Further crack extension takes place at
the central region, but at higher load because the plane stress regions require increasing load to be deformed\(^{(44)}\). Finally, a maximum load is reached, as shown in Fig.2.6(b), when the shear strength of the surface plane stress regions is surmounted by the increase in applied stress due to the decrease in specimen area. The entire fracture becomes unstable and failure occurs in a laminate fashion: with the central square fracture tunnels in while the boundary regions lag behind and fail by shear rupture on planes inclined at \(45^\circ\) to the loading direction forming the so-called shear lips\(^{(23,45)}\). At the thick end of region B, the fracture surface will therefore consist of a large proportion of square fracture, and the proportion of slant fracture decreases until it becomes insignificant (<15% total)\(^{(46)}\) in region C. On the other hand, the proportion of slant fracture increases as the plate thickness is decreased until \(B_0\) is reached, as shown in Fig.2.4, when the square fracture just disappears forming either double (or V-slat\(^{(45)}\)) or single shear fracture.

The thickness effect is strongly dependent on the size of the crack tip plastic zone relative to the thickness of the plate specimen. This aspect will be discussed later in section 3.7.

2.4 FRACTURE TOUGHNESS - THICKNESS CURVE

The dependency of \(K_c\) upon specimen thickness, \(B\), as shown in Fig.2.4, is redrawn in Fig.2.9 where \(B_{SL}\) is the thickness of the shear lips.

Many models have been proposed to establish quantitatively a basis for predicting this experimentally observed relationship between \(K_c\) and \(B\).
Essentially, from the principle of energy balance, the total critical fracture energy, $G_c$, is the sum of the fraction of energy dissipated in the shear lip formation and the fraction of energy dissipated in the square fracture. Mathematically, that is,

$$E_T = E_{SL} + E_{FF} \quad \text{(2.17)}$$

where $E_T$ is the total fracture energy per unit fracture area, and $E_{SL}$ and $E_{FF}$ are the per unit area fracture energy of the shear lip and flat fracture, or square fracture, respectively. It is noted that by definition, $E_T = G_c^{(44)}$.

In his model, Bluhm$^{(44)}$ made the assumptions that

a) the size of the shear lip, $B_{SL}$, beyond the critical plate thickness, $B_o$, is independent of the plate thickness,

b) flat fracture is a surface phenomenon while the formation of shear lip is a volume sensitive mechanism.

Assumption (a) implies that

$$B_{SL} = \frac{1}{2} B_o \quad \text{(2.18)}$$

From assumption (b), it can be deduced that, for unit length,
\[ E_{SL} = \begin{cases} \frac{1}{2} B, & B \leq B_o \\ \frac{1}{2} B_o^2 / B, & B > B_o \end{cases} \] .......................... 2.19

and

\[ E_{ff} = \left( 1 - \frac{B}{B_o} \right) \]

It follows that since \( G_c = E_T \)

\[ G_c = \begin{cases} \frac{1}{2} k_{SL} B_o \left( \frac{B}{B_o} \right), & \frac{B}{B_o} \leq 1 \\ \frac{1}{2} k_{SL} B_o + k_{ff} \left( 1 - \frac{B}{B_o} \right), & \frac{B}{B_o} > 1 \end{cases} \] .......................... 2.20

where \( k_{SL} \) and \( k_{ff} \) are assumed to be material constants which are evaluated experimentally.

For very thick specimens where \( G_c = G_{IC}, B_o / B \ll 1 \). Equation 2.20 gives

\[ G_{IC} = k_{ff} \] .......................... 2.21

The maximum \( G_c \) value occurs at \( B = B_o \) as shown in Fig.2.9. At this point, from equation 2.20, \( G_c \) is given by

\[ G_{c_{max}} = \frac{1}{2} k_{SL} B_o \text{ for } \frac{B_o}{B} = 1 \] .......................... 2.22

A very similar model to Bluhm's is postulated by Krafft, Sullivan and Boyle (47). Making the same assumptions as before, the model yields (48).

\[ G_c = \left( \frac{dW}{dA} \right) (1 - S) + \left( \frac{dW}{dV} \right) \frac{B_o^2}{2} \] .......................... 2.23

where \( (dW/dA) \) is the work done in producing unit area of square fracture, \( (dW/dV) \) is the plastic work density associated with the shear lip formation, and \( S \) is the fractional part of the fracture surface occupied by the shear lips. It is observed that equation 2.23 can essentially be resolved to

\[ G_c = S^2 G_{c_{max}} + (1 - S) G_{IC} \] .......................... 2.24
However, the analysis of Tetelman and McEvily\textsuperscript{(30)}, yields

\[ G_C = S G_C (45^\circ) + (1 - S) G_{IC} \] \hspace{1cm} 2.25

where \( G_C (45^\circ) \) is the work done in the 45\(^\circ\) slant fracture.

The 'necking-in' or depression associated with shear lip formation has been demonstrated by Hahn et al.\textsuperscript{(49,50)} to be related to the fracture energy of the shear lip. The relationships are, for per unit area, \( R_{SL} \), and per unit volume, \( \psi_{SL} \), shear lip fracture energies, respectively,

\[
\begin{aligned}
R_{SL} &= \overline{\sigma} A / B_{SL} \\
\psi_{SL} &= 2 \overline{\sigma} A / B_{SL} \lambda
\end{aligned}
\] \hspace{1cm} 2.26

where \( \overline{\sigma} \) is the average flow stress of the material, \( A \) and \( \lambda \) are the area and width respectively of the depression as shown in Fig.2.10.

Fig.2.10: Cross-sectional view of plate showing shear lip depression.

From energy balance, the total fracture energy \( R \) is therefore given by the expression\textsuperscript{(49)},

\[ R = 2 \left( B_{SL} / B \right) R_{SL} + \left( 1 - \frac{2B_{SL}}{B} \right) R_F \] \hspace{1cm} 2.27

where \( R_F \) is the fracture energy per unit area of the square fracture.

Substituting equation 2.26 into equation 2.27, \( R \) is reduced to

\[ R = \left( B_{SL} / B \right) \psi_{SL} + \left( 1 - \frac{2B_{SL}}{B} \right) R_F \] \hspace{1cm} 2.28
It can be noted that although different parameters are used in the analyses, equations 2.20 of Bluhm (44), 2.23 of Krafft et al. (47) and 2.28 of Hahn et al. (49) are analogous. The model of Tetelman and McEvily is, however, different in that equation 2.25 contains only an \( s \) term instead of an \( s^2 \) term (compare with equation 2.23 for example). This is critical because the agreement of these theoretical relations with experimental data appears to rest on the use of the square of the fraction of shear lips (48).

Other analyses of the relation between \( G_c \) and specimen thickness are given in References (51-53).

2.5 EFFECT OF TEST AND METALLURGICAL VARIABLES

2.5.1 Effect of Temperature and Strain Rate

The functional dependence of plane strain fracture toughness on stress and strain can be expressed by the equation (54)

\[
G_{IC}(T) = f\left[\frac{\sigma_{YS}(T)}{\varepsilon_f(T)}\right]
\]

where \( \varepsilon_f \) is the true strain at fracture and \( T \) is the temperature. Like other material properties, temperature has a great influence on the fracture toughness also. However, unlike the dependence of yield stress on temperature which is well understood, the temperature effect of \( K_{IC} \) of a high strength material has not heretofore been predictable.

Generally in steel, as temperature increases the yield strength level falls and the material exhibits more ductility in the fracture thus giving an increase in toughness (for example, see References (55-58)). This temperature effect on \( K_{IC} \) is greatest in relatively low strength structural steel. For very high-strength steels and non-ferrous alloys, the variation of toughness with temperature is small (59). This trend is analogous to the brittle-ductile transition obtained in a conventional Charpy impact test. Indeed, due to the difficulties in testing rather large specimens of low yield strength high toughness steel, attempts have been made to relate \( K_{IC} \) to Charpy impact energy (60,61) so that small test specimens can be employed to determine the \( K_{IC} \) value. Such empirical relationships have been found between plane strain \( K_{IC} \) and impact Charpy V-notch test results (60).

The transition temperature behaviour, which is observed to be unaffected by the \( K_{IC} \) to \( K_c \) stress-state transition, is thought to be
associated with the change in microscopic fracture mechanism.

Corten and Shoemaker\(^{(62)}\) postulated that \(K_{IC}\) is not only dependent on temperature, but also the strain rate \(\dot{\varepsilon}\) of the material through the rate parameter \(T \ln A/\dot{\varepsilon}\), where \(A\) is the frequency factor. For strain rate sensitive, relatively low strength structural steels, it is found that \(K_{IC}\) decreases with decreasing temperature and increasing strain rate. Corten and Shoemaker's rate parameter is however not applicable to high strength steels since these materials are fairly strain rate insensitive. Eftis and Kraft\(^{(63)}\) showed that the fracture toughness of mild steel varies with strain rate as shown in Fig.2.11.

At high strain rate (in a dynamic test, for example), the fracture toughness of a material using precracked, side-notched, impact Charpy specimen has often been found to be larger than the value obtained from static loading of the same type of specimen\(^{(64)}\). Hahn et al.\(^{(49)}\) demonstrated that the fracture toughness of high strength SAE 4340 steel at a crack speed of about 900 m/sec, is almost double that obtained in a static test.

2.5.2 Effect of Anisotropy

Deformation during processing produces anisotropy which affects the properties of the material. Fracture toughness tests show that \(K_{IC}\) values are sensitive to the testpiece orientation in the parent material\(^{(65-68)}\). This effect is partly due to fibering generally found in rolled products and partly to chemical segregation banding in the material\(^{(69)}\). As a result, it is easier for cracks to propagate in
20. a direction parallel to the fibre direction. Fig.2.12 shows the designation of the different fracture toughness specimen orientations suggested by Goode (70).

![Diagram showing different orientations](image)

**Fig.2.12**: Coöde system for crack orientation in plate

Generally, it is found that L-T orientation gives the highest $K_{IC}$ value while the S-L orientation gives the lowest. The difference between values depends on the thickness of the material (65,66). Thicker material shows a larger difference in $K_{IC}$ values. Zinkham (65) explains that the greater degree of anisotropy in thicker material is due to the better alignment of the inclusion stringers and the more elongated grain structure.

### 2.5.3 Effect of Impurities

Fracture toughness parameters are very sensitive to the microstructures of the materials. The volume fraction, distribution, hardness and morphology of inclusions in a material will affect its value of $K_{IC}$ (71). Evans et al. (72) showed that the $K_{IC}$ of high-purity steel of En 24 is about double that for the same but commercially produced material. This improvement in toughness may be attributed to the reduction in inclusion content in the purer steel.

Sulphur and phosphorus contents in high levels are proven to have deleterious effects on the fracture toughness of high strength steels (73,74). Fig.2.13(a) illustrates the effect of sulphur on the fracture toughness of a medium carbon, nickel-chrome molybdenum steel heat treated to a yield strength of about 1460 MPa (212 Ksi). In steels with pearlitic micro-
structures (carbon and silicon contents fall in the range from 0.10 to 0.30% and 0.05 to 0.25% respectively), decrease in carbon or increase in silicon level was found to improve the fracture toughness (75).

Fig. 2.13: Effect of (a) sulphur, (b) silicon on fracture toughness

Using AISI 4340 type ultrahigh strength steel, Wei (68) obtained the effect of silicon level on fracture toughness to be that shown in Fig. 2.13(b). In the latter material, it is believed that the mechanisms to improve toughness are inhibited (68). Cottrell (74) has demonstrated that the addition of sulphur, phosphorus, arsenic and tin to a pure 2% nickel-chromium-molybdenum steel reduces the value of its fracture toughness. Beside the above alloy impurities, interstitial alloy elements such as carbon, nitrogen and oxygen can produce a deleterious effect on $K_{IC}$ value. While chromium, manganese and vanadium improve the $K_{IC}$ value, nickel and molybdenum appear to reduce the toughness after a certain tempering temperature (77).

In the case of aluminium alloys, it is found that the Aluminium Association 7000 series is superior in toughness than the 2000 series due to the lower contents of insoluble intermetallic phases (78). Carmen et al. (79) working with 7075-T6 alloy showed that reducing the iron and silicon contents produces remarkable improvement in $K_{IC}$ value. For the 7000 series high strength heat treatable alloys, it is found that both zinc and magnesium additions will reduce the fracture toughness value (80). Addition of copper in the yield strength range of 400 to 500 MPa is believed to improve toughness. However, silicon addition is undesirable although it increases the fracture toughness of the 7000 series aluminium alloy. This is because of the formation of $\text{Mg}_2\text{Si}$ which reduces the yield strength of the material (81).
2.6 **SHEAR LIPS**

The development of $45^\circ$ slant fracture or shear lips at the specimen surfaces in cracking plates of high strength alloys is a familiar fractographic phenomenon. It is suggested (for example, Reference 82) that shear lips only occur in materials that can undergo plane stress yielding and in specimens where constraints are such that plane strain yielding can take place at most in the specimen interior.

Shear lips are observed to occur either on parallel or perpendicular planes (see Fig.2.14) in thick specimens. In thin specimens, however, they generally appear on parallel planes. This preference may be due to the asymmetry in the crack tip stress field introduced as a result of the appearance of a small shear lip on one side of the specimen (83). The size of the shear lips, $B_{SL}$, once fully developed, is constant along the length of the fracture path.

![Shear lips on parallel planes.](image1.png)

![Shear lips on perpendicular planes.](image2.png)

Fig.2.14: Shear lips on free surfaces of plate

In addition, it is usually observed although exceptions do occur (86-88), that $B_{SL}$ of a fully grown shear lip (beyond $B_0$ of Fig.2.4) is independent of the specimen thickness (24, 47, 57, 89, 90). At a constant thickness, however, shear lip size increases with the increase in temperature (3) and crack speed (49). The temperature dependence of the amount (56) of shear lip or the absolute shear lip size (57) is approximately coincident with the temperature dependence of the material fracture toughness.

The profile of the slant fracture or shear lip is conventionally assumed to incline at an angle of $45^\circ$ to the plane of crack propagation. It appears that this general shape does not change with the testing condition (91). This is plausible since the plastic zone size for a slant fracture has a maximum at $45^\circ$ (92, 93). Trudeau (94, 95), considering that
Shear lips are actually zero-isoclinic surfaces characterized by purely normal displacements and zero shear, postulated that the shear lip profile starts at 60° to the plane of the square fracture and gradually decreases to an angle of 45°. Fracture contours from a number of crack-notch toughness specimens were found to follow closely an elastic zero isoclinic even when the fracture was preceded by a few percent plastic flow. On the other hand, the work of Yusuf showed that an angle in the range of 50-59° is found for DTD.687 and 7075-T6 aluminium alloys and 35° for the 6% Al, 4% V titanium alloy. The well defined profile of the shear lip is thought to have a well defined mechanical cause, and its simultaneous appearance on both sides of a plate must be due to the satisfaction of some requirements, for example, the attainment of a critical K_c value and the existence of a plane stress state (although plane stress does not necessarily lead to shear lip formation, as in beryllium and some titanium alloys). Indeed, Krafft et al. (47,97) suggested that shear lip is the elastic plastic boundary of the plastic zone at the free surface and hence a measure of the size of this plastic zone. Mathematically, therefore, \[ \text{shear lip size} = \frac{2\pi}{\sin^{-1}(\frac{K_c}{\sigma_y})} \]

The relationship shows that the fracture surface shear lip is a measure of the first trace of plastic shear reaching the free surface and in turn the plane stress plastic zone radius. Equation 2.30 has also been suggested by Gran et al. (98). The validity of the above plastic zone/shear lip size relationship has been demonstrated by Mowbray et al. (57) using Ni-Mo-V steels, Knott (36) using 7075-T6 aluminium alloy data of Krafft et al. (47) and Holt, Khor and Lai (99) using 3/8" thick single-edge-cracked tension specimens of Comsteel 3140. In a failure analysis study of a 17.8mm thick plate of D6AC steel, Hertzberg (100) showed that equation 2.30 can be used to estimate the fracture toughness value of a material. Carr et al. (101,102), have suggested that the shear lip is formed inside the plastic zone and consequently, the plastic zone size can be larger but not smaller than the size of the shear lip (47). This is schematically illustrated in Fig.2.15.

Shear lips are not only associated with metals, they are also observed in the fracturing of many organic thermoplastics like polyurethane polymers (103), polycarbonates (82), polystyrene, vinyl chloride-vinyl acetate copolymers (104) and Ni-Fe base metallic glasses (105,106).
Fig. 2.15: Relationship between plastic zone and shear lip in plate specimen.

Even in these materials, shear lips are found to incline at an angle of 45° to the tensile axis or the plane of fracture propagation (105).

In fatigue cracking of plate specimens, it is observed that shear lips are always developed at the specimen surfaces after the crack has propagated a certain distance in a square or 90 degree fracture mode (17,40,107-111) as shown in Fig.2.16.

Fig.2.16: The transition of a fatigue crack in plate

From considerations of stress distribution, Liu (112) suggested that the transition from one mode of fracture to another is dependent on the thickness of the specimen tested. Wilhem (113), however, experimentally demonstrated that the transition in 2024-T3 aluminium is independent of the specimen thickness but dependent only on the stress intensity factor at transition.

The formation of shear lips has been postulated to occur by shear rupture on planes that are inclined at 45° to the plane of the notch (30). Elongated shear dimples are generally observed on the shear planes (114,115). Open elongated dimples (dimples with the open end of the parabolas toward the fracture edge) are found exclusively on the
acute angle shear lip while closed elongated dimples (dimples with the closed end of the parabolas toward the fracture edge) along the obtuse angle shear lip \(^{(116)}\).

2.7 STANDARD PLANE STRAIN FRACTURE TOUGHNESS TESTING

The procedure for standard plane strain fracture toughness testing is fully discussed in Ref.\(^{(117)}\) and documented in the ASTM\(^{(118)}\) and British\(^{(119)}\) Standards. The main aim is to incorporate the requirements to be fulfilled to produce plane strain conditions at the crack tip so that a reproducible value for the lower limiting critical fracture toughness of a metallic material can be obtained. Basically, the procedure involves testing a notched, fatigue precracked specimen of specified thickness and geometry under tensile loading or three point bending. The load corresponds to a 2\(^{\circ}\) increment of crack is established from the load/displacement curve that is recorded autographically during the test. The critical fracture toughness of the material is then calculated from this load using theoretical expression analysed for the particular specimen geometry. If the requirements with respect to specimen thickness and crack length are satisfied, the test is valid and the critical fracture toughness is taken as the plane strain fracture toughness \(K_{IC}\) value.

Many types of specimens described by Srawley and Brown\(^{(20,45)}\), are suitable for fracture toughness testing but only two are recommended in the standards\(^{(118,119)}\). These are the three-point bend specimen and the compact tension specimen as shown in Fig.2.17(a) and (b) respectively. In designing the specimen dimensions, it is necessary that a state of plane strain is established at the crack tip. To ensure this, the specimen must be thick enough so that the shear lip contribution to the toughness can be approximately omitted. Secondly, the plastic zone size at the crack tip must be small compared to the length of the crack, \(a\), so that a single-valued \(K\) parameter can be used to characterise the fracture at critical condition. Finally, the width of the specimen ligament, \(W-a\), must be large compared to the plastic zone size so that the crack tip stress field is not influenced by the free boundary of the specimen. It is obvious that these requirements should be a multiple of the plastic zone size which is expressed in terms of the characteristic dimension \((K_{IC}/\sigma_{YS})^2\). The recommendation is that\(^{(20,31)}\)

\[
a, W-a, B > 2.5 \times \left(\frac{K_{IC}}{\sigma_{YS}}\right)^2 \quad \ldots \quad 2.31
\]
which is sixteen times the plane stress plastic zone size of $r_y = \frac{1}{2\pi} \left( \frac{K_{IC}}{\sigma_{YS}} \right)^2$.

![Diagram](image)

**Fig.2.17:** ASTM standard specimens

Under these restrictions, $K_{IC}$ values are reproducible to about $\pm 10\%$ (46).

It is necessary to provide a sharp crack at the tip of the stress concentrator in all specimens. As shown in Fig.2.18 since the fracture toughness decreases with the notch root radii until a limiting radius is reached (120), fatigue cracking is generally used to sharpen the starter notch. To ensure that a sharp fatigue crack is achieved, the maximum stress intensity during the final stage of fatigue crack extension must not exceed 60% of the $K_{IC}$ value subsequently determined.

![Graph](image)

**Fig.2.18:** Effect of root radius on fracture toughness
This requirement may be appreciated from the results of May (121) or Brown and Srawley (122) as shown in Fig. 2.19.

![Diagram](image)

*Fig. 2.19: Effect of fatigue cracking stress intensity on fracture toughness (122).*

The experimental procedure is to record autographically the load on the test specimen and the displacement across the opposite faces of the stress concentrator. The latter is monitored with a double-cantilever beam clip gauge to be described in section 5.5.1. The fracture toughness value is evaluated at a load, $P_{Q}$, corresponding to a 2% increment in crack extension. This is graphically determined by a specified deviation from linearity on the load versus displacement record (118, 123). The calculated fracture toughness value is considered as valid $K_{IC}$ if the ratio $P_{max}/P_{Q}$ where $P_{max}$ is the maximum load sustained by the specimen, is less than 1.10, and if the specimen dimension requirements described above are satisfied.

Although the ASTM standard requirements for plane strain fracture toughness testing are generally applicable to most materials, there are cases in which they are believed to be inadequate. In a recent study of the effect of specimen size on plane strain fracture toughness of 2219-T851 aluminium alloy, Kaufman and Nelson (124) found that in order to obtain size independent
K, results, it is necessary to increase the crack length limit to $a > 5(K_{IC}/\sigma_V)^2$ even within the general plane strain region. This has the effect of keeping the net-section stress below two-thirds of the yield stress. On the effect of specimen size using Ti-6Al-4V, Munz et al. (125) concluded that the ASTM recommendation for specimen thickness cannot avoid thickness dependent $K_Q$ values and that $P_{max}/P_Q < 1.0$ requirement for valid test is unable to exclude the effect of thickness. The inclusion of the $P_{max}/P_Q$ ratio to restrict the contribution of crack tip plasticity to the displacement at the measurement point $P_Q$ is therefore questionable (126). Specimen size dependent $K_Q$ values, even when ASTM size requirement was satisfied, were also obtained by other investigators (90, 127, 128).
CHAPTER 3

CRACK TIP PLASTIC ZONE
CHAPTER 3

CRACK TIP PLASTIC ZONE

3.1 CRACK TIP STRESS FIELD

In fracture mechanics, the interest lies in the elastic state at the vicinity of the crack tip. The fundamental principle is that the stress field ahead of a sharp crack can be characterised by three quantities, \( K_I, K_{II} \) and \( K_{III} \) each associated with a local mode of deformation as shown in Fig.2.5. The quantities are known as the stress intensity factors which are a measure of the stress singularity at the crack tip. The present discussion deals only with the opening mode, Mode I which is associated with local displacement in which the crack surfaces move directly apart.

The stress fields in the vicinity of the crack tip subjected to Mode I deformation (Refer to Appendix A for derivation) are given by (7,12,13,18),

\[
\sigma_x = \frac{K_1}{\sqrt{2\pi r}} \cos \frac{\theta}{2} \left[ 1 - \sin \frac{\theta}{2} \sin \frac{3\theta}{2} \right] \\
\sigma_y = \frac{K_1}{\sqrt{2\pi r}} \cos \frac{\theta}{2} \left[ 1 + \sin \frac{\theta}{2} \sin \frac{3\theta}{2} \right] \quad \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots 3.1 \\
\tau_{xy} = \frac{K_1}{\sqrt{2\pi r}} \sin \frac{\theta}{2} \cos \frac{\theta}{2} \cos \frac{3\theta}{2} \\
\sigma_z = \nu(\sigma_x + \sigma_y) \\
\tau_{xz} = \tau_{yz} = 0
\]

where the coordinates \( r \) and \( \theta \) and the stress components are shown in Fig.3.1, and \( \nu \) is the Poisson's ratio of the material. Equations 3.1 are written for the case of plane strain conditions, that is, \( \sigma_z = \nu(\sigma_x + \sigma_y) \), but can be applied to the plane stress situation by taking \( \sigma_z = 0 \). It is noted that the equations have been derived by omitting the higher-order terms in \( r \).
As such, they are exact in the limit as \( r \) approaches zero but are only a good approximation in region where \( r \) is small compared to the \( x - y \) planar dimensions of the specimen such as crack length. The equations also show that the magnitude of the elastic stress field can be described by the stress intensity factor, \( K_I \), which is independent of the coordinates \( r \) and \( \theta \). From dimensional analysis, \( K_I \) must be linearly related to the magnitude of the applied stress and the square root of a characteristic length, the crack size. Accordingly, \( K_I \) reflects the redistribution of the stress in a body as a result of the introduction of a crack.

3.2 CRACK TIP PLASTIC ZONE

Equations 3.1 indicate that the elastic stress in the vicinity of a crack where a stress singularity exists, can be very large at \( r \ll a \). The stress at the crack tip is higher than the overall applied stress and at a sharp crack, localised plastic deformation occurs when the appropriate yield criterion is satisfied even at very small applied stress levels. Hence a stress singularity cannot exist and a plastically deformed region is created in front of the crack tip. This plastic region is generally known as the crack tip plastic zone. If the plastic zone is small enough relative to the elastically stressed region around it, the elastic stress singularity described by equations 3.1 will still be a good approximation to the stress field. This condition is readily satisfied if the plastic zone size is small compared to the length of the crack and to the distance to the next nearest boundary of the specimen. The size and shape of the plastic zone depend on the mode of deformation that acts on the crack and the yield criterion (Tresca or Von Mises) that is applicable. A rough estimate of the plastic zone size is simply to treat the problem as one of
plane stress and to assume that the plastic zone would cover the area where the elastic solution gives stresses greater than the tensile yield stress, $\sigma_{ys}$, of the material.

The stress $\sigma_y$ distribution in the vicinity of a crack tip is shown in Fig.3.2. Until a distance $r_y$, the stress is greater than the yield stress $\sigma_{ys}$ of the material. The size of the plastic zone is thus, to a first approximation, that of distance $r_y$, which can be determined by substituting $\sigma_{ys}$ into equations 3.1 for $\sigma_y$ and $\theta = 0$.

![Stress distribution at crack tip and first estimation of plastic zone size.](image)

Therefore,

$$\sigma_{ys} = \frac{K}{\sqrt{2\pi r_y}}$$

or

$$r_y = \frac{1}{2\pi} \left( \frac{K}{\sigma_{ys}} \right)^2$$

3.2

Within the plastic zone, the maximum tensile stress is reduced from the elastic value shown by the curve $\sigma_y = K/\sqrt{2\pi r}$, to the tensile yield stress $\sigma_{ys}$ of the material. Accordingly, in order to take up the load that is carried by the elastic material in the region $0<r<r_y$ (the shaded area), the plastic zone must extend to a distance greater than $r_y$.

A second approximation of the plastic zone size can be obtained by evaluating the total area under the curve $\sigma_y = K/\sqrt{2\pi r}$ between $r = 0$
and \( r = r_y \). It can be shown that this is equal to \( 2\sigma_{ys} r_y \); consequently, the shaded area must be \( \sigma_{ys} r_y \). Hence the plastic zone effectively extends in front of the crack tip to a distance approximately \( 2r_y \). As shown in Fig.3.3,

\[
x_p = 2r_y = \frac{1}{\pi} \left( \frac{K}{\sigma_{ys}} \right)^2
\]

**Fig.3.3:** Second estimation of plastic zone size.

It was postulated by Irwin\(^{(2)}\) that the occurrence of the plastic zone causes the plate to behave as if it contained a crack of slightly larger in size. The effective half crack size is then equal to \( a + r_y \), the physical crack size plus a correction term, \( r_y \). This supplemental term is known as Irwin's plastic zone correction term for matching an equivalent elastic crack to an elastic-plastic crack\(^{(129)}\).

If the plastic zone correction is applied to the stress intensity factor, \( K \) would be given by

\[
K = \sigma_y \sqrt{2\pi (a + r_y)}
\]

**Fig.3.3:** Second estimation of plastic zone size.

The use of equation 3.4 involves an iteration procedure. But this can be avoided by taking \( K = \sigma\sqrt{2\pi a} \) to estimate \( r_y \) and then evaluating the corrected value of \( K \) from equation 3.4. Plastic zone correction is not necessary in situations where linear elastic fracture mechanics is applicable since the plastic zone size is negligible compared to the crack size. However, when plastic zone size is large which happens when the net section stress exceeds about \( 0.8\sigma_{ys} \)\(^{(15)}\), the application of a plastic zone correction becomes somewhat redundant due to the limited accuracy of the expression for the stress intensity factor. In practice, therefore, the
33. Plastic zone correction term is not often used.

3.3 PLASTIC ZONE SIZE PREDICTION

3.3.1 Dugdale Model

A more precise approach for determining the size of the plastic zone for a sharp tensile crack under plane stress condition in a non-strain-hardening material is given by Dugdale (130). Like Irwin (129), Dugdale considered an effective crack of length longer than the actual physical crack as shown in Fig.3.4.

![Fig.3.4: Spread of plasticity from a crack: Dugdale's model.](image)

The plasticity spreads from the tip of the crack to a further distance ρ. The region ρ is however not a crack because it has been closed up by an internal tensile stress. Since internally applied forces are in static equilibrium, this internal tensile stress must be equal to σ_{ys} because the tensile stress existing in the plastic zone is σ_{ys}. The size of the region ρ is so chosen that the stress singularity disappears, that is, K = 0. To acquire this, the stress intensity K₀ due to the uniform applied stress σ must compensate the stress intensity Kₚ due to the wedge forces σ_{ys}.

Mathematically, then

\[ K₀ = Kₚ \]
The stress intensity due to the wedge forces \( K_p \) can be computed using Green's function (13), and the result is given as,

\[
K_p = 2\sigma \frac{\sqrt{a + \rho}}{\pi} \cos^{-1} \left( \frac{a}{a + \rho} \right) \quad \text{(3.6)}
\]

From equation 3.5, putting \( K_\rho = \sigma\sqrt{a + \rho} \), \( \rho \) can be evaluated.

\[
\frac{a}{a + \rho} = \cos \left( \frac{\pi}{2} \frac{\sigma}{\sigma_{ys}} \right) \quad \text{(3.7)}
\]

With Westergaard stress functions, Burdekin and Stone(131) showed that the displacement at the crack tip in the \( y \)-direction, denoted by \( \delta \) is given by

\[
\delta = \frac{8}{\pi} \frac{\sigma_{ys}}{E} a \ln \left[ \sec \left( \frac{\pi}{2} \frac{\sigma}{\sigma_{ys}} \right) \right] \quad \text{(3.8)}
\]

where \( E \) is the Young's Modulus. \( \delta \) is generally known as the Crack Opening Displacement which will be discussed in Chapter 4.

Equation 3.7 can be simplified by neglecting the higher order terms in the expansion of the right-hand-side to give

\[
\rho = \frac{\pi}{8} \frac{(K_p)}{\sigma_{ys}} \quad \text{(3.9)}
\]

Equation 3.9 compares well with equation 3.3 in which

\[
r_p = \frac{1}{\pi} \frac{(K_p)}{\sigma_{ys}} \quad \text{(3.9)}
\]

It should be noted that at high \( \sigma/\sigma_{ys} \), equation 3.6 gives a better estimation of the plastic zone size than equation 3.9.

A very similar analysis to Dugdale's is also given by Barenblatt (132). An alternative method by Bilby, Cottrell and Swinden which yields similar results is described in the following section.
3.3.2 BILBY - COTTRELL - SWINDEN MODEL

Fig. 3.5 shows an infinite isotropic elastic body of shear modulus \( \mu \), subject to a uniform applied shear stress \( \sigma_{yz} = q \) at infinity.

![Diagram](image)

Mode III deformation

**Fig.3.5 : Representation of plastic zone by arrays of dislocations**

The plastic zone, \( \rho \), and the crack of length, \( 2a \), are represented as arrays of long straight screw dislocations lying parallel to the \( z \) axis in the \( xz \) plane. The dislocations are spaced closely at the tip of the crack but widely as the plastic-elastic interface is reached so as to accommodate the displacement gradient. The resistance to motion in the yield region \( \rho \) is taken as \( \tau_y \).

From the theory of distributions of dislocations, it is possible to formulate a singular integral equation which expresses that when the system is in equilibrium, the resultant shear stress on any dislocation in the distribution is zero. The size of region \( \rho \) evaluated from such an equation can be shown (133) to be

\[
\frac{a}{a + \rho} = \cos \left( \frac{\pi}{2} \frac{q}{\tau_y} \right) \quad \ldots \ldots \ldots \ldots \ldots \quad 3.10
\]

and the relative displacement at the crack tip in the \( y \) direction is

\[
\phi = \frac{4\tau_y}{\pi\mu} a \ln \left[ \sec \left( \frac{\pi}{2} \frac{q}{\tau_y} \right) \right] \quad \ldots \ldots \ldots \ldots \ldots \quad 3.11
\]
Equations 3.10 and 3.11 are identical to equations 3.7 and 3.8 respectively, and in fact, if plane stress Mode I tension case is considered, equation 3.11 can be rewritten as

\[ \delta = \frac{8}{\pi} \frac{\sigma_{ys}}{E} a \ln \left[ \sec \left( \frac{\pi}{2} \frac{\sigma}{\sigma_{ys}} \right) \right] \]

which is the same as equation 3.8.

3.4 PLASTIC ZONE SIZE UNDER PLANE STRAIN CONDITION

Owing to the fact that the effective yield stress of a non-strain hardening material in plane strain is larger than that in plane stress, the plastic zone size in plane stress is rather different from that in plane strain. The maximum effective stress in the plane strain plastic zone can be as much as three times the nominal yield stress (15), whereas at the tip of the crack where a state of plane stress exists, the stress is limited to the yield stress \( \sigma_{ys} \). Fig.3.6 shows the stress distribution under plane stress and plane strain conditions.

Fig.3.6: Stress distribution in (a) plane stress, (b) plane strain conditions

The stress increases rapidly with increasing distance from \( \sigma_{ys} \) at the crack tip to a maximum value of \( 3\sigma_{ys} \) (15) in the plane strain condition. From equation 3.2, the plastic zone size under plane strain state can thus be computed as

\[ r_{y} = \frac{1}{2\pi} \left( \frac{K}{3\sigma_{ys}} \right)^2 = \frac{1}{18\pi} \left( \frac{K}{\sigma_{ys}} \right)^2 \quad \ldots \quad 3.12 \]
Actually, since a stress perpendicular to a free surface is not possible, plane strain condition does not exist at the specimen surfaces. The plastic constraint factor which is the ratio of the maximum stress to the yield stress, or mathematically,

\[ \text{plastic constraint factor} = \frac{\sigma_{\text{max}}}{\sigma_y}, \quad \ldots \ldots \ldots \ldots \quad (3.13) \]

will have an average value of less than 3. Plastic constraint factors are experimentally found to be between 1.5 to 2 \( (134) \). McClintock and Irwin \( (1) \) from empirical considerations suggested that the plane strain plastic zone size be smaller than the plane stress case given in equation 3.2 by about one third. For numerical simplicity, Irwin \( (1) \) used a plastic constraint factor of \( (2)^{3/2} \) which reduces equation 3.2 to

\[ r_y = \frac{1}{4\pi r^2} \left( \frac{K}{\sigma_y} \right)^2 = \frac{1}{6\pi} \left( \frac{K}{\sigma_y} \right)^2 \quad \ldots \ldots \ldots \ldots \quad (3.14) \]

Equation 3.14 is generally taken as the plane strain plastic zone correction factor.

3.5 SHAPE OF THE PLASTIC ZONE

The shape and size of the crack tip plastic zone depend on the plastic flow properties of the material \( (20) \), the specimen size, boundary conditions and the state of stress whether plane stress or plane strain. In the previous sections, only the extent of the plastic zone along the \( x \) axis was discussed. The shape of the plastic zone is complex and difficult to determine \( (135,136) \) although a first impression of its shape can be gotten if the yield condition for all values of \( \theta \) in equations 3.1 is considered. This involves the choice of an appropriate yield criterion which generally is either the Tresca yield criterion or the Von Mises yield criterion. The two criteria yield different shape and size of plastic zone.

The extent of the plastic zone as a function of \( \theta \) using the Tresca yield criterion can be shown as:

For plane stress: \[ r_p (\theta) = \frac{1}{2\pi} \left( \frac{K}{\sigma_y} \right)^2 \left[ \cos \frac{\theta}{2} \left( 1 + \sin \frac{\theta}{2} \right) \right]^2 \quad \ldots \ldots \ldots \ldots (3.15(a)) \]
For plane strain: the larger of

$$r_p(\theta) = \frac{1}{2\pi} \left( \frac{K}{\sigma_y} \right)^2 \cos \frac{\theta}{2} \left[ 1 - 2\nu + \sin \frac{\theta}{2} \right]^2$$

and

$$r_p(\theta) = \frac{1}{2\pi} \left( \frac{K}{\sigma_y} \right)^2 \cos \frac{\theta}{2}$$

3.15(b)

On the other hand, with Von Mises yield criterion,

For plane stress: $$r_p(\theta) = \frac{1}{4\pi} \left( \frac{K}{\sigma_y} \right)^2 \left[ 1 + \frac{3}{2} \sin^2 \theta + \cos \theta \right]$$

3.16(a)

For plane strain: $$r_p(\theta) = \frac{1}{4\pi} \left( \frac{K}{\sigma_y} \right)^2 \left[ \frac{3}{2} \sin^2 \theta + (1 - 2\nu)^2 (1 + \cos \theta) \right]$$

3.16(b)

The boundary of the plastic zone predicted in equations 3.15 is depicted in Fig.3.7 where the axes are expressed in non-dimensional parameter $r_p/\left[ \frac{1}{\pi} \left( \frac{K}{\sigma_y} \right) \right]$.

![Diagram](image)

Fig.3.7: Plastic zone shape according to (a) Von Mises, (b) Tresca yield criteria.
It is observed that the Tresca yield zones are larger than the Von Mises yield zones. The difference between the plane strain and plane stress loci of the plastic zone is not due to the stress components in the plane, but to the existence or non-existence of the transverse constraint.

In the case of a thick plate specimen, since stresses perpendicular to the outer surfaces are not possible, plane stress stress state always exists at the surface while plane strain stress state operates in the interior. Therefore, at the surface

\[ \sigma_z = \sigma_3 = 0 \]  \hspace{1cm} \text{(3.17)}

where \( \sigma_3 \) is the third principal stress always acting perpendicular to the plate: \( \sigma_z \equiv \sigma_3 \). The stress \( \sigma_3 \) increases gradually from zero at the surface to the plane strain value at the interior. Accordingly, the plastic zone size decreases from its plane stress value at the surface to a smaller plane strain value at the plate interior. A three-dimensional impression of the plastic zone at the tip of a through-the-thickness Mode I crack based on Von Mises yield criterion is shown in Fig.3.8.\(^{(20)}\)

Fig.3.8: Three-dimensional plastic zone according to Von Mises yield criterion.
The plastic zone at the specimen surfaces extends further than it does at the interior where the surface influence is small. The surface influence extends into the thickness of the specimen for a distance that is proportional to the characteristic dimension \( (\sigma_y/\sigma) \). The size and shape of the crack tip plastic zone illustrated in Figs. 3.7 and 3.8 are, however, only approximate since the same error was made, as in deriving equation 3.2: by limiting the stress to the yield stress of the material and allotting the extra load to the elastic material outside the supposed boundary of the plastic zone. Correction to such an error involves great complexity. The problem has attracted large number of experimentalists and theoreticians like Hahn and Rosenfeld, Underwood and Kendall, Stimpson and Eaton, Liu, Jacobs, Prandtl, Hult and McClintok, McClintok, Allen and Southwell, Hill, Green, Bilby and Swinden, Rice and Rosengren, and Tuba. In general, the crack tip plastic zone size can be expressed in the form

\[
\begin{align*}
    r_y (\pi/2) &= B (\sigma_y/\sigma)^2 \\
    r_y (\pi) &= A (\sigma_y/\sigma)^2
\end{align*}
\]

along the x and y direction respectively. The values of the coefficients A and B under plane strain stress state conditions, for small scale yielding are briefly summarized in Table 3.1 after Reference (152). Under plane stress conditions, the maximum plastic zone size, \( r_{y_{\text{max}}} \) and that along the x direction, \( r_y (\pi) \) are given in Table 3.2 after Ref (155).

In Table 3.2, all the yield criteria suggest that the maximum extent of the plastic zone always inclined at an angle \( \phi \) of about 60° to the crack plane as shown in Fig. 3.9. Tuba, Rice et.al, and Levy et.al all obtained an angle of about 70° in their analyses. However, from stress analysis, since maximum shear stress occurs at \( \theta = \pm 90° \) and the maximum octahedral stress at \( \pm 70° \), the plastic zone is expected to incline at an angle of \( \pm 70 \) to 90° to the crack plane. Using etching technique, Hahn and Rosenfield demonstrated that the maximum extent of the plastic
<table>
<thead>
<tr>
<th>Calculation</th>
<th>A</th>
<th>B</th>
</tr>
</thead>
<tbody>
<tr>
<td>Elastic stress field: Combined with Von Mises or</td>
<td>0.025</td>
<td>0.21</td>
</tr>
<tr>
<td>Tresca yield criterion and plane strain (153)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Strip yield models: Dugdale (130)</td>
<td>0.40</td>
<td>0.14</td>
</tr>
<tr>
<td>Bilby, Cottrell and Swinden (133)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Bilby and Swinden (149)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Continuum Plasticity: Rice and Rosengren (150)</td>
<td>0.007</td>
<td>0.14</td>
</tr>
<tr>
<td>Levy, Marcal, Ostergren and Rice (154)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Etching experiments: Hahn and Rosenfield (138)</td>
<td>0.03</td>
<td>0.20</td>
</tr>
</tbody>
</table>

**Table 3.1: Summary of Plane Strain Plastic Zone Size (152)**
<table>
<thead>
<tr>
<th>Failure Criterion</th>
<th>Proposer</th>
<th>Plastic zone size, $r_y$</th>
<th>$r_y (0)$</th>
<th>$r_{ymax}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Von Mises</td>
<td>Rice (156-158)</td>
<td>$0.239 \left( \frac{K}{\sigma_{YS}} \right)^2 \cos^2 \frac{\Theta}{2} \left[ 1 + \sin \frac{\Theta}{2} \cdot \sin \frac{3\Theta}{2} \right]^2$</td>
<td>$0.239 \left( \frac{K}{\sigma_{YS}} \right)^2$</td>
<td>$0.40 \left( \frac{K}{\sigma_{YS}} \right)^2$</td>
</tr>
<tr>
<td></td>
<td>Hutchinson (159,160)</td>
<td>$0.239 \left( \frac{K}{\sigma_{YS}} \right)^2 \left[ \cos^2 \frac{\Theta}{2} + \frac{3}{4} \sin^2 \Theta \right]$</td>
<td>$0.239 \left( \frac{K}{\sigma_{YS}} \right)^2$</td>
<td>$0.32 \left( \frac{K}{\sigma_{YS}} \right)^2$</td>
</tr>
<tr>
<td>Octahedral shear stress</td>
<td>Liu (112)</td>
<td>$0.358 \left( \frac{K}{\sigma_{YS}} \right)^2 \cos^2 \frac{\Theta}{2} \left[ 1 + 3 \sin^2 \frac{\Theta}{2} \right]$</td>
<td>$0.358 \left( \frac{K}{\sigma_{YS}} \right)^2$</td>
<td>$0.47 \left( \frac{K}{\sigma_{YS}} \right)^2$</td>
</tr>
<tr>
<td>Tresca</td>
<td>Paris (161)</td>
<td>$0.318 \left( \frac{K}{\sigma_{YS}} \right)^2 \cos^2 \frac{\Theta}{2} \left[ 1 + \sin \frac{\Theta}{2} \cdot \sin^3 \frac{\Theta}{2} \right]^2$</td>
<td>$0.318 \left( \frac{K}{\sigma_{YS}} \right)^2$</td>
<td>$0.54 \left( \frac{K}{\sigma_{YS}} \right)^2$</td>
</tr>
<tr>
<td>Plastic Cohesive Force</td>
<td>Tomkin (162)</td>
<td>$0.392 \left( \frac{K}{\sigma_{YS}} \right)^2$ at $\pm 45^\circ$</td>
<td>-</td>
<td>$0.392 \left( \frac{K}{\sigma_{YS}} \right)^2$</td>
</tr>
</tbody>
</table>

**TABLE 3.2: SUMMARY OF PLANE STRESS PLASTIC ZONE SIZE (155)**
zone occurs at $\phi = 65 \pm 8^\circ$ which agrees well with the results of Tuba and Levy et al. but differs considerably from Rice and Rosengren's model (150) as shown in Fig. 3.10 where $\phi = 100^\circ$.

![Diagram showing various zones and stress shapes]

Fig. 3.9: More accurate plane stress plastic zone shape

![Diagram showing plastic zone shapes with equations]

Fig. 3.10: Plastic zone shape according to Rice and Rosengren (151).
Further, Hahn and Rosenfield (138) showed that in comparison with etching data, the furthest extent of the plastic zone was quite well estimated by Bilby and Swinden (149), Rice and Rosengren (150) and Tuba (151). However, current theoretical analyses seem unable to obtain a satisfactory value for \( r_y^{\infty} \).

Recent analyses (164) have shown that the non-singular stress term in the stress field at the crack tip acting parallel to the crack plane affects the size of plane strain plastic zone. Taking into account the effect of this T-stress term, \( r_{y\max}^{\infty} \) in plane strain conditions for compact tension specimens has been found (165,166) to be \( 0.15(\sigma_{ys}/K)^2 \) which compares favourably with the results of other investigators as shown in Table 3.1. The angle \( \phi \) of 81° is however slightly different.

The size and shape of the plastic zone are affected by the stress state. On the other hand, the plastic zone influences the state of stress. When the plastic zone is large in relation to the thickness of a plate specimen, yielding can freely take place in the through-thickness direction and thickness reduction or necking occurs during fracture. When the plastic zone size is small, because of the constraint of the surrounding elastic material, yielding in the through-thickness direction cannot take place. The result is that the plate may crack, in the former case with plane stress or slant-type fracture while in the latter, with plane strain or square-type fracture. A convenient method to measure the tendency for plane stress or plane strain fracture is by comparing the size of the plane stress plastic zone with the thickness of the specimen. The original index (2) was a non-dimensional ratio defined as

\[
\beta = \frac{1}{B} (\frac{K}{\sigma_{ys}})^2 \quad \text{...........................................} \quad 3.19
\]

where \( B \) is the thickness of the specimen. Irwin (22) however prefers another index \( \alpha \) where \( \alpha = \frac{1}{\beta} \). In general, the simple rule is that plane stress fracture initiates under plane stress deformation when at fracture \( r_y \geq 2B \), that is,

\[
r_y = \frac{1}{\pi} (\frac{K}{\sigma_{ys}})^2 \geq 2B \quad \text{...........................................} \quad 3.20
\]
or $\beta > 2\pi$. Hahn and Rosenfield (32, 168) however showed experimentally that the plane strain plastic zone size, equal to $B/4$, is a reasonable estimate of the beginning of a significant shift away from plane strain.

It is noted that equation 3.18 contains the uniaxial yield stress, $\sigma_{ys}$, of the material, it might therefore be expected that the size and shape of the crack tip plastic zone would be affected by strain hardening properties of the material. The influence of strain hardening is schematically pictured in Figs. 3.10 (150) and 3.11 (168). Some examples of such influence are given by Gerberich (169) in Fig. 3.12.

![Image of plastic zones](image)

**Fig. 3.11:** Effect of strain-hardening exponent on plastic zone

![Image of materials](image)

**Fig. 3.12:** Examples of plastic zone with different values of $n$.

This observation is confirmed by Underwood et al. (170) using 1018 steel, 70-30 brass, 6061-T6 aluminium and 90-10 brass ($n = 0$, 0.01, 0.04 and 0.16 respectively).
3.6 EFFECT OF PLASTIC ZONE

The state of stress at the tip of a through-the-thickness crack is largely influenced by the thickness of the plate. To obtain the plane strain fracture toughness $K_{IC}$ of a material, the effect of plastic zone deformation must be maintained small with respect to the thickness of the specimen. Since plastic zone size is proportional to $(k/\sigma_{ys})^2$, it can be observed that a thicker plate is required of a material with low yield stress and high toughness than a material with high yield stress and low toughness. In the case of steel, the decrease in yield stress is generally found to be accompanied by the increase in toughness when the steel is tempered at progressively higher tempering temperature. Therefore, for fracture toughness testing of steel, larger specimen thickness is needed at higher tempering temperature.

It was described in section 2.3 that for a given material at a given temperature and testing speed, the fracture toughness is found to vary with the thickness of the specimen as shown in Fig.2.4. The fracture mode change, from plane stress to plane strain can be ascribed to the growth of the plastic zone at the leading edge of the crack (15,19,22,122). When the thickness of the plate is large, the size of the plastic zone is small with respect to the plate thickness and yielding in the thickness direction cannot operate easily due to the constraint by the surrounding elastic material. Consequently, the strain in the thickness direction is zero, resulting in a stress state of plane strain. Triaxial stress system dominates most part of the thickness, and thus the plate fails at a relatively lower stress. In the case of thin sheet, plastic zone is large compared to the plate thickness. The stress in the thickness direction tends to zero and so a biaxial or plane stress stress state occurs (31,35). The higher strain in such a specimen assists it to fail at increasingly higher external stress. When the thickness is increased to an optimum thickness, the toughness reaches its highest level as illustrated in Fig.2.4.

At intermediate thickness range, the plastic zone at the crack tip is neither, in relation to the plate thickness, large enough that a plane stress state can exist in the entire plate thickness, nor small enough that it can be neglected so that plane strain is assumed dominant. Instead the thickness is such that the plane stress region which exists near the plate surface and the central plane strain region are of comparable size. It is
due to this combination of different stress systems that the plate fails at an intermediate stress level. As the proportion of plane stress to plane strain region decreases with the increase in specimen thickness, failure stress gradually decreases until a limiting value is reached at $K_{IC}$ or $G_{IC}$.

At the optimum thickness, $B_0$ in Fig. 2.4, where the value of $K_c$ is maximum, the size of the plastic zone on the specimen surface in the $\theta = \pi/2$ direction is equal to half the specimen thickness $(112)$. Under this circumstance, the two planes both inclined at $45^\circ$ to the specimen surface as well as the loading direction are able to cut through the specimen within the plastic zone, forming either a full slant or V-slant fracture. Using equation 3.2, for example, $B_0$ can be estimated to be $(30,69)$

$$B_0 = \frac{1}{\pi} \left( \frac{K_{IC}}{\sigma_s} \right)^2$$

3.21
CHAPTER 4

YIELDING FRACTURE MECHANICS
CHAPTER 4

YIELDING FRACTURE MECHANICS

4.1 INTRODUCTION

The use of stress-intensity factors in elastic situations is justified when non-linear effects such as yielding are on a sufficiently small scale such that the elastic stresses at the vicinity of the crack tip are only slightly affected. This is usually the case when fracture occurs at a stress appreciably below the yield stress of the material and under the conditions of plane strain where the size of the plastic yield zone is small in comparison to the characteristic dimension of the crack. Under such circumstances, linear elastic fracture mechanics is useful and $K_{IC}$ provides a reasonable description of the crack tip stress field.

However, when yielding becomes extensive, the size of the plastic zone will no longer be small compared to the size of the crack. Consequently, a stress intensity factor corrected for plasticity effect does not provide an accurate failure prediction.

It has been generally considered that the plastic zone size will become significant if the net section stress is equal to the yield stress of the material. But in practice, it has been suggested that the accuracy of $K_{IC}$ is acceptable only when the average net section stress at instability does not exceed 80% of the 0.2% offset tensile yield stress (171-175). Recent evidence on the other hand shows that the net section stress should be less than two-thirds or 66% of the yield stress (124,176,177) for linear elastic fracture mechanics to apply.

Many attempts to characterize fracture in the presence of large scale yielding have been proposed. Notable amongst these are the techniques based on the $J$-contour integral by Rice (157,178), the crack tip opening displacement criterion proposed by Wells (179,180), and the semi-empirical equivalent-energy concept of Witt (181,182). They are briefly described in the following sections. Other less common techniques are the modification of the Dugdale model of strip yielding by Heald, Spink and Worthington (183), the curve fitting technique of Chell and Milne (184), both the semi-empirical method (185) and pseudo-elastic load/displacement curve fitting method (186) by Chell, the generalized theory of fracture mechanics by Andrews (187,188), the non-linear energy method by Liebowitz (189,190) and other experimental
methods involving the measurement of lateral notch root or crack tip contraction, angle of bend and stretch zone width.

4.2 CRACK TIP OPENING DISPLACEMENT

4.2.1 Concept of Crack Tip Opening Displacement

In section 3.2, it was observed that, in order to correct for plasticity occurring at the tip of a crack, the real physical crack was replaced by a slightly longer crack by adding on Irwin's plastic zone correction term. With the analysis due to Westergaard (12), it is possible to evaluate the displacement within the elastic crack in the direction of applied stress. Thence, the opening separation of the surfaces of the equivalent elastic crack at the point corresponding to the tip of the real crack can be obtained. This opening separation which is contained by the crack tip plastic zone, is known as the crack opening displacement or in abbreviation, C.O.D. It is believed that the crack opening displacement can be used as a measure of the work done in extending the crack even in situation where the amount of plastic flow at the vicinity of the crack is considerable (179).

Based on the analyses of Westergaard (12) and Muskhelishvili (18), Burdekin and Stone (13) showed that for plane stress condition, the value of C.O.D., \( \delta \), as shown in Fig. 4.1, is given by (equation 3.8)

\[
\delta = \frac{8}{\pi} \frac{\sigma_{ys}}{E} \ln \left[ \sec \left( \frac{\pi}{2} \frac{\sigma}{\sigma_{ys}} \right) \right] \quad \text{.................. 4.1}
\]
Wells (191) confirmed equation 4.1 by using finite element analysis. Hahn and Rosenfield (32), Bilby, Cottrell, Smith and Swinden (192) and Cottrell (193) also suggested a similar relationship.

Series expansion of equation 4.1 gives

$$\delta = \frac{8}{\pi} \frac{\sigma_{ys}}{E} a \left\{ \frac{1}{2} \left( \frac{\pi \sigma}{\sigma_{ys}} \right)^2 + \frac{1}{12} \left( \frac{\pi \sigma}{2 \sigma_{ys}} \right)^4 \right\} \quad \cdots \cdots \quad 4.2$$

When $$\left( \sigma / \sigma_{ys} \right)$$ is small, fracture occurs well before general yield, and equation 4.2 can be rewritten as

$$\delta = \frac{\pi}{E} \frac{\sigma^2}{\sigma_{ys}} a \quad \cdots \cdots \quad 4.3$$

Using equation 2.10, equation 4.2 can be further reduced to

$$G = \sigma_{ys} \delta \quad \cdots \cdots \quad 4.4$$

It can be seen, therefore, at low applied stress that the value of $$\delta$$ is related to the value of $$G$$ or $$K$$ in the linear elastic fracture mechanics. A critical $$\delta$$ value is to be expected if fracture occurs at critical $$G$$ or $$K$$ value.

Theoretically, equation 4.4 is only applicable to plane stress conditions. Under plane strain stress state, the relationship can be expressed as

$$G = \lambda \sigma_{ys} \delta \quad \cdots \cdots \quad 4.5$$

where $$\lambda$$ takes into account the constraint imposed by the plane strain loading.

Many values of $$\lambda$$ have been reported. For a non-strain hardening material, Rice (178) evaluated that $$\lambda = 1.48$$ while Rice and Johnson (194), 1.27, Levy et al. (154), and Hayes and Turner (195) respectively gave the value 2.14 and 2. On the other hand, Thornton (196), Rooke (197) and Parry and Mills (198) experimentally showed that $$\lambda=1$$. By infiltrating silicon rubber at the midsection of precracked three-point bend specimens and measuring the rubber casting for C.O.D., Robinson and Tetelman (199, 200)
evaluated that the value of $\lambda$ for material with work hardening exponents ranging from 0.033 to 0.2 to be equal to unity. Finite element analysis by Sumpter et al.\textsuperscript{(201)} suggested a value of 1.155. While the value of $\lambda$ is dependent on the location in which $\delta$ is measured, a value of 2 or 2.2 has been adopted by Green and Knott\textsuperscript{(202)}. This value is also suggested by recent finite element analysis\textsuperscript{(195,203)}.

Parallel to fracture toughness $K$ value, the C.O.D. of a given material is influenced by temperature, test piece geometry, triaxiality of stress and strain rate.

4.2.2 CRACK TIP OPENING DISPLACEMENT TESTING

In practice, C.O.D. testing involves mainly measuring a value of $\delta$ at the tip of a sharp crack usually in tough material where fracture occurs after general yielding. In early attempts, paddle type of C.O.D. meter\textsuperscript{(131)} was extensively used\textsuperscript{(204,205)}. This has been found to require complex instrumentation and the results are unreliable\textsuperscript{(206)}. Later development made use of double-notched specimens\textsuperscript{(207)} and notch root contraction\textsuperscript{(208)} which are well documented by Burdekin\textsuperscript{(209)}.

The most recent recommended procedure for C.O.D. testing\textsuperscript{(210)} is that established by Nichols et al.\textsuperscript{(206)}. This method makes use of single notched bend specimen in three point bending. The surface crack opening displacement, $V_g$, which is measured with a double cantilever beam clip gauge, is converted to the crack tip opening displacement using the relation

$$\delta = \frac{V_g}{1 + \left[ (a + z) / r(w - a) \right]}$$

where, as shown in Fig.4.2, $z$ is the thickness of the knife edge and $r$ is the rotational factor relating $V_g$ to $\delta$. Consider the plastic deformation associated with the bending, because the whole ligament is above yield, it may be treated as a plastic hinge with the centre of rotation at a distance $r(w - a)$ from the tip of the crack. This rotational factor has to be experimentally evaluated. Various values of $r$ have been found\textsuperscript{(206,211-213)} but for a 25mm thickness compact tension specimen and a value of $V_g$ greater than approximately $50 \times 10^2$mm, it is shown\textsuperscript{(214,215)} in Fig.4.3 that $r = 0.4$. 
Fig. 4.2: Relationship between crack opening displacement and knife-edge displacement

Fig. 4.3: Variation of rotational factor with knife-edge displacement for 25mm compact tension specimen

An alternative method for converting $V_g$ value to $\delta$ is based on a theoretical approach. The required equations are (210),

$$\delta = \frac{0.45 (W - a) \gamma_s}{0.45 W + 0.55a + Z} \left[ V_g - \frac{\gamma_s W (1 - v^2)}{E} \right] \quad \text{(4.7)}$$

for $V_g \geq 2 \gamma_s W (1 - v^2) / E$, and

$$\delta = \frac{0.45 (W - a) \gamma_s}{0.45 W + 0.55a + Z} \left[ \frac{V_g^2 E / 4 \gamma_s W (1 - v^2)}{E} \right] \quad \text{(4.8)}$$

for $V_g < 2 \gamma_s W (1 - v^2) / E$
where \( \gamma = \frac{V'E}{\sigma_{ys}} W(1-V^2) \) is a non-dimensionalized limiting value of elastic clip gauge displacement and \( V' \) is the limiting elastic clip gauge displacement.

In the C.O.D. Standard \(^{(210)}\), it is specified that the critical C.O.D. value should be determined at the instant of fracture. Usually this value is obtained from the load/displacement test record at maximum load \(^{(209)}\). However, slow crack growth usually occurs before maximum load is reached. Consequently it is found that C.O.D. value at maximum load, \( \delta m \), is dependent on specimen geometry \(^{(216)}\), loading mode \(^{(211,217,218)}\) and specimen thickness \(^{(219)}\). Knott \(^{(48)}\) suggested the use of C.O.D. value at the initiation of fracture, \( \delta_i \), which has been demonstrated to be a material constant \(^{(202,211,218)}\) relatively independent of specimen dimensions \(^{(216)}\). \( \delta_i \) bears many resemblances to the plane strain \( K_{IC} \) value. However, the experimental determination of \( \delta_i \) is generally tedious and expensive in terms of the number of test specimens that are required. Many techniques have been attempted, this includes the use of silicon rubber casting of the crack tip by Robinson and Tetelman \(^{(199,200)}\), the use of side-grooved specimens \(^{(220,221)}\) and the use of electrical potential drop \(^{(222)}\) and ultrasonic \(^{(218)}\) methods to detect the onset of ductile crack growth. Unambiguous value of \( \delta_i \) can also be obtained by unloading specimens from different locations along the load/displacement curve and fracturing them apart in liquid nitrogen to produce a change in fracture mode. The fibrous crack growth is then measured to yield a plot of C.O.D. against fibrous crack length. Thence, by extrapolating the plot to zero crack length, \( \delta_i \) can be determined.

In spite of the extensive investigation done on C.O.D., there is still doubt regarding its definition and practical significance \(^{(215,223)}\); even the less ambiguous \( \delta_i \) value is believed to be practically insignificant \(^{(224,225)}\).

4.3 THE J-CONTOUR INTEGRAL

4.3.1 Concept of the J-Contour Integral

It has been seen that when a specimen fractures after gross yielding, linear elastic fracture mechanics is invalidated. Therefore, in order to measure fracture toughness with excess plasticity, it is necessary
to devise a fracture criterion that is not based on the assumption of linear elastic behaviour. Such an attempt was made by Rice \((157,178)\) in defining the J-contour integral \((226)\), which for two-dimensional case, is given as

\[
J = \int_{\Gamma} (W d y - \bar{n} \frac{\partial U}{\partial x} ds)
\]

\[
\text{4.9}
\]

where \(\Gamma\) is any close contour surrounding the crack tip (Fig.4.4), \(T\) is the tension vector perpendicular to \(\Gamma\) in the direction of \(\bar{n}\), \(T_i = \sigma_{ij} n_j\), \(U\) is the displacement vector in the \(x\) direction and \(ds\) is an element of \(\Gamma\).

\[J = \int_{\Gamma} (W d y - \bar{n} \frac{\partial U}{\partial x} ds)
\]

\[
\text{4.9}
\]

\[W, \text{the strain energy density, is given by}
\]

\[
W = W(x,y) = W(\varepsilon) = \int_0^\infty \sigma_{ij} \varepsilon_{ij} \text{d}t
\]

\[
\text{4.10}
\]

Equation 4.9 can be shown \((178)\) to be path independent. This property of the J-integral is formally equivalent to the change in potential energy when the crack is extended. To express in more applicable terms, J-integral can be redefined as \((227)\)

\[
J = \int_{\Gamma} (\frac{\partial P}{\partial a} V) \text{d}V, \text{ or } J = \int_{\Gamma} (\frac{\partial V}{\partial a} P) \text{d}P
\]

\[
\text{4.11}
\]

where \(P\) is the force per unit length of crack front, and \(V\) is the displacement at the load point. Equation 4.11 expresses the relation between \(J\) and the rate of change with respect to crack length, \(a\), of the area under the load/displacement, \(P-V\), curve. As such \(J\) is a generalised relation
for the energy released when a crack is extended and thus, is applicable even when there is excessive plasticity at the crack tip. For small-scale yielding, the J-integral is identical to the strain-energy release rate, G, in plane strain (228,229):

\[ J_{IC} = G_{IC} = \frac{K_{IC}^2}{E} (1 - v^2) \]  

The J-integral approach, though appears promising for fracture at large scale yielding, has two major limitations (228). Firstly, under large plastic deformation, slow crack growth which always precedes fracture invalidates the path independency of the J-integral. Experimentally, however, it has been shown (230) that J-integral is path independent for large scale yielding up to an applied load level of \( \sigma_{applied}/\sigma_{ys} = 0.75 \), and even for unloading and rotations of stress and strain components with crack extension. Secondly, since the J-integral is only expressed in two dimensions, it is accordingly limited to plane strain and generalized plane stress cases only.

### 4.3.2 J-COUNTOUR INTEGRAL TESTING

The J-integral was first measured by Begley and Landes (228,229) with a compliance technique using compact tension specimens and bend bars. The procedure consisted of loading specimens containing cracks of different lengths. From the output load/displacement at loading points records, the area under the curves is determined at various displacements. These areas are converted to absorbed energy per unit specimen thickness, \( P \). Since from equation 4.11 (227),

\[ J = - \frac{dP}{da} \]  

the J-integral at a particular displacement can therefore be determined from

\[ J = - \frac{P_1 - P_2}{a_1 - a_2} \]

where \( P_1 \) and \( P_2 \) are the per unit thickness energy absorbed by specimens having crack length \( a_1 \) and \( a_2 \) respectively at a particular displacement.
The compliance technique of evaluating \( J \), as observed above, requires the use of several specimens. Estimation formulas (227) and procedures (231, 232) were later developed to evaluate \( J \) from a single specimen test using notch round, bend, centre-notched and compact tension specimens. A tentative \( J_{IC} \) uniform test procedure was proposed by Landes and Begley (233). The test requires the use of deeply notched, fatigue pre-cracked or bend type specimens which are unloaded from different points along the load/displacement curve. After marking the crack, with heat tinting, fatigue mark or dye penetrant, the specimen is fractured open and the maximum crack extension, \( \Delta a \), is measured. \( J \) is estimated from the load versus load point displacement record using the expression

\[
J = \frac{2A}{bB} \quad \text{................................. 4.15}
\]

where \( A \) is the area under the curve up to the point of unloading and \( b \) is the unfractured ligament. By constructing the plot of \( J \) versus \( \Delta a \) and the stretch zone line \( J = 2\Delta a\sigma_{\text{flow}} \), \( J_{IC} \) can be obtained from the intersection of the two curves. The stress, \( \sigma_{\text{flow}} \), is here taken to be the average of the yield and the ultimate stresses.

Measurements of \( J \)-integral using the compliance method and the estimation formula have been found to agree within about 15%. This has been demonstrated by Robinson (234) using three point bend specimen and centre cracked tension specimens of En32 steel and Yoder et al. (235) using titanium 6Al - 4V alloy. The method proposed by Landes and Begley (233) using estimation formula is rapidly gaining popularity. Good agreement of results between different testing laboratories have been achieved (236).

A procedure which is compatible with the linear elastic fracture mechanics and the C.O.D. testing to experimentally determine the critical value of \( J, J_c \), has recently been devised by Sumpter and Turner (237). The method makes use of the load/displacement record of a fracture testing of a three-point bend specimen with \( 0.25 < a/W < 0.60 \).
4.4 THE EQUIVALENT ENERGY METHOD

4.4.1 The Equivalent Energy Concept

The equivalent energy concept, proposed by Witt (238, 239) for predicting gross plastic fracture, is a simple empirical modelling method relating fracture in a large specimen to that in a geometrically similar model. The concept is useful in determining the lower bound $K_{IC}$ values. The technique can be understood as follows: Using a series of compact tension specimens, Witt postulated that when the load/displacement curves of geometrically similar specimens are normalized (by dividing the load quantity by the square of the specimen thickness and correspondingly dividing the displacement quantity by the specimen thickness), the curves will coincide, as shown in Fig. 4.5, with thicker specimens tending to fracture at lower points on the curve.

![Normalized load/displacement curve to maximum load for compact tension specimen](image)

Fig. 4.5: Normalized load/displacement curve to maximum load for compact tension specimen

The area under the normalized load/displacement curve to the point of fracture is known as the volumetric energy, which essentially is the energy per unit volume of the specimen. The basic value that characterizes fracture behaviour in the equivalent energy concept is the volumetric energy ratio which simply is the normalized energy of one testpiece tested to maximum load divided by the corresponding normalized energy of another geometrically similar testpiece again tested to maximum load. That is,
$S_{m,p} = \frac{E_m}{E_p}$ ................................................. 4.16

where $E_m$ and $E_p$ are respectively the normalized energy absorbed up to maximum load of the model and the prototype and $S_{m,p}$ is the ratio of volumetric energy between them. The equivalent energy concept is based on the hypothesis that this volumetric energy ratio is a unique variable function \(^{(240)}\). When the model and the prototype both behave in a frangible manner, the specimen size ratio is found to be the volumetric energy ratio \(^{(239,241,242)}\). In addition, Witt found that the variation of size effect versus specimen size is approximately a straight line \(^{(182,242,243)}\) as shown in Fig.4.6 and that by using compact tension specimens of thickness larger than 1 inch machined from ASTM A533 grade B class 1 steel plate, the volumetric energy ratio is independent of the specimen geometry but dependent only on specimen thickness \(^{(242,239)}\).

Consequently, the fracture of a large structure can therefore be modelled, using the equivalent energy concept, by extrapolation of the test results of a smaller model specimen.

\[ K_{ICd} = C \sigma \sqrt{\pi a S_{e}} \] .......................... 4.17

**Fig.4.6:** Size effects on energy-absorbing capacity of impact specimens \(^{(182)}\)

In terms of an elastic-plastic fracture mechanics procedures, Witt formulated that the fracture formula is given as \(^{(238,239)}\)
where $K_{ICd}$ is the fracture toughness of specimen thickness $d$, $S_{d,n}$ is the volumetric energy ratio between the model specimen of thickness $d$ and the structure of thickness $f$, $\sigma_f$ is the equivalent stress at maximum load determined by linearly extrapolating on the linear portion of the load/displacement record for the local stress with the flaw not present, $a_d$ is the size of the flaw scaled to a specimen of thickness $d$ and $C$ in the constraint or shape factor used in linear elastic fracture mechanics.

The equivalent energy concept is compatible with the linear elastic fracture mechanics (244) and good agreement between $K_{ICd}$ and $K_{IC}$ has been experimentally obtained (245,246). However, as the approach is empirical, its range of applicability is difficult to assess (244). Its usefulness is nevertheless directly proportional to the ease with which the volumetric energy ratio can be determined (240). If for a material the volumetric energy ratio depends on the thickness of the test specimen and the testing temperature, the approach should be applicable.

### 4.4.2 Equivalent Energy Concept Testing

The procedure to obtain $K_{ICd}$ value from standard fracture toughness tests is documented in References (238,247 and 248). Basically, the method makes use of two quantities on the load/displacement record of a fracture test, namely, as shown in Fig.4.7, the area $A_2$ under the curve to any point $P_B$ on the linear portion of the curve and the area $A_1$ under the curve to the point of maximum load, $P_{max}$.

![Typical nonlinear load/displacement curve for a compact tension specimen](image-url)
The fracture toughness property of the material tested in thickness d can then be calculated from the following expression:

\[ K_{ICd} = \frac{b^2 p_B}{bdv(2bd)} f\left(\frac{a}{w}\right) \]  \hspace{1cm} \text{(4.18)}

where \( b \) is the ratio of the area \( A_1 \) divided by the area \( A_2 \) and \( f\left(\frac{a}{w}\right) \) is obtainable from the ASTM standard \(^{118} \). For standard compact tension specimen in which \( W = 2d \), equation 4.18 can be rewritten as \(^{245} \)

\[ K_{ICd} = \frac{p_B \sqrt{A_1/A_2}}{d \sqrt{w}} f\left(\frac{a}{w}\right) \]  \hspace{1cm} \text{(4.19)}

### 4.5 CONCLUSION

It is noted that the crack opening displacement, \( \delta \), the \( J \)-contour integral and the equivalent energy concept of \( K_{ICd} \) are not only compatible, but they can also be estimated from the test record of a single test.

As both the C.O.D. and the \( J \)-integral attempt to characterize ductile fracture process by using single parameters which may be related to the critical energy release rate, it is apparent that a relationship exists between them. Finite element analyses \(^{154,195,249} \) show that the relationship is of the form

\[ J = M \sigma_y \delta \]  \hspace{1cm} \text{(4.20)}

where \( M \), although not normally found to be constant, has a value that lies between 1.15 and 2.98 \(^{250} \). Equation 4.20 appears to be material dependent, affected possibly by the work hardening rate \(^{234} \). On the other hand, Turner and Burdekin \(^{225} \) suggested that the value of \( M \) in the above relation is likely to depend on the definition used for \( \delta \).

The compatibility of the \( J \)-integral and the equivalent energy concept, although not applicable to all load/displacement curves in general, can be observed to be justified for deeply notched bend bars and compact tension specimens \(^{244} \). Using compact tension specimens of A533 grade B class 1 low-alloy steel (HSST plate 04), Riccardella and
Swedlow (245) found that both the J-integral and the equivalent energy approaches gave highly consistent results in spite of the wide range of specimen sizes. This is confirmed by Chell and Worthington (246) using Ducol W30A steel for compact tension specimen with a/W > 0.5. Using precracked Charpy specimens, however, Robinson and Tetelman (251) found that the equivalent energy approach produced unreliable estimation of $K_{IC}$. This is consistent with the results of Begley and Landes (244).
CHAPTER 5

EXPERIMENTAL PROCEDURES
CHAPTER 5

EXPERIMENTAL PROCEDURES

5.1 INTRODUCTION

The aim of the present investigation was to study the size of the shear lips as a function of the material fracture toughness and yield stress. The parameters were altered in the case of steel, by the process of tempering the material to various temperatures.

The plane strain fracture toughness data of the materials were obtained using compact tension specimens and test method which was in accordance to the recommended ASTM standard E399-74(118). In cases where yielding was believed to be excessive, yielding fracture mechanics approaches were employed to evaluate the fracture toughness parameter of the material.

The standard properties of the materials were investigated by means of tensile tests using specimens machined from the compact tension specimens.

5.2 MATERIAL

5.2.1 Steel

The specimen material used in the investigation was Comsteel En 25, a high-strength low-alloy steel manufactured and supplied by Commonwealth Steel Company Ltd. It was supplied in the form of 125mm diameter black bar and in the as-received condition, was heat treated and hardened to 950 - 1100 MPa tensile strength. Two bars of four feet length were furnished for this work. Their nominal composition is shown in Table 5.1 while the complete analysis is tabulated in Appendix B.

<table>
<thead>
<tr>
<th>Element</th>
<th>C</th>
<th>Mn</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>Weight Per Cent</td>
<td>0.31</td>
<td>0.58</td>
<td>2.33</td>
<td>0.60</td>
<td>0.52</td>
</tr>
</tbody>
</table>

Table 5.1: Nominal Composition of Comsteel En 25
It was found that the sulphur and phosphorus contents of the two bars were slightly different. As a result, it was necessary to differentiate the two steel compositions by denoting that with the higher sulphur and phosphorus contents composition O while the other steel, composition N (as can be seen in Appendix B) in situations where the experimental data were particularly dependent on steel compositions. The silicon content of steel composition N was higher than expected for this grade of steel.

5.2.2 Aluminium

One lot of 160mm thick 7075-T6 high strength commercially produced aluminium plate was employed. The composition of the alloy was analysed by New Zealand Aluminium Smelters Limited in accordance with ASTM E 227-67 (1972) and E 406-70. This is shown in Table 5.2 and tabulated in detail in Appendix C. The alloy was

<table>
<thead>
<tr>
<th>Element</th>
<th>Cu</th>
<th>Mg</th>
<th>Zn</th>
</tr>
</thead>
<tbody>
<tr>
<td>Weight Percent</td>
<td>1.46</td>
<td>2.51</td>
<td>5.76</td>
</tr>
</tbody>
</table>

Table 5.2: Nominal composition of 7075-T6 aluminium.

selected for the present study because it is widely used for high strength structural application such as aircraft and missile components. Moreover, knowledge of its fracture toughness characteristics in various crack orientations and specimen thicknesses is easily available from the literature (for example, References 33, 65, 100). The alloy, in addition, offers metallurgical properties that are different from those of Comsteel En 25 and hence this serves as a means of comparison as well, permits generalization of the conclusion of the investigation with a reasonable degree of confidence. It was later found that the alloy was in the as-cast condition, as a result the effect of anisotropy was unable to be studied.

5.3 SPECIMEN GEOMETRY

Specimens currently recommended in the ASTM (118) or British Standard (119) are the three-point bend specimen and the compact tension specimen. With the materials being supplied in round bar of 125mm diameter for Comsteel En 25 and blocks of 80 x 100 x 160mm for 7075-T6
Specimen thickness, B = 25mm .......... A
22mm .......... B
19mm .......... C
16mm .......... D
13mm .......... E
10mm .......... F
7mm .......... G
4mm .......... H

Fig.5.1: Dimensions of compact tension specimen
aluminium alloy, the choice of using compact tension specimen was obvious for reasons that,

(a) it offered a more efficient use of the material,

(b) heat treatment could be easily carried out due to its length being shorter,

(c) lower load was required in testing,

(d) different crack plane orientations were obtainable without difficulty,

(e) specimens of small thickness could be tested without buckling.

Dimensions of the compact tension specimen are shown in Fig.5.1. The specimen thickness, B, varied over a range from 4mm to 25mm with a 3mm incremental interval with the exception of steel of composition N where only 25mm thickness specimens were used. The specimens were machined such that the path of the fracture was parallel to the rolling direction of the bar or plate. For the steel specimens, according to the ASTM standard(118), the crack plane orientation is designated C-L or R-L while for the aluminium specimen, T-L. To investigate the effect of anisotropy on fracture toughness of the materials, however, R-C and L-R orientations for steel and S-T and S-L orientations for aluminium were also tested using 25mm thickness specimens.

The stress intensity at the tip of the crack for compact tension specimen was calculated using the expression

\[ K_I = \frac{P}{BW^{1/2}} f \left( \frac{a}{W} \right) \] .............................. 5.1

where

- \( K_I \) = stress intensity factor in opening mode (mode I) cracking, MPa\( \sqrt{m} \)
- \( P \) = load, N,
- \( B \) = specimen thickness, mm,
\[ W = \text{specimen depth, mm}, \]
\[ a = \text{crack length, mm}. \]
\[
\begin{align*}
 f\left(\frac{a}{W}\right) &= 29.6\left(\frac{a}{W}\right)^{1/2} - 185.5\left(\frac{a}{W}\right)^{3/2} + 655.7\left(\frac{a}{W}\right)^{5/2} \\
&- 1017.0\left(\frac{a}{W}\right)^{7/2} + 638.9\left(\frac{a}{W}\right)^{9/2}
\end{align*}
\]

To facilitate calculation of \( K_I \), the values of \( f\left(\frac{a}{W}\right) \) are graphed in Appendix D with values of \( a/W \) between 0.45 and 0.55. This is also found in tabulated form in References (118,119).

5.4 SPECIMEN PREPARATION

5.4.1 Machining

The compact tension specimens were machined to the dimensions shown in Fig.5.1 with the materials in the as-received condition. All dimensional tolerances were in accordance to the ASTM standard (118). A straight through notch was used not only because this offered simpler machining than the chevron notch but also because severe quench cracking was found to occur frequently in the Comsteel specimens (255) with chevron notches. The notches were formed by climb milling with a 3.13mm wide slitting saw. The notch root radii measured using a Nikon profile projector were found to be about 0.15mm. The specimen surfaces were surface ground and the loading-pin holes were bored and reamed. Two 3mm tapped holes at 12mm apart across the crack starter (see Fig.5.1) were used to secure the knife edges whereby the double cantilever beam clip gauge (see section 5.5.1) could be attached. As the specimen thickness tolerance was less critical, to reduce machining time and cost, it was relaxed to \( \pm 0.125 \)mm for all specimens.

5.4.2 HEAT TREATMENT

5.4.2.1 Steel Specimen

The specimens were austenitized at 850°C for one hour in a molten salt bath. The bath temperature was controlled by means of a chromel-alumel thermocouple wired to an Ether controller, and was constantly checked using independent chromel-alumel thermocouple wired
to a Honeywell potentiometer. The temperature was maintained to within \( \pm 10^\circ C \) throughout the entire operation. After this, the specimens were quenched into a flowing column of oil. The oil bath was designed such that oil was pumped through a vertical P.V.C. tube of 75mm diameter. The upward flow of the oil cools the quenched specimen uniformly, hence reduces thermal gradients during the cooling process. This however did not completely eliminate quench cracks especially in specimens machined from steel with composition N (see Appendix B).

After the quench, apart from the as-quenched specimens, all other specimens were tempered in molten nitrate salt baths at one of the following temperatures: 200, 300, 400, 450, 500, 550, 600, and 650\(^\circ\)C while the 100\(^\circ\)C temper was carried out in an oil bath. The tempering temperatures were maintained constant by means of chromel-alumel thermocouples attached to Ether controllers and again rechecked as described above. The temperature variation was minimised to within \( \pm 5^\circ C \). All specimens were tempered for one hour. After being cooled in the air, the specimens were washed in warm water to remove all traces of salt.

In the case of the specimens with steel composition N, the 600 and 650\(^\circ\)C tempering process was carried out in an electric muffle furnace. To prevent oxidation, the specimens were wrapped in two layers of type 321, 0.075mm thick stainless steel foil to form an airtight enclosure. The temperature of the specimens was measured by a chromel-alumel thermocouple connected to a Honeywell potentiometer. The furnace temperature was monitored by an independent thermocouple wired to an Ether controller.

After the tempering process, the side faces of the specimens were reground to remove decarburization and the slight central bulge caused by the change in volume associated with the transformation from austenite to martensite during the quench.

5.4.2.2 Aluminium Specimen

In the as-received T6 temper condition the alloy is solution heat treated and artificially aged where cold work from flattening or straightening may not be recognized in applicable specifications. No further heat treatment to the alloy was performed.
5.4.3 Fatigue Cracking

Fatigue cracking of the specimens was carried out at the New Zealand Pottery and Ceramics Research Association, Lower Hutt using an M.T.S. electrohydraulic structural testing equipment. The control console of the testing machine is shown in Fig. 5.2. Fig. 5.3 shows the arrangement of the equipment for fatigue cracking. This consisted of, going from top to bottom of Fig. 5.3, a universal joint to take up minor misalignment, clevis and compact tension specimen (the clip gauge was required only in the fracture test), adapter and load cell. The clevis which was built in accordance with the ASTM Standard(118), after the design of Bubsey et al.(257), had a slot width of 25.1 mm to accommodate specimens of thickness $B = 25$ mm. For specimen thickness less than 25 mm, brass spacers were used so that the specimen was kept at the centre of the clevis. The clevis and the loading pins were made from AISI H11 tool steel hardened to give a tensile strength of about 1700 MPa.

In this investigation, all fatigue cracking was performed with continuous tension cycling under load control. A high initial load not exceeding 70% of $K_i$ value was used to start the fatigue crack. The fatigue crack extension of the final 2.5% of the overall length of the notch plus crack was done strictly according to the requirements of the ASTM Standard B399-74(118). The frequency used was in the range of 2 to 5 Hertz using the M.T.S. hydraulic actuator and 8 to 10 Hertz with the MAND actuator depending on the crack length of the specimen and the loading range. The total number of cycles required to grow the fatigue crack varied from 10,000 to 34,000 for the steel specimens and from 7,500 to 21,000 for the aluminium specimens.

5.5 COMPACT TENSION SPECIMEN TESTING

5.5.1 Clip Gauge Displacement Measurement

The surface crack opening displacement across the notch of the specimen was measured using a clip gauge. The clip gauge, as shown in Fig. 5.4, was constructed following the design of Fisher et al.(258) as recommended by the standards(118,119). A titanium alloy (Ti-8Mo-8V-2Fe-3Al) which has a high yield stress to elastic modulus ratio, $\frac{\sigma_{ys}}{E}$, was used for the cantilever beams while for the spacer block, aluminium was used to reduce the cantilevered weight.
Fig. 5.4 Double cantilever beam clip gauge.

Fig. 5.6 Clip gauge calibration equipment.
Strain gauges of 120 ohm (Kyowa type KFC-5-C1-11) were attached onto each side of the cantilever beams close to the spacer block using Cyanoacrylate glue, CC-15A. The gauges were then coated with low-melting-point wax to reduce the effect of moisture.

Fig. 5.5 shows the bridge network used to connect the strain gauges on the cantilever beams.

![Bridge Network Diagram](image)

**Fig. 5.5:** Clip gauge bridge network
The strain gauge network was excited by 3.00 volts D.C., supplied by a Hewlett Packard type 6218A power supply. The sensitivity of the clip gauge was about 1.4 mV/V per millimeter of the gauge opening.

Before the clip gauge was used, it was checked for linearity using slip-gauges in a calibration rig as shown in Fig.5.6. Readings were taken at eleven equally spaced intervals over the working range of the gauge from 4.95 to 6.35mm. As can be seen from Appendix E, the linearity of the clip gauge was better than the value of 0.0025mm maximum deviation required in the standards (118,119).

The clip gauge was mounted on the specimen using two attachable knife edges screwed to the specimen as shown in Fig.5.7. The spacing between the knifed edges was 5mm. This was established by inserting a machined spacer between the knife edges while tightening the screws. The knife edges were made following the design recommended by the ASTM standard (118).

5.5.2 Load Measurement

The load cell, attached permanently to the end of the hydraulic actuator, consisted of a four active arm strain bridge which had an output signal proportional to the applied load. The output from the load cell was fed into the M.T.S. control console and amplified to give ±10.00 volts output for each loading range. The system was calibrated in each range using a shunt calibrator resistor.

The specifications of the load cells are given in Table 5.3.

<table>
<thead>
<tr>
<th></th>
<th>M.T.S. Actuator</th>
<th>MAND Actuator</th>
</tr>
</thead>
<tbody>
<tr>
<td>Rated capacity, KN.</td>
<td>670</td>
<td>100</td>
</tr>
<tr>
<td>Bridge resistance, ohm.</td>
<td>350</td>
<td>350</td>
</tr>
<tr>
<td>Linearity, % of rated capacity</td>
<td>±0.13</td>
<td>±0.15</td>
</tr>
<tr>
<td>Output, mV/V at rated capacity</td>
<td>2.367</td>
<td>2.473</td>
</tr>
</tbody>
</table>

Table 5.3: Specifications of load cells
Fig. 5.7  Compact tension specimen with clip gauge and knife edges.

Fig. 5.8  (a) Standard calibration block  
(b) Metal block for load calibration
5.5.3 Fracture Test

Essentially, the fracture test involved loading the fatigue pre-cracked compact tension specimens to fracture so that a plot of the load versus the displacement across the notch at the specimen edge could be autographically recorded on an X-Y recorder.

After the load cell was calibrated, a metal block was loaded in the clevis to calibrate the Y axis of the M.F.E. 815M X-Y recorder. This was done by loading the block to a predetermined load to correspond to the full scale load on the Y axis. The output signal was fed into the recorder and the recorder gain was adjusted to give full scale deflection (25cm). The system was then unloaded. The metal block removed and the compact tension specimen to be tested was placed in the clevis.

The clip gauge was then mounted and the clip gauge bridge network balanced. The output was fed into the amplifier in the M.T.S. control console and from this to the X-axis of the X-Y recorder.

The compact tension specimen was manually cycled a few times up to a load less than the maximum load used in the final stage of the fatigue cracking. This was done to ensure that the clip gauge was seated properly in the knife edges and that an initial slope of about 1.5 was obtained for the load/displacement record. The X-axis of the recorder was then calibrated by mounting the clip gauge on a series of standard blocks (Fig.5.8) in which the knife edges were accurately spaced at 0.200, 0.210, 0.220, 0.240 and 0.250 in.

The specimen was finally loaded to failure by winding the set-point manually such that the rate of increase of stress intensity of the specimen was within the range of 0.55 to 2.75 MPa√m/sec recommended by the ASTM Standard (118).

5.5.4 Fracture Toughness Calculation

Specimen thickness, B, was measured with a micrometer at three positions between the notch and the unnotched edge of the specimen before it was fractured. The average of the three readings was used in the fracture toughness calculation.
The depth, $W$, of the specimen was determined using a Nikon profile projector at 10x magnification. $W$ was taken as the average of the distances from the unnotched edge of the specimen to the near edge of the loading-pin hole and the far edge of the same hole. Measurement was made for each loading-pin hole on both sides of the specimen and the average of the four values was recorded.

After the fracture test, the fracture surfaces of the specimens were photographed using a Carl Zeiss C35M Tessovar photomacrographic system. With the photograph magnified about 7 to 8 times, the distance from the unnotched edge of the specimen to the tip of the fatigue crack was measured at three positions: at the centre of the crack front, and midway between the centre and the end of the crack front on each side. The average of the three measurements was subtracted from the specimen width, $W$, to obtain the crack length, $a$.

To establish that a valid $K_{IC}$ value had been determined, it was necessary first to calculate a conditional result, $K_{Q}$, which involved a construction on the load/displacement record obtained from the fracture test. The procedure, illustrated in Fig. 5.9 was as follows:

![Diagram of load/displacement records](attachment:image.png)

**Fig. 5.9:** Types of load/displacement records for $K_{IC}$ determination

A tangent OA to the initial linear part of the test record was first drawn. A secant line $OP$ from the origin of the test record with
the slope 5% less than that of tangent OA, i.e. slope \((P/V)_o = 0.95(P/V)_o\) where \((P/V)_o\) is the slope of OA, was then constructed. If the load at every point on the test record which preceeded \(P_5\) was lower than \(P_5\) as illustrated in Type I of Fig 5.9, then \(P_5\) was taken as \(P_Q\). However, if there was a maximum load preceding \(P_5\) which exceeded it, then \(P_Q\) was defined by this maximum load (Fig 5.9 Type II and III).

From the values of \(B, W, a\) and \(P_Q\), a conditional value of the fracture toughness, \(K_Q\), of the specimen was then determined using the expression

\[
K_Q = \frac{P_Q}{BW^{1/2}} f\left(\frac{a}{W}\right) \quad \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots 5.2
\]

where

\[
f\left(\frac{a}{W}\right) = 29.6 \left(\frac{a}{W}\right)^{1/2} - 185.5 \left(\frac{a}{W}\right)^{3/2} + 655.7 \left(\frac{a}{W}\right)^{5/2}
- 1017.0 \left(\frac{a}{W}\right)^{7/2} + 638.9 \left(\frac{a}{W}\right)^{9/2}
\]

A plot of \(\frac{a}{W}\) versus \(f\left(\frac{a}{W}\right)\) for specific values of \(\frac{a}{W}\) between 0.450 and 0.550 is presented in Appendix D.

5.5.5 Rejection Criteria

The calculated conditional fracture toughness, \(K_Q\), was considered to qualify as a valid \(K_{IC}\) value if the following criteria laid down in the ASTM Standard E399-74 (118) were satisfied:

1) Loading -

\[
P_{max}^{\frac{1}{P_Q}} \quad \text{did not exceed 1.10, where } P_{max} \text{ was the maximum load that the specimen was able to sustain.}
\]

2) Fatigue crack stress intensity at terminal 2.5% of the overall length of notch plus crack -

\[
K_f(max) \quad \text{did not exceed } 0.60 K_Q \quad \text{and}
K_f(max)/E \quad \text{did not exceed } 0.00032m^{1/2}
\]
3) Shape of fatigue crack -

a) the maximum difference between any two of the three crack length measurements described in section 5.5.4 did not exceed 5% of the average crack length,

b) any part of the crack front was not closer to the machined notch than 5% of the average crack length, a, or 1.3mm minimum,

c) the length of either surface trace of the crack was more than 90% of the average crack length,

d) the crack plane was parallel to both the specimen width and thickness directions within ±10°; and

e) the average crack length, a, was between 0.45 and 0.55 times the depth, W, of the specimen.

4) Specimen size -

both the average crack length, a, and the specimen thickness, B, were greater than 2.5 \((K_c/\sigma_{ys})^2\) where \(\sigma_{ys}\) was the 0.2% proof stress.

5.5.6 \(K_c\) Calculation

The values of \(K_c\) for both the steel and the aluminium specimens were determined from the maximum load and the corresponding crack length. The maximum load sustained by the specimen and the corresponding displacement were noted on the load/displacement record. From the dimensionless compliance of the specimen, BEV/P, where B is the specimen thickness, E is the Young's Modulus of the specimen material, V is the displacement at maximum load, P, an effective dimensionless crack length ratio, a/W, was
evaluated (the variation of BEV/ $P$ versus $a/W$ at the load line and at
the knife edges $^{(259,260)}$ is plotted in Appendix F). The value of $K_c$
was calculated with this effective $a/W$ ratio using the following
expression from the ASTM Standard $^{(118)}$ (equation 5.2):

$$K_c = \frac{P}{B\sqrt{W}} f\left(\frac{a}{W}\right) \quad \text{5.3}$$

where $f\left(\frac{a}{W}\right)^*$, obtainable from the Standard, corresponded to the effective
$a/W$ ratio.

5.6 CALCULATION OF YIELDING FRACTURE MECHANICS PARAMETERS FROM FRACTURE
TOUGHNESS TEST RECORD.

5.6.1 Crack Opening Displacement Approach

From the fracture toughness test record of load/displacement,
Fig.5.10, the surface clip gauge displacement, $V_g$, at point C where the
maximum load was first attained, was measured. $V_g$ was then converted to
crack opening displacement, $\delta$, using one of the following equations
(equation 4.7 or 4.8):

$$\delta = \frac{0.45(W-a)}{0.45W+0.55a+Z} (V_g - V') \quad \text{for } V_g \geq 2V' \quad \text{5.4(a)}$$

or

$$\delta = \frac{0.45(W-a)}{0.45W+0.55a+Z} \left(V_g^2/4V'\right) \quad \text{for } V_g < 2V' \quad \text{5.4(b)}$$

where $V'$ is the limiting elastic clip gauge displacement. For compact
tension specimen, $Z$ was taken as the sum of the distance from the centre
of the loading pins to the notched face of the specimen and the thickness
of the knife edge $^{(214)}$. $V'$, the limiting elastic clip gauge displacement
was obtained from the load/displacement record, shown in Fig.5.10, as the
distance OF. Line OEDA is the tangent to the linear portion of the load/
displacement record.
5.6.2 J-Integral Approach

As the J-integral involved the use of the area under the load/displacement curve that represented the total work input to the specimen, it was necessary to redraw the test record obtained during the fracture test (section 5.5.3) so that the displacement was measured at the load point. This was carried out using the theoretical boundary collocation of the specimen tested\(^{(231)}\). The ratio of the displacement at load line, \(V_{LL}\), over the displacement at the knife edges, \(V_{KE}\), is determined from Appendix F where the compliance of the specimen, \(\frac{BEV}{F}\), at the load line and at the knife edges is plotted against \(a/W\). A new load/displacement at load line curve is redrawn from point O of Fig.5.10 to the point of maximum load C, as shown in Fig.5.11.

The area, \(A\), under the load/displacement at load line curve \(OE'C'H'\) was measured with a planimeter. The J-integral at maximum load was then calculated from the expression,

\[
J = \frac{2A}{B(W - a)}
\]
5.6.3 Equivalent Energy Approach

The area, $A_2$, under the load/displacement record (area OEC in Fig. 5.10) to maximum load at C and the area, $A_1$, to any load, P, on the linear portion of the test record (area OPG in Fig. 5.10) were measured using a planimeter. The fracture toughness, $K_{ICd'}$ was then calculated using the expression

$$K_{ICd} = \frac{P\sqrt{A_2/A_1}}{d\sqrt{W}} f\left(\frac{a}{W}\right) \quad \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots 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5.7 MEASUREMENT OF CRACK OPENING DISPLACEMENT AND J-CONTOUR INTEGRAL

AT CRACK INITIATION

5.7.1 Specimen Testing

Compact tension specimens of 25mm thickness were used to evaluate the value of crack tip opening displacement at initiation, $\delta_i$, and $J_{IC}$. The specimens were austenitized at 850°C for one hour in a salt bath, quenched in oil and then tempered for one hour at 600°C.
The specimens were fatigue cracked as outlined in section 5.4.3.

After the X and Y axes of the M.F.E. 815M X-Y recorder had been calibrated using the double cantilever clip gauge together with the standard blocks (Fig.5.8) and the metal block (described in section 5.5.3) respectively, the specimens were loaded up and unloaded from different points along the load/displacement curve as shown schematically in Fig.5.12.

Fig.5.12: Load/displacement record showing unloading points

Crack extension that occurred in the specimen was marked by heat tinting the crack or by fatiguing after unloading. Heat tinting was carried out by heating the specimen to 300°C for 30 minutes in a salt bath, while fatigue marking was performed using a stress intensity less than 60% of the $K_Q$ value. The specimen was subsequently broken apart on a Universal Instron testing machine, model TT-D.

5.7.2 $\delta_1$ Analysis

The crack extension, $\Delta a$, on the fracture surface was measured with a travelling microscope. $\Delta a$ was taken to be the maximum distance from the prefatigue mark to the end of the heat tint mark or fatigue mark.
The fracture surfaces of all the specimens tested were photographed for permanent record.

To evaluate $\delta_i$, the crack opening displacement at the point of unloading, calculated as described in section 5.6.1 was plotted against the measured crack extension, $\Delta a$. By extrapolating the C.O.D. versus $\Delta a$ curve to zero crack extension, $\delta_i$ was determined unambiguously.

5.7.3 $J_{IC}$ Analysis

The original load/displacement record to the point of unloading was firstly redrawn using the procedure described in section 5.6.2 such that a load/displacement at loading line curve was obtained. The area under the redrawn curve was then measured with a planimeter and the value of $J$ calculated using the estimation formula

$$J = 2A/B(W - a) \quad \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldotted
5.9 **Below Room Temperature Fracture Testing**

Compact tension specimens were fractured at temperatures below room temperature in an attempt to reduce the size of the shear lips formed at the free surfaces. Three aluminium and six steel specimens were tested. The steel specimens were austenitized at 850°C and tempered at 600°C while the aluminium specimens were tested in the as-received condition.

The test temperatures used were 0°C, -78°C and -196°C. The 0°C temperature bath was obtained using a mixture of ice and water while the -78°C bath, a mixture of solid carbon dioxide and acetone. At these testing temperatures, both the compact tension specimen and the clevis were submerged in the mixture which was contained in a 10cm deep x 10cm diameter P.V.C. tank. At -196°C, in order to avoid the possibility of damaging the clevis and the loading pins, only the specimens were submerged in the liquid nitrogen bath. The specimen was rapidly transferred onto the clevis where it was pulled to fracture. In all cases, submersion time was over one hour.
The fracture testing was performed on a 100KN capacity Instron testing machine, model TT-D using a GR load cell. The crosshead speed used was 0.02 in/min (0.5mm/min). The C.O.D. and load outputs were recorded on a model 7004B Hewlett Packard X-Y recorder. The C.O.D. was monitored with a clip gauge (see section 5.5.1) while the load was directly fed from the console of the Instron testing unit.

5.10 TENSILE TESTING

5.10.1 Specimen Preparation

Tensile specimens were machined from the fractured half of a compact tension specimen as indicated by the dotted lines in Fig. 5.14.

![Tensile Specimen Diagram](image)

Fig.5.14: Method of obtaining tensile specimen

The axis of the specimen was normal to the direction of rolling and the crack plane direction. The specimens were ground under flood coolant to the dimensions shown in Fig.5.15. At least two specimens for each tempering temperature and for each batch of material were machined and tested.
Fig. 5.15: Dimensions of tensile specimen from fractured compact tension specimen

Due to the short length of the tensile specimens obtained from the compact tension specimens, an extensometer could not be used to accurately measure the elongation during testing so that the elastic modulus and the strain hardening exponent of the material could not be obtained. This was overcome by using longer tensile specimens, as shown in Fig. 5.16.

Fig. 5.16: Dimensions of tensile specimen for work-hardening exponent determination.

These specimens which had the same orientation as those above were, for Comsteel En25, machined from a 125mm diameter bar and for 7075-T6, a 160mm thickness block. The heat treatment procedure for the Comsteel
En25 specimens was the same as that for the compact tension specimens, that is, austenitize at 850°C for one hour, quench into oil then temper for one hour at one of the following temperatures: 100,200,300,400, 450,500,550,600 and 650°C. Two specimens were tested at each tempering temperature and for each batch of material.

5.10.2 Tensile Test

All tensile tests were carried out at room temperature on a Universal Instron testing machine, model TT-C, using a FR load cell. The crosshead speed was 0.005 in/min (0.127 mm/min) and the load versus machine extension was autographically recorded on a chart using a chart speed of 0.5 in/min (12.7 mm/min). The 0.2% proof stress, the ultimate tensile stress, the percentage elongation to fracture, the Young's modulus of elasticity and the strain hardening exponent (described in Appendix G) were determined from the tensile test record. The specimen reduction in area was measured with a Hounsfield Reduction-in-Area-Gauge.

In testing the long tensile specimens, a standard Instron strain gauge extensometer, model G-51-16, was used. The initial gauge length was 13.3mm. The opening of the extensometer was calibrated on the recording chart in the Instron console using a drum micrometer while the zero setting was adjusted with an oscilloscope.

The procedure used in testing tensile specimens at temperatures below ambient was identical to that described above. In low temperature tests, the specimen was submerged in a tank of ice water mixture or acetone dry ice mixture or liquid nitrogen for at least 30 minutes to attain a temperature of 0°C, -78°C and -196°C respectively.

5.11 SHEAR LIP SIZE MEASUREMENT

Like the measurement of crack length described in section 5.5.4, the shear lip size measurements were taken from the enlarged fracture surface photographs. The size of the shear lip, schematically shown in Fig.5.17, was defined as the distance from the surface of the specimen to the interface between the slant and the square fracture, C, on the fracture surface. Measurements were made on both slant surfaces of a fractured half of a specimen and at three locations: one quarter,
one half and three quarters the distance from the tip of the fatigue
crack to the unnotched edge of the specimen. The average of the six
measurements was recorded.

![Diagram of shear lip size and square fracture]

**Fig.5.17: Determination of shear lip size**

5.12 HARDNESS MEASUREMENT

Hardness measurements on all specimens were carried out on an
Avery testing machine using the Rockwell C scale for the Comsteel En25
specimens and Rockwell B scale for the 7075-T6 aluminium specimens.
The surfaces were machine ground before testing. A minimum of four
hardness tests were taken for each specimen. Generally the variation
in hardness for a particular specimen was 1.5 HRC for the steel
specimens and about 3 HRB for the aluminium specimens.

Variations in the performance of the hardness testing machine
throughout the entire investigation were corrected by taking hardness
tests on a standard block before each test period. This variation was
observed to be 1.5 HRC for the steel specimens. Since only one batch
of aluminium alloy specimens was tested, this variation was entirely
eliminated.
5.13 MICRO-HARDNESS MEASUREMENT

Micro-hardness measurements were taken on selected specimens at the shear lip and the surrounding area (region ABC in Fig. 5.17) using a Tukon micro-hardness tester. The specimens were sectioned in a spark cutter and polished before testing. Using a 136° diamond pyramid indentor and a 1kg load, measurements were made in a grid system of 0.2mm x 0.4mm.

5.14 OPTICAL MICROSCOPY

Selected specimens were sectioned, polished, etched and examined on a Reichert MeF2 microscope. To study the deformation of the grains in the shear lip regions, polished specimens of Comsteel En25 that had been tempered up to 300°C were etched in a 50% diluted solution of 15g picric acid and 213g "Teepol" at 75°C. At higher tempering temperatures, the etchant used was a solution of 100 ml of saturated picric acid, 2ml "Teepol" and 6 drops conc. HCl (261). The primary role of the "Teepol" was to act as a "wetting" reagent. Before being polished and etched, the specimens were nickel plated so that deformation of the grains close to the fracture surface could be observed. The plating process was carried out using an electrolyte of nickel sulphate and boric acid with pure nickel foil as the anode.

Standard Fry's reagent (100cc H2O, 120cc Conc.HCl and 90g CuCl2) and modified Fry's reagent (100cc H2O, 80cc Conc.HCl, 20g CuCl2 and 20g FeCl3) (262) were used for macrographic etching of the specimens to show the distribution of plastic deformation at the shear lip region.

In the case of the aluminium alloy, the etchants used were 1% HF, Kellar's reagent (2cc HF (48%), 3cc conc. HCl, 5cc Conc. HNO3 and 190cc H2O) and Barker's reagent (263) (14% HBF4, stainless steel cathode, 5V, 30 seconds).

5.15 ELECTRON FRACTOGRAPHY

Immediately after fracturing, the fracture surfaces of the specimens were washed in acetone to remove any traces of oil present. Clear nail polish was coated on the fracture surfaces to prevent rust and atmospheric deterioration. Before viewing the specimen in the
scanning electron microscope (S.E.M.), the nail polish was thoroughly removed with acetone in an ultrasonic cleaner.

The fracture surfaces of the steel and aluminium specimens, tempered at different temperatures or tested in different crack plane orientations or at different testing temperatures, were examined using a JEOL J.S.M. U3 scanning electron microscope operating in the secondary electron image mode at 25 KV. For elemental analysis of the fracture surface, an EDAX 505 energy dispersive X-ray analyser attached to the electron microscope was employed.

5.16 SPECIMEN NOMENCLATURE

The identification code for the steel specimens consisted of a letter in between two numbers, for example,

<table>
<thead>
<tr>
<th>Specimen Number</th>
<th>Specimen Thickness</th>
<th>Tempering Temperature</th>
</tr>
</thead>
<tbody>
<tr>
<td>2</td>
<td>E</td>
<td>45</td>
</tr>
<tr>
<td>1 to 8 max.</td>
<td>A - 25mm</td>
<td>0 - As-quenched</td>
</tr>
<tr>
<td></td>
<td>B - 22mm</td>
<td>1 - 100°C</td>
</tr>
<tr>
<td></td>
<td>C - 19mm</td>
<td>2 - 200°C</td>
</tr>
<tr>
<td></td>
<td>D - 16mm</td>
<td>3 - 300°C</td>
</tr>
<tr>
<td></td>
<td>E - 13mm</td>
<td>4 - 400°C</td>
</tr>
<tr>
<td></td>
<td>F - 10mm</td>
<td>45 - 450°C</td>
</tr>
<tr>
<td></td>
<td>G - 7mm</td>
<td>5 - 500°C</td>
</tr>
<tr>
<td></td>
<td>H - 4mm</td>
<td>55 - 550°C</td>
</tr>
</tbody>
</table>

Hence, 2 E 45 represents the second 13mm thick specimen tempered at 450°C.

Unless stated, all steel specimens were of Composition O (Refer Appendix B) and tested in the C-L or R-L orientation. Specimens made from Composition N were designated N, for example, 3A2N.
The basic code for the aluminium specimens, tested in the T-L orientation, was similar to that of steel: a number followed by a letter. Thus,

```
<table>
<thead>
<tr>
<th>Specimen Number</th>
<th>Specimen Thickness</th>
<th>7075-T6 Aluminium</th>
</tr>
</thead>
<tbody>
<tr>
<td>3</td>
<td>F</td>
<td>Al</td>
</tr>
</tbody>
</table>
```

represents the third 10mm thick 7075-T6 aluminium specimen tested in the T-L orientation.

The basic crack plane orientation identification code \((70,118)\) was used to identify specimens tested with crack plane orientations different from those mentioned above, for example, 1A5N(R-C) or 3AA\(\bar{l}\)(S-T).
CHAPTER 6

EXPERIMENTAL RESULTS
6.1 INTRODUCTION

Results obtained in the present investigation are presented either in graphical or tabulated form in this chapter. The tensile properties of the materials are described first. The fracture toughness, in the elastic and yielding regimes, and its correlations with shear lip size and other parameters are then shown and briefly discussed. The fractographic results, both optical and scanning electron microscopy, are next presented together with an interpretation of the observed features. The final section contains the metallographic and microhardness data. It is noted that the metallographic section deals mainly with the spread of the plasticity as a result of the shear lip formation rather than the metallurgy of the materials which has been well studied in the literature.

The brief discussion of the results in each section is carried out independently. The combination of the above sections, to form a complete and comprehensive understanding of the investigation, is presented in the succeeding chapter.

6.2. TENSILE DATA

Using tensile test specimens described in section 5.10.1, the room temperature tensile properties of Comsteel En25 and 7075-T6 aluminium alloy were determined. The average results for Comsteel En25 at each heat treatment are presented graphically in Fig.6.1 and tabulated together with the aluminium alloy data in Appendix H. The slight variations in composition did not produce a marked difference in the tensile properties, but the general trend was that the steel with composition N, being cleaner than that with composition O, had a slightly higher ultimate tensile stress and 0.2% proof stress. The stresses for the steel with composition N increased slightly upon tempering at 100 - 200°C but then decreased steadily with the increase in tempering temperature.
For the steel of composition O, the proof and the ultimate tensile stresses decreased with increasing tempering temperature over the whole range.

Distinct yield plateaux were observed for all specimens tempered at 300°C and above. This can be seen in Fig.6.2 where the typical stress-strain curves at various tempering temperatures are presented for the steel of composition N.
Fig. 6.2: Effect of tempering temperature on stress-strain curves

The hardness results tabulated in Appendix H are the average of all specimens of the same heat treatment. As shown in Fig. 6.1, the hardness data correlate well with the tensile stress in that the variation of hardness with tempering temperature for both steel compositions follows closely that of tensile stress versus tempering temperature.

The percentage reduction in area of the tensile specimens generally increased with the increase in tempering temperature over
the entire tempering range as shown in Fig.6.3.

Fig.6.3: Effect of tempering temperature on area reduction and elongation

Within experimental error, the difference in steel compositions did not cause the reduction in area and the elongation to vary significantly, although the elongation of the steel with composition N appeared to be very slightly lower. The percentage elongation for both steel compositions showed a decrease in the 300°C to 400°C tempering region. Tempering above this region gradually increased the percentage elongation of the specimens.
The dependence of strain hardening exponent, \( n \), on tempering temperature depicted in Fig.6.4 (Tabulated in Appendix I) was roughly parallel to that of percentage elongation on tempering temperature shown in Fig.6.3 in that a decrease in \( n \) was observed in the same tempering region. At low tempering temperatures, the steel of composition O appeared to have a slightly higher \( n \) as well as percentage elongation. However, at high tempering temperatures this observation was not conclusive.

![Graph showing the effect of tempering temperature on strain hardening exponent](image)

**Fig.6.4**: Effect of tempering temperature on strain hardening exponent
Comparing the tensile properties of the 500°C tempered specimens tested in different orientations (Appendix H), it was observed that the specimen orientation produced a marked increase in the percentage reduction in area and the specimen elongation. The increase in area reduction from 30% in the R-L and R-C orientations to 59% in the L-R orientation was nearly twofold.

The tensile properties of the 7075 aluminium alloy in the T6 condition given in Appendix H and I are summarised in Table 6.1 below. It was noticed that specimen orientation had a slight effect on the material properties particularly the elongation, the reduction in area and the strain hardening exponent.

<table>
<thead>
<tr>
<th>Specimen orientation</th>
<th>T-L/T-S</th>
<th>L-T</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.2% proof stress, MPa</td>
<td>482</td>
<td>547</td>
</tr>
<tr>
<td>Ultimate tensile stress, MPa</td>
<td>547</td>
<td>597</td>
</tr>
<tr>
<td>Percentage reduction in area, %</td>
<td>5</td>
<td>10</td>
</tr>
<tr>
<td>Percentage elongation, %</td>
<td>6</td>
<td>16</td>
</tr>
<tr>
<td>Strain hardening exponent</td>
<td>0.053</td>
<td>0.13</td>
</tr>
<tr>
<td>Modulus of Elasticity, MPa</td>
<td>63300</td>
<td>63400</td>
</tr>
</tbody>
</table>

Table 6.1: Tensile Properties of 7075-T6 Aluminium Alloy

As the testing temperature decreased below ambient, the 0.2% proof stress and the ultimate tensile stress of both the steel and the aluminium alloy were found to increase as shown in Table 6.2. The steel tensile specimens were obtained from compact tension specimens that had been tempered at 600°C.
Table 6.2: Tensile Properties at Testing Temperatures Below Ambient

<table>
<thead>
<tr>
<th>Test Temperature, °C</th>
<th>Comsteel En25</th>
<th>7075-T6 Aluminium</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Ambient</td>
<td>0</td>
</tr>
<tr>
<td>$\sigma_{ys}$, MPa</td>
<td>896</td>
<td>963</td>
</tr>
<tr>
<td>$\sigma_{UTS}$, MPa</td>
<td>985</td>
<td>1042</td>
</tr>
<tr>
<td>% reduction in area</td>
<td>40</td>
<td>40</td>
</tr>
<tr>
<td>% elongation</td>
<td>18</td>
<td>14</td>
</tr>
</tbody>
</table>

Definite yield plateaux were observed in all the steel specimens while the phenomenon was absent in the aluminium alloy specimens. The reduction in area remained approximately constant as temperature was decreased until at -196°C, a drop in value was noted for both materials. No changes could be observed in the elongation of the materials although in the case of the steel specimens, a general decrease in percentage was evident when compared with the room temperature test result.

6.3 FRACTURE TOUGHNESS DATA

6.3.1 Effect of Tempering Temperature

The effect of tempering temperature on the plane strain fracture toughness, $K_{IC}$, of Comsteel En25 compact tension specimens of 25mm thickness tested in the R-L or C-L crack orientation is shown graphically in Fig.6.5 and tabulated in Appendix J. The average plane strain fracture toughness of the material in the as-quenched condition was 48 MPa/\text{m}. The toughness increased steadily to about 62 MPa/\text{m} at 200°C. Tempering at 300°C, however, produced a distinct drop in fracture toughness value to about 60 MPa/\text{m} indicating the occurrence of tempered martensite embrittlement at approximately 350°C. At 400°C and above, the fracture toughness of Comsteel En25 was found to increase with the increase in tempering temperature.

Fig.6.5 also shows the effect of slight variation in steel composition on the fracture toughness. It can be observed that the steel with composition N produced fracture toughness values that were
consistently higher than those of steel composition O. This is particularly noticeable at the tempered martensitic embrittlement and high tempering temperature regions.

Fig. 6.5: Effect of tempering temperature on fracture toughness of Comsteel En25

The change in shear lip size, $b_{SL}$, with tempering temperature was found to follow very closely the variation of fracture toughness with tempering temperature (see Fig. 6.6). However, the change in shear lip size in the 350°C region was not as marked as in the change in fracture toughness.
Fig.6.6: Effect of tempering temperature on shear lip size

At high tempering temperatures, shear lips were found to be less well defined and formed. This resulted in inaccuracy in the shear lip size measurement and consequently, scatter in result can be seen in Fig.6.6.

The variation in steel composition did not produce noticeable change in shear lip size.
6.3.2 Effect of Specimen Thickness

The influence of specimen thickness, \( B \), on the fracture toughness, \( K_c \), of Comsteel En25 (Steel composition 0) at various tempering temperatures is shown in Fig. 6.7 and tabulated in Appendix J.

![Graph showing the effect of specimen thickness on fracture toughness](image)

Fig. 6.7: Effect of specimen size on fracture toughness of Comsteel En25

For all tempering temperatures, \( K_c \) decreased towards the limiting \( K_{IC} \) values as the specimen thickness was increased. At tempering temperatures of 400\(^\circ\)C and below, the value of \( K_c \) for the 25mm thick specimens was approximately equal to the average \( K_{IC} \).
value at that tempering temperature as can be seen in Table 6.3. However, above 400°C, the difference between the two toughness parameters increased with increasing tempering temperature. The difference between the maximum $K_c$ value and the constant $K_{IC}$ value was also observed to increase with temperature.

<table>
<thead>
<tr>
<th>Tempering Temperature (°C)</th>
<th>$K_{IC}$* (MPa√m)</th>
<th>$K_c$** (MPa√m)</th>
</tr>
</thead>
<tbody>
<tr>
<td>As Quenched</td>
<td>55.28</td>
<td>52.73</td>
</tr>
<tr>
<td>100</td>
<td>54.94</td>
<td>55.80</td>
</tr>
<tr>
<td>200</td>
<td>64.66</td>
<td>66.65</td>
</tr>
<tr>
<td>300</td>
<td>62.23</td>
<td>60.83</td>
</tr>
<tr>
<td>400</td>
<td>64.68</td>
<td>64.67</td>
</tr>
<tr>
<td>450</td>
<td>78.21</td>
<td>85.57</td>
</tr>
<tr>
<td>500</td>
<td>84.01</td>
<td>90.43</td>
</tr>
<tr>
<td>550</td>
<td>91.90</td>
<td>104.89</td>
</tr>
<tr>
<td>600</td>
<td>92.05</td>
<td>109.51</td>
</tr>
</tbody>
</table>

Table 6.3: Comparison of $K_{IC}$ with $K_c$ values at different tempering temperatures.

* Average over all ASTM valid tests
** Average over the 25mm thickness specimens

The latter difference varied from about 1.5 times in the as-quenched state to about 2 times in the 600°C temperature.

At high tempering temperatures, most specimens that failed to meet the ASTM standard\(^{(118)}\) had thicknesses too small to meet the $2.5 \left(\frac{K_c}{\sigma_y}\right)^2$ requirement.

The load/displacement records at a given tempering temperature showed a marked change in shape as the specimen thickness decreased; the amount of plastic behaviour appeared to increase.
Hence the ratio of $\frac{P_{\text{max}}}{Q}$ increased with the decrease in specimen thickness. A typical set of load-displacement records is illustrated in Fig.6.8 where the $\frac{P_{\text{max}}}{Q}$ ratio increased from an average value of 1.04 for 25mm thickness specimens to 1.66 for 4mm thickness specimens.

Shear lip size measurements showed that $B_{SL}$ was within experimental error, constant along the crack length of the specimens. For a given tempering temperature, the size of the shear lip was found to be independent of the thickness of the test specimen as shown in Fig.6.9 and tabulated in Appendix J.
Shear lips were well defined and well formed in specimens tempered at a temperature below 400°C; but as the tempering temperature was increased, the distinctness of the shear lip formation was found to diminish (see Fig.6.10).

For the 7075-T6 aluminium alloy, the variation of $K_c$ with specimen thickness followed a similar trend to the Comsteel En25. As shown in Fig.6.11, the $K_c$ values for specimen thicknesses larger than 19mm remained constant at a value of approximately 29.6 MPa/m (see Appendix K for tabulated data).
Fig. 6.10 Shear lip formation on Comsteel En25 specimens.

Fig. 6.12 Shear lip formation on 7075-T6 aluminium specimen.
Fig. 6.11: Effect of specimen size on fracture toughness of 7075-T6 aluminium alloy

The average $K_{IC}$ value was 27.7 MPa m. Specimens of less than 10 mm thickness were found to give invalid ASTM tests either because of the specimen thickness or the $P_{max}/P$ requirements. Although the specimens were sufficiently thick to attain a state of plane strain, they did not appear to be thin enough to give the true plane stress fracture toughness, where $K_c$ has a maximum value.

As with Comsteel En25, the shear lip size in the aluminium alloy was found to be invariant with the thickness of the test specimen and the length of the fracture path. However, the formation of the shear lips along the free surfaces of the specimen was observed to be discontinuous, as shown in Fig. 6.12.
6.3.3 Effect of Specimen Orientation

The fracture toughness data for specimens of different crack plane orientation, tabulated in Appendix J for Comsteel En25 and Appendix K for the 7075-T6 aluminium alloy, are summarised in Table 6.4. The tempering temperature of the steel specimens was 500°C.

<table>
<thead>
<tr>
<th>K_{IC}/B_{SL}</th>
<th>Orientation</th>
<th>R-L</th>
<th>T-L</th>
<th>R-C</th>
<th>S-T</th>
<th>L-R</th>
<th>L-S</th>
</tr>
</thead>
<tbody>
<tr>
<td>Comsteel En25</td>
<td>(MPa√m)</td>
<td>82.73</td>
<td>87.35</td>
<td>150.36*</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Shear lip size</td>
<td>(mm)</td>
<td>1.6</td>
<td>1.9</td>
<td>4.5</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>7075-T6 aluminium alloy</td>
<td>(MPa√m)</td>
<td>27.67</td>
<td>29.60</td>
<td>29.67</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Shear lip size</td>
<td>(mm)</td>
<td>1.3</td>
<td>1.1</td>
<td>1.3</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Table 6.4: Average K_{IC} (MPa√m) and shear lip size (mm) for specimens of different crack plane orientations.

* denotes K_{IC} value only.

The difference in crack plane orientation did not produce a significant effect on the fracture toughness of the aluminium alloy. Nevertheless, toughness in the S-T and L-S orientations was generally slightly higher than that in the T-L orientation. For Comsteel En25, the fracture toughness was significantly affected by the crack plane orientation. While the toughness in the R-C orientation was noticeably higher than in the R-L orientation, a nearly two-fold increase was obtained in the L-R orientation. As a result of the high toughness value in the L-R orientation, the data were invalidated by the ASTM standard specimen thickness requirement of 2.5(K_{IC}/σ_{YS})^2 to be less than the actual specimen thickness, B.

The fracture surfaces for the specimens of different crack plane orientation are shown in the macrographs in Fig.6.13.
Fig. 6.13 Effect of crack plane orientation on fracture surface.
The central portion of the shear lips along the crack length of the R-L and R-C orientated specimens was of approximately constant size; but in the L-R orientation, the specimen width, \( W \), did not appear to be large enough for the shear lips to attain a constant size. Table 6.4 shows that the average size of the shear lip in the L-R orientation is about three times that in the R-L specimens.

The shear lip size for the aluminium alloy was almost constant for all orientations of the crack plane (see Fig. 6.13 and Table 6.4).
6.3.4 Effect of Testing Temperature

The effect of testing temperature, for temperatures below ambient, on Comsteel En25 (composition N) tempered at 600°C and 7075-T6 aluminium alloy in the as-received condition, is shown in Table 6.5 below.

<table>
<thead>
<tr>
<th>Specimen designation</th>
<th>Comsteel En25</th>
<th>7075-T6Al</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>12A6N</td>
<td>13A6N</td>
</tr>
<tr>
<td>Testing Temperature, °C</td>
<td>0</td>
<td>-78</td>
</tr>
<tr>
<td>Specimen thickness, B, mm</td>
<td>25.03</td>
<td>25.05</td>
</tr>
<tr>
<td>Specimen width, W, mm</td>
<td>49.16</td>
<td>49.77</td>
</tr>
<tr>
<td>Crack length, a, mm</td>
<td>24.54</td>
<td>26.39</td>
</tr>
<tr>
<td>$P' Q'$</td>
<td>60.7</td>
<td>54.3</td>
</tr>
<tr>
<td>$K_Q$</td>
<td>104.67</td>
<td>102.41</td>
</tr>
<tr>
<td>$P_{max}/P'Q'$</td>
<td>1.05</td>
<td>1.06</td>
</tr>
<tr>
<td>$2.5(K_QO)^2$</td>
<td>29.53</td>
<td>25.31</td>
</tr>
<tr>
<td>ASTM test validity</td>
<td>NO</td>
<td>NO</td>
</tr>
<tr>
<td>Shear lip size, mm</td>
<td>1.9</td>
<td>1.8</td>
</tr>
</tbody>
</table>

Table 6.5: Fracture toughness data at test temperatures below ambient.

Comparing with the room temperature fracture toughness, listed in Appendices J and K, for 600°C tempered steel, an average $K_Q$ value of 105.24 MPa/√m and for 7075-T6 aluminium alloy a $K_{IC}$ value of 27.67 MPa/√m, it can be seen that decreasing the test temperature below ambient does not cause any significant change in the fracture toughness values for both materials. However, due to the increase in proof stress (see section 6.2) with decreasing testing temperature, the steel specimen tested at -196°C was found to give a valid test according to the ASTM Standard E399-74(118). The $K_{IC}$ value of the aluminium alloy at -196°C was slightly lower than expected. The load/displacement records were of the type shown in Fig.2.6(c) where the maximum load, $P_{max}$, is identical to the 5% secant load, $P'Q'$. 
6.4 ERROR ANALYSIS

6.4.1 Tensile Data

The estimated error in the tensile test data was expressed as a ratio of the standard deviation to the mean. This ratio, known as the coefficient of variation, gives an indication of the relation between the scatter and the mean, and when multiplied by 100 can be regarded as the percentage error. The values of the coefficient of variation for tensile specimens tempered at various temperatures as well as those for the 7075-76 aluminium alloy are listed in Appendix H where $S$ is the standard deviation and $\bar{x}$ is the mean of the specimens tested.

6.4.2 Fracture Toughness Data

Systematic errors occurring in the $K_{IC}$ determination arose mainly from the inaccuracies in measuring the various parameters that were required for the fracture toughness calculation. From the fracture toughness equation for the compact tension specimen,

$$K_{IC} = \frac{P\sqrt{a}}{B^{1/2}W^{1/2}} f\left(\frac{a}{W}\right)$$

where measurements were required to be made of the 5% offset secant load, $P$, specimen width, $W$, specimen thickness, $B$, and crack length, $a$, the estimated errors in these parameters were:

\[
\begin{align*}
B &= \pm 0.01 \text{ mm} \\
W &= \pm 0.05 \text{ mm} \\
a &= \pm 0.02 \text{ mm}
\end{align*}
\]

and $P = 2\%$, equal contribution from the load determination and the secant construction.

Therefore, by using the error propagation formula (see Appendix L), it can be shown that the error in the $K_{IC}$ calculation is

$$\left(\frac{\sigma_{K_{IC}}}{K_{IC}}\right)^2 = \left(\frac{\sigma_P}{P}\right)^2 + \left(\frac{\sigma_B}{B}\right)^2 + 0.25 \left(\frac{\sigma_W}{W}\right)^2 + 2.22 \left(\frac{\sigma_a}{a}\right)^2$$

where the various terms on the right-hand-side of the expression can be obtained from the estimated errors listed above. These were,
\[ \frac{\sigma_p}{P} = 0.02 \]
\[ \frac{\sigma_B}{B} = 0.001 \text{ for } B = 25\text{mm} \]
\[ = 0.003 \text{ for } B = 4\text{mm} \]
\[ \frac{\sigma_W}{W} = 0.001 \text{ for } W = 50\text{mm} \]
\[ \text{and } \frac{\sigma}{\frac{a}{W}} = 0.001 \text{ for } a = 25\text{mm} \]

Consequently, the error for the \( K_{IC} \) value was approximately 2% for all specimens thicknesses.

6.5 Correlation of Fracture Toughness with Shear Lip Size

The fracture toughness and shear lip size data for Comsteel En25 and 7075-T6 aluminium alloy specimens are tabulated in Appendices J and K respectively. From Fig.6.9 it can be seen that the shear lip size at a given tempering temperature is independent of specimen thickness. Table 6.6 gives a summary of the average \( K_{IC} \) and shear lip size values for all the

<table>
<thead>
<tr>
<th>Heat Treatment</th>
<th>Steel Composition O</th>
<th>Steel Composition N</th>
<th>Aluminium</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>( K_{IC} ) (MPa(\sqrt{m}))</td>
<td>( B_{SL} ) (mm)</td>
<td>( K_{IC} ) (MPa(\sqrt{m}))</td>
</tr>
<tr>
<td>As-Quenched</td>
<td>55.28</td>
<td>0.47</td>
<td>47.20</td>
</tr>
<tr>
<td>100°C Temper</td>
<td>54.94</td>
<td>0.49</td>
<td>49.93</td>
</tr>
<tr>
<td>200°C Temper</td>
<td>64.66</td>
<td>0.71</td>
<td>62.51</td>
</tr>
<tr>
<td>300°C Temper</td>
<td>62.23</td>
<td>0.90</td>
<td>61.75</td>
</tr>
<tr>
<td>400°C Temper</td>
<td>64.68</td>
<td>1.11</td>
<td>68.99</td>
</tr>
<tr>
<td>450°C Temper</td>
<td>78.21</td>
<td>1.49</td>
<td>-</td>
</tr>
<tr>
<td>500°C Temper</td>
<td>84.01</td>
<td>1.72</td>
<td>82.73</td>
</tr>
<tr>
<td>550°C Temper</td>
<td>91.90</td>
<td>1.82</td>
<td>-</td>
</tr>
<tr>
<td>600°C Temper</td>
<td>92.05</td>
<td>2.26</td>
<td>105.24*</td>
</tr>
<tr>
<td>650°C Temper</td>
<td>-</td>
<td>-</td>
<td>110.70*</td>
</tr>
<tr>
<td>T6</td>
<td>-</td>
<td>-</td>
<td>27.67</td>
</tr>
</tbody>
</table>

Table 6.6 Average \( K_{IC} \) and \( B_{SL} \) values for Comsteel En25 and 7075-T6 aluminium alloy.
materials used in this investigation. Data that were invalid ASTM tests are marked with asterisks.

The correlation between plane strain fracture toughness, $K_{IC'}$ and size of the shear lip, $B_{SL}$, is presented graphically in Fig. 6.14 where $B_{SL}$ was plotted against the relative toughness parameter, $(K_{IC'}/\sigma_{YS})$. The theoretical prediction of the size of the crack tip plastic zone formulated by Rice (158), Liu (112) and Irwin (19), are also shown in the figure.

![Diagram showing the relationship between fracture toughness and shear lip size on linear plot.](image-url)

**Fig. 6.14:** Relationship between fracture toughness and shear lip size on linear plot.

Considering the data for the steel specimens it can be seen that at low tempering temperatures, the relationship between $B_{SL}$ and $(K_{IC'/\sigma_{YS}})$
can be described using an expression similar to that proposed by Rice (158) for the plastic zone size; namely,

\[ B_{SL} = B \left( \frac{K_{IC}}{\sigma_{YS}} \right)^m \] .............................. 6.1

where \( B \) and \( m \) are constants.

At tempering temperatures higher than 450°C, Fig.6.14 shows that the relationship between \( B_{SL} \) and \( (K_{IC}/\sigma_{YS}) \) deviates from equation 6.1. The deviation increases with the increase in tempering temperature.

The constants \( B \) and \( m \) in equation 6.1 were determined by plotting \( B_{SL} \) and \( (K_{IC}/\sigma_{YS}) \) on log-log scales. This is shown in Fig.6.15. At the low tempering temperatures, below 450°C, the data follows approximately a linear relationship while at high tempering temperatures, deviation from linearity is evident. Regression analysis of the data below 450°C resulted in a value of \( B = 0.41 \) and \( m = 2.02 \). Thus equation 6.1 can be rewritten as

\[ B_{SL} = 0.41 \left( \frac{K_{IC}}{\sigma_{YS}} \right)^{2.02} \] .............................. 6.2

The significance of equation 6.2 will be discussed fully in the next chapter.

Figures 6.14 and 6.15 show that the variation in steel composition did not affect the relationship between \( B_{SL} \) and \( (K_{IC}/\sigma_{YS}) \). It can also be observed that the data for the 7075-T6 aluminium alloy fell very close to the linear relationship expressed in equation 6.2, showing that the relationship is applicable to the aluminium alloy.

6.6 CORRELATION OF \( K_{IC} \) WITH OTHER PARAMETERS

6.6.1 Tensile Data

A relationship between smooth tensile properties and \( K_{IC} \) of the type reported by Jones and Brown (265), that is, \( K_{IC} \propto \sigma_{YS}^{-3} \), was not found for Comsteel En25. Also there did not appear to be a correlation between either \( \sigma_{UTS} \) or \( \sigma_{YS} \) and \( K_{IC} \). Nevertheless, a plot of relative toughness, \( K_{IC}/\sigma_{YS} \), against \( \sigma_{YS} \) provided a linear relationship over a limited range of tempering temperatures (see Fig.6.16). A distinct change in gradient occurs at a tempering temperature of about 400°C.
Fig. 6.15: Relationship between fracture toughness and shear lip size on log-log plot.

Fig. 6.16: Relationship between fracture toughness and tensile data.
6.6.2 Stretch Zone

The stretch zone is defined as the region of shear deformation between the fatigue crack and the over-load fracture region of a fracture toughness testpiece. The width of this region, $t_{SZ}$, was determined on the scanning electron microscope using a magnification of between 600 and 2,000 times. No distinct stretch zone was observable for tempering temperatures below 300°C. Well formed stretch zones were however observed in specimens tempered above this temperature. The results are tabulated in Table 6.7.

<table>
<thead>
<tr>
<th>Material</th>
<th>Tempering Temperature (°C)</th>
<th>Specimen</th>
<th>Stretch Zone Width (µm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Comsteel En25</td>
<td>650</td>
<td>4A65N</td>
<td>29.2</td>
</tr>
<tr>
<td></td>
<td>600</td>
<td>1A6</td>
<td>17.9</td>
</tr>
<tr>
<td></td>
<td>550</td>
<td>1A55</td>
<td>16.0</td>
</tr>
<tr>
<td></td>
<td>500</td>
<td>1A5N</td>
<td>12.3</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1A5</td>
<td>12.2</td>
</tr>
<tr>
<td></td>
<td></td>
<td>2A5</td>
<td>12.1</td>
</tr>
<tr>
<td></td>
<td></td>
<td>3A5</td>
<td>12.1</td>
</tr>
<tr>
<td></td>
<td>450</td>
<td>1A45</td>
<td>9.0</td>
</tr>
<tr>
<td></td>
<td>400</td>
<td>1A4</td>
<td>5.8</td>
</tr>
<tr>
<td></td>
<td></td>
<td>2A4</td>
<td>6.6</td>
</tr>
<tr>
<td></td>
<td></td>
<td>3A4</td>
<td>6.5</td>
</tr>
<tr>
<td>725-T6 Aluminium Alloy</td>
<td></td>
<td>Not Observed</td>
<td></td>
</tr>
</tbody>
</table>

Table 6.7: Stretched zone width for Comsteel En25 and 7075-T6 aluminium alloy.

The relationship between the stretch zone width and relative toughness, $(K/\sigma_{YS})$, plotted on log-log scales is shown in Fig.6.17. The straight line in the figure, given by the equation
Relative toughness, $\frac{K}{\sigma_{ys}}$, $\sqrt[1.6]{m}$.

Fig. 6.17: Relationship between fracture toughness and stretch zone width

$\tau_{sz} = 0.5 \left( \frac{K}{\sigma_{ys}} \right)^{1.6}$ mm

was determined by Bates and Clark (266) for a number of aluminium and steel alloys. All the Comsteel En25 data were seen to lie above the Bates and Clark line indicating that the stretch zone widths were larger than expected.
Fig. 6.18 Fibrous crack extension marking by (a) heat tinting, (b) fatigue.
6.7 FRACTURE TOUGHNESS FROM YIELDING FRACTURE MECHANICS

6.7.1 Crack Opening Displacement at Crack Initiation, $\delta_i$

The results from the six 25mm thickness compact tension specimens tempered at 600°C and unloaded at different amount of fibrous crack growth are tabulated in Appendix M. Where possible, the values of $K_Q$ are also included in the Appendix.

Fig.6.18 shows the typical fracture surfaces of the heat tinted and fatigue marked specimens. It can be observed that the regions of fibrous crack growth are easily discernible.

The variation of the crack opening displacement, $\delta$, at the point of unloading with the amount of fibrous crack growth, $\Delta a$, as shown in Fig.6.19, follows approximately a linear relationship. $\delta$ was calculated using the theoretical expression (equation 4.8 of section 4.2.2) from Reference (210).

![Graph showing linear relationship between crack opening displacement and fibrous crack length.](image)

**Fig.6.19:** Determination of C.O.D. at crack initiation
Extrapolation of the line in Fig 6.19 to zero fibrous crack growth gave a value of \( \delta_1 \) of 0.039 mm. The fracture toughness calculated from equation 4.5 taking \( \lambda = 1 \) was found to be 84.64 MPa\( \sqrt{m} \). This was much lower than that determined from the ASTM standard analysis (118) using a 2% increment of crack extension where an average \( K_Q \) value of 105.24 MPa\( \sqrt{m} \) was obtained.

6.7.2 J-Contour Integral

The relation of \( J \) at the point of unloading versus the length of fibrous crack growth, \( \Delta a \), is graphically presented in Fig. 6.20 and tabulated in Appendix M. An approximate linear relationship was observed. The value of \( J_{IC} \) from the intersection of the \( J-\Delta a \) line and the \( J = 2\Delta a\sigma_{\text{flow}} \) line, where \( \sigma_{\text{flow}} = \frac{1}{2}(\sigma_{YS} + \sigma_{UTS}) \) and \( \sigma_{YS} = 896 \) MPa, \( \sigma_{UTS} = 985 \) MPa, was found to be 28.5 Nmm/mm\(^2\).

![Graph showing J-Contour Integral](image-url)
The plane strain fracture toughness of Comsteel En25 tempered at 600°C using equation 4.12,

\[ K_{IC} = \sqrt{J_{IC}} \frac{E}{(1-\nu^2)} \]

was therefore 80.13MPa/m. This compares well with the value of 84.64 MPa/m obtained from \( \delta_i \) analysis.

The minimum specimen size for valid \( J_{IC} \) using

\[ B, a, W-a > \alpha J_{IC} / \sigma_{flow} \] ................................. 6.4

where \( \alpha = 50 \) was found to be 1.5mm. This is smaller than the dimensions of the 25mm thick specimens.

6.8 FRACTURE TOUGHNESS FROM SIDE-GROOVED COMPACT TENSION SPECIMEN

Of the three side-grooved compact tension specimens, only one was successfully tested. The other two were found to possess quench cracks that were too severe to render any usefulness. Results of the successful test are summarised in Table 6.8 while the load/displacement record is shown in Fig. 6.21 where for comparison purposes, the typical load/displacement record of a non side-grooved specimen tempered at the same temperature is included.

<table>
<thead>
<tr>
<th>Width</th>
<th>Ungrooved thickness</th>
<th>Crack Length</th>
<th>Depth of Groove</th>
<th>( P_Q )</th>
<th>( P_{max} )</th>
<th>( K_{COD} )</th>
<th>( J ) at max. load</th>
<th>( K_J )</th>
</tr>
</thead>
<tbody>
<tr>
<td>W</td>
<td>B</td>
<td>a</td>
<td>t</td>
<td>(mm)</td>
<td>(KN)</td>
<td>(KN)</td>
<td>(MPa/m)</td>
<td>(MPa/m)</td>
</tr>
<tr>
<td>49.33</td>
<td>25.04</td>
<td>24.97</td>
<td>2.96</td>
<td>40.0</td>
<td>42.8</td>
<td>81.87</td>
<td>28.80</td>
<td>80.55</td>
</tr>
</tbody>
</table>

Table 6.8 Results of side-grooved specimen (steel composition N) tempered at 600°C.
No shear lips on the fracture surface were detectable. This was expected since the average shear lip size at 600°C tempering temperature on non-side-grooved specimens was only 2.65mm, whereas the depth of the side-groove was 2.96mm. Effect of side ligaments was therefore eliminated and the COD at maximum load occurred at approximately the same position as COD at initiation. This can be seen to be so by comparing the fracture toughness values in Table 6.8 with the results in section 6.7. The fracture toughness values from crack opening displacement and the corresponding J-integral at maximum load were 81.87 and 80.55 MPa/m respectively. These compared well with those calculated from $\delta_i$ and $J_{IC}$. 
The results in sections 6.7 and 6.8 show that the fracture toughness at 600°C temper calculated using the ASTM standard analysis is different from that obtained using the \( J_{IC} \) approaches and the side-grooved specimen. The discrepancy may be due to the invalidity of the ASTM \( K_{IC} \) values as a result of excessive yielding at high tempering temperatures. This will be fully discussed in Chapter 7. It is, however, necessary to state here that \( K_{IC} \) will henceforth be regarded as the fracture toughness obtained merely from the ASTM Standard 5% offset secant analysis and that it may not necessarily be the true plane strain fracture toughness of the material. Where the true value is meant, it will be specified.

### 6.9 Comparison of \( K_O \) with \( K_{COD} \), \( K_J \) and \( K_{EE} \)

The fracture toughness values for the 25mm thick specimens, determined at the point of maximum load on the load/displacement fracture test record using the crack opening displacement approach, \( K_{COD} \), the \( J \) integral approach, \( K_J \), and the equivalent energy approach, \( K_{EE} \), are shown as a function of tempering temperature in Fig.6.22 and tabulated in Table 6.9. The toughness determined using the theoretical approach of Wells (210) is denoted by \( K^T_{COD} \), and the experimental approach using a constant rotational factor of \( r = 0.4 \) suggested by Ingham et al. (214) and Elliott et al. (215), is denoted by \( K^E_{COD} \). In converting the crack opening displacements to the fracture toughness values, \( \lambda \) was taken as unity. The justification of this assumption will be discussed in the next chapter. The maximum load \( K_C \) values are also included in Table 6.9 for comparison purposes.
<table>
<thead>
<tr>
<th>Specimen</th>
<th>$K_{IC}$ (MPa√m)</th>
<th>$K_C$ (MPa√m)</th>
<th>$K_{COD}^T$ (MPa√m)</th>
<th>$K_{COD}^E$ (MPa√m)</th>
<th>$K_J$ (MPa√m)</th>
<th>$K_{EE}$ (MPa√m)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1A6</td>
<td>92.05</td>
<td>109.51</td>
<td>94.50</td>
<td>165.65</td>
<td>86.92</td>
<td>111.93</td>
</tr>
<tr>
<td>1A55</td>
<td>94.98</td>
<td>104.89</td>
<td>96.95</td>
<td>171.73</td>
<td>83.72</td>
<td>110.59</td>
</tr>
<tr>
<td>1A5</td>
<td>75.45</td>
<td>83.62</td>
<td>90.39</td>
<td>158.79</td>
<td>67.95</td>
<td>88.33</td>
</tr>
<tr>
<td>2A5</td>
<td>81.22</td>
<td>94.46</td>
<td>97.85</td>
<td>166.60</td>
<td>73.43</td>
<td>97.88</td>
</tr>
<tr>
<td>3A5</td>
<td>86.23</td>
<td>93.22</td>
<td>92.40</td>
<td>168.16</td>
<td>72.19</td>
<td>95.08</td>
</tr>
<tr>
<td>1A45</td>
<td>78.98</td>
<td>85.57</td>
<td>94.19</td>
<td>165.92</td>
<td>68.47</td>
<td>90.11</td>
</tr>
<tr>
<td>1A4</td>
<td>59.29</td>
<td>63.32</td>
<td>84.98</td>
<td>150.03</td>
<td>49.25</td>
<td>63.43</td>
</tr>
<tr>
<td>2A4</td>
<td>63.22</td>
<td>67.15</td>
<td>84.42</td>
<td>154.37</td>
<td>51.03</td>
<td>68.78</td>
</tr>
<tr>
<td>3A4</td>
<td>60.52</td>
<td>64.54</td>
<td>82.92</td>
<td>147.42</td>
<td>49.22</td>
<td>65.98</td>
</tr>
<tr>
<td>1A3</td>
<td>56.14</td>
<td>59.64</td>
<td>89.34</td>
<td>150.74</td>
<td>49.91</td>
<td>65.05</td>
</tr>
<tr>
<td>2A3</td>
<td>57.76</td>
<td>60.23</td>
<td>87.62</td>
<td>152.67</td>
<td>48.49</td>
<td>61.27</td>
</tr>
<tr>
<td>3A3</td>
<td>59.12</td>
<td>62.62</td>
<td>83.90</td>
<td>152.53</td>
<td>48.45</td>
<td>64.13</td>
</tr>
<tr>
<td>1A2</td>
<td>60.39</td>
<td>64.59</td>
<td>93.34</td>
<td>163.85</td>
<td>50.48</td>
<td>68.24</td>
</tr>
<tr>
<td>2A2</td>
<td>60.83</td>
<td>68.71</td>
<td>95.84</td>
<td>162.55</td>
<td>53.34</td>
<td>65.91</td>
</tr>
<tr>
<td>1A1</td>
<td>48.71</td>
<td>55.80</td>
<td>86.31</td>
<td>147.20</td>
<td>44.25</td>
<td>54.36</td>
</tr>
<tr>
<td>1A0</td>
<td>48.50</td>
<td>52.73</td>
<td>82.18</td>
<td>146.40</td>
<td>40.93</td>
<td>53.31</td>
</tr>
</tbody>
</table>

Table 6.9: Fracture toughness from yielding fracture mechanics analyses.
Fig. 6.22 shows that $K_{COD}^E$ grossly overestimates the values of $K_c$ and $K_{IC}$ at all tempering temperatures by a factor of 2 to 3 times. To a lesser extent, $K_{COD}^T$ also overestimates $K_c$ at tempering temperatures below 500°C but at 550 and 600°C, $K_{COD}^T$ is less than $K_c$. Although $K_J$ is consistently lower than $K_c$ and $K_{IC}$ over the whole range of tempering temperature, it predicts well the shape of the curve of fracture toughness versus tempering temperature. $K_{EE}$ correlates favourably with $K_c$ as can be seen in Table 6.9. The discrepancy was found to be less than 10%. However, instead of being the lower bound, the values of $K_{EE}$ are consistently higher than $K_{IC}$ values.

Fig.6.22: Comparison of fracture toughness obtained from yielding fracture mechanics approaches.
6.10 FRACTOGRAPHY

6.10.1 Optical Microscopic Observations

Representative macro fracture features of Comsteel En25 austenitized at 850°C and tempered at various temperatures can be seen in Figure 6.23 and 6.24 where the specimen orientation was either C-L or R-L.

The as-quenched specimen (Fig.6.23(a)) had a rather flat and featureless fracture appearance in both the fatigue cracked region and the over-load region. The fatigue crack front was well defined. Similar fracture features were observed at a tempering temperature of 100°C. However, at 200°C a significant change occurred in that a quite uneven appearance was noticed. Tempering at 300°C produced a flat and rather featureless fracture surface similar to that in the as-quenched and 100°C tempered specimen (see Fig.6.23(b) and 6.23(d)). As the tempering temperature was further increased, the amount of unevenness gradually increased such that at 600 and 650°C, as can be seen in Figs.6.24(d) and 6.24(e), the fracture appearance became extremely rugged. The large contrast between the uneven over-load region and the relatively smooth fatigued region made the fatigue crack front clearly distinguishable. At high tempering temperatures, the fatigue crack front appeared to bow outward so that the crack length at the centre of the specimen was greater than at the free surfaces. The amount of contraction in the specimen thickness direction also increased with increasing tempering temperature while the formation of shear lips along the free surfaces of the testpieces became less and less well defined.

A common macro fracture feature that appeared in all the specimens shown in Figs. 6.23(a) to 6.24(e) was the lines parallel to the direction of crack propagation and coincident with the rolling direction.

The effect of specimen orientation on the macro fracture appearance is shown in Fig.6.13. The testpieces have been austenitized at 850°C and tempered at 500°C. The major differences between each orientation are the size and shape of the shear lips and the direction of the lines on the fracture surface which for the R-C and L-R orientations are perpendicular to the direction of crack propagation and for L-R orientation also perpendicular to the rolling direction. While the fracture features in the
Fig. 6.23 Macro fracture features of Comsteel En25
(as-quenched to 400°C temper)
Fig. 6.24 Macro fracture features of Comsteel En25
(450°C temper to 650°C temper).
R-L and R-C orientations appeared rugged, the L-R orientation however appeared comparatively less uneven.

The effect of anisotropy was not discernable in the aluminium specimens as shown in Fig.6.13. At the initial high stress intensity region, the fatigue crack surface generally appeared to be dark and spongy while a shiny surface was observed when the maximum stress was reduced to less than 60% of the final fracture toughness $K_f$ value. The fracture appearance of the overload region generally had rather coarse features.

### 6.10.2 Scanning Electron Microscopic Observations

The fracture surfaces of the Comsteel En25 specimens tempered at various temperatures and the 7075 aluminium alloy in the T-6 condition were observed in the JEOLJ.S.M.U3 scanning electron microscope. The area of observation was confined mainly to the central 10mm of test-piece width and up to about 15mm behind the fatigue crack tip. In all the micrographs discussed below the direction of crack propagation is from the bottom of the page towards the top. The discussion is divided into three sub-sections as follows:

For Comsteel En25,
(a) Effect of tempering temperature (R-L orientation),
(b) Effect of crack plane orientation on specimens tempered at 500°C.

For 7075-T6 aluminium alloy,
(c) Effect of crack plane orientation.

A general introduction to the various modes of fracture, namely, dimple rupture, cleavage, fatigue and decohesive rupture, is included in Appendix N.

(a) **Effect of Tempering Temperature (R-L orientation)**

The typical fracture morphology of the as-quenched specimens is shown in Figs.6.25,26,27(a) and 27(b). At low magnification, the overload fracture surface was rather flat with few MnS stringers, as shown in stereo micrographs in Figs.6.25(a) and 6.25(b).
a. Overload fracture.
b. Overload fracture.
c. Fatigue.
d. Fatigue.

Fig. 6.25 Fracture morphology of as-quenched specimens.
a. Crack tip.
b. Crack tip.
c. Overload fracture.
d. Overload fracture.

Fig. 6.26 Fracture morphology of as-quenched specimens.
Fig. 6.27 Fracture morphology of as-quenched (a,b) and 100°C tempered (c,d) specimens.
a. Crack tip.

b. Overload fracture.

c. Overload fracture.

d. Overload fracture.

Fig. 6.28 Fracture morphology of 100°C tempered specimens.
Fig. 6.29 Fracture morphology of 100\(^\circ\)C (a,b) and 200\(^\circ\)C (c,d) tempered specimens.
a. Overload fracture.  

b. Overload fracture.  

c. Dimples.  

d. Dimples.  

Fig. 6.30 Fracture morphology of 200°C tempered specimens.
Fig. 6.31 Fracture morphology of 300°C tempered specimens.
Fig. 6.32 Fracture morphology of 300°C (a,b) and 400°C (c,d) tempered specimens.
a. Fatigue.

b. Crack tip.

c. Cleavage.

d. Dimples.

Fig. 6.33 Fracture morphology of 400°C tempered specimens.
Fig. 6.34 Fracture morphology of 400°C (a,b) and 450°C (c,d) tempered specimens.
Fig. 6.35 Fracture morphology of 450°C (a, b) and 500°C (c, d) tempered specimens.
a. Fatigue.

b. Overload fracture.

c. Crack tip.

d. Crack tip.

Fig. 6.36 Fracture morphology of 500°C tempered specimens.
Fig. 6.37 Fracture morphology of 500°C (a,b) and 550°C (c,d) tempered specimens.
Fig. 6.38 Fracture morphology of 550°C tempered specimens.
a. Overload fracture.
b. Overload fracture.
c. Fatigue.
d. Fatigue.

Fig. 6.39 Fracture morphology of 600°C tempered specimens.
a. Dimples.

b. Dimples.

c. Shear lip.

d. Shear lip.

**Fig. 6.40** Fracture morphology of 600 C tempered specimens.
a. Overload fracture.
b. Crack tip.
c. Crack tip.
d. Dimples.

Fig. 6.41 Fracture morphology of 650°C tempered specimens.
a. Overload fracture.

b. Overload fracture.

c. Crack tip.

d. Crack tip.

Fig. 6.42 Fracture morphology in R-C orientation
(500°C temper).
Fig. 6.43 Fracture morphology in R-C orientation (500°C temper).
Fig. 6.44 Fracture morphology in L-R orientation
(500°C temper).
Fig. 6.45 Fracture morphology in L-R orientation
(500°C temper).
Fig. 6.46 Fracture morphology in L-S orientation
(7075-T6 aluminium).
Fig. 6.47 Fracture morphology in L-S orientation
(7075-T6 aluminium).
a. Fatigue.

b. Fatigue.

c. Crack tip.

d. Crack tip.

Fig.6.48 Fracture morphology in S-T orientation
(7075-T6 aluminium).
Fig. 6.49 Fracture morphology in S-T orientation (7075-T6 aluminium).
Fig. 6.50 Fracture morphology in T-L orientation (7075-T6 aluminium).
Fig. 6.51 Fracture morphology in T-L orientation

(7075-T6 aluminium)
The mode of fatigue crack propagation was found to be dependent on the stress intensity used. In Fig. 6.25(c) where the maximum stress intensity was low (about 23 MPa/m) the fracture mode was a mixture of transgranular and intergranular; but at high fatigue stress intensity (about 29 MPa/m), only the transgranular mode was observed as in Fig. 6.25(d). The line of demarcation between fatigue crack and overload fracture was difficult to define, although generally the fracture mode at the tip of the crack was smooth intergranular with rather extensive secondary cracks as can be observed in Fig. 6.26(a). Immediately in front of the fatigue crack, the fracture was a mixture of dimples and cleavage facets (Figs. 6.26(a) and 26(b)). This was carried onto the main overload fracture where the fracture mode contained approximately equal proportion of dimples and cleavage (see Figs. 6.26(c) and 26(d)). Occasionally, few patches of intergranular dimples could be observed, as in Fig. 6.27(a). Many of the dimples contained small impurity particles which were analysed, using energy dispersing X-ray analysis, to be MnS. Fig. 6.27(b) shows the typical elongated shear dimples on the obtuse angle shear lip of a 4mm compact tension specimen.

The fracture morphology of the 100°C tempered specimen was almost identical to that in the as-quenched state. The overall overload fracture mode remained relatively smooth with few MnS stringers being exposed, as can be observed in the stereo micrographs in Fig. 6.27(c) and (d). At the tip of the fatigue crack, smooth intergranular fracture with extensive secondary cracking dominated the behaviour (Fig. 6.28(a)) while in the main overload region, the fracture morphology consisted of roughly equal proportion of dimples and cleavage (see Figs. 6.28(c) and 28(d)). A magnified view of the rather shallow dimples and the cleavage facets can be seen in Figs. 6.28(b) and 29(a) respectively. In the shear lip region, it was observed that fracture occurred by shearing between different layers or bands of MnS stringers, thus producing what appeared to look like steps (see Fig. 6.29(b)).

Tempering at 200°C produced a fracture morphology that was significantly different from that in the as-quenched or 100°C tempered state. Here, the overload region showed a large number of stringers being exposed during the fracture and the overall surface becoming more rugged (see Fig. 6.29(c)). This surface relief was a consequence of the formation of steep and abrupt shear facets between different layers or bands of MnS.
stringers. The fractograph in Fig.6.29(d) shows that the crack tip can be clearly identified and that the immediate crack tip contains mainly ductile dimple fracture. Even in the overload region the fracture morphology was observed to be ductile dimple fracture (see Fig.6.30(a) which is the higher magnification view of the central area of the main fracture shown in Fig.6.29(c) ). The surface consisted of a larger number of stringer troughs. Enlarged views of dimples can be seen in Figs.6.30(b), 30(c) and 30(d). The cleavage facets which were common in the as-quenched and 100°C tempered specimens, were found to occur only in few isolated areas.

The fracture morphology underwent a drastic change when the tempering temperature was increased to 300°C. Although the main fracture region at low magnification still appeared to be relatively similar to that at the 200°C temperature (Fig.6.31(a) and 31(b) ) in that the surface was uneven and consisted of numerous stringers, at high magnification the fracture was one of extensive cleavage facets. At the crack tip, the morphology was a mixture of ductile dimples and cleavage (Fig.6.31(c) ), but as the crack propagated, the latter fracture mode dominated the larger part of the fracture. Figs.6.31(d) and 32(a) show some typical examples of such cleavage facets. It can be observed in Fig.6.32(a) that even in areas of extensive cleavage, fine dimples formed by microvoid coalescence are found on the shear planes. A noticeable portion of cleavage facets (270) was also observed on the shear lips (Fig.6.32(b) ), although the overall morphology here could be classified as shear facets containing elongated dimples.

The 400°C tempered specimens were observed to have a further increase in surface relief as shown in the stereo fractographs in Figs. 6.32(c) and 32(d). The fatigue crack surface consisted of transgranular fatigue fracture with numerous secondary cracks, that appeared to be fatigue striations (Fig.6.33(a) ). The crack tip was characterized by dimples (Fig.6.33(b) ). The main fracture morphology was mixed mode: with shear facets linking between different layers of MnS stringers and the flat areas between the shear facets containing both cleavage and dimples as shown in Figs.6.33(c) and 33(d) respectively. The fracture mode on the shear lips was similar to that in the 300°C tempered specimens. Fig.6.34(a) shows the occasional occurrence of cleavage on the shear lip surface and Fig.6.34(b) shows a low magnification view of the step profile and the shear facets on the acute angle shear lip.
At the 450°C tempering temperature, the fracture morphology of the overload region was dominated by ductile dimples and abrupt shear facets as shown in Fig. 6.34(c). Cleavage, common to the 400°C tempered specimens, was seldom observed. Well formed ductile dimples (Fig. 6.34(d)) were found on the flat area between stringer troughs. At the immediate crack tip, the fracture mode was ductile (see stereo fractographs in Figs. 6.35(a) and 35(b)). Stereo viewing of the crack tip showed there to be a well formed stretched zone.

The fracture surfaces of materials tempered at 500, 550, 600 and 650°C were very similar to one another. These are shown in Figs. 6.35(c) to 37(b) for a 500°C temper, 6.37(c) to 38(d) for 550°C, 6.39(a) to 40(d) for 600°C and 6.41(a) to (d) for a 650°C temper. The surface relief in the overload region increased with the increase in tempering temperature so that at high tempering temperatures, the surfaces were extremely uneven and rugged with abrupt perpendicular shear facets linking different elevations of MnS stringer bands (compare, for example, the stereo fractographs in Figs. 6.35(c) and 35(d) with Figs. 6.39(a) and 39(b)) thus forming terrace-type of fracture. In the fatigue area, irregular striation like crackings (see Figs. 6.39(c) and 39(d)) were observed to occur frequently as the tempering temperature was increased. The immediate crack tips were dominated by shearing (Figs. 6.36(c), 36(d) and 41(b)) and ductile dimples (Figs. 6.37(d) and 41(c)). Well defined stretch zones were observed (Figs. 6.36(c) and 36(d)). In the main fracture area, well formed dimples of various sizes (Figs. 6.37(a), 37(b), 38(a), 38(b) and 41(d)) with spherical MnS inclusions (Figs. 6.40(a) and 40(b)) were found. Stereo fractographs (Figs. 6.38(c), 38(d), 40(c) and 40(d)) show elongated dimples on the shear lip surfaces.

A summary of the effect tempering temperature has on the fracture morphology of compact tension specimens austenitized at 850°C and fractured at room temperature for the R-L orientation is given in Table 6.10 below:
### Table 6-10: Effect of tempering temperature on fracture mode.

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<th>Tempering Temperature °C</th>
<th>Immediate Crack Tip</th>
<th>Overload Fracture Area</th>
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<td>Dimples and cleavage</td>
</tr>
<tr>
<td>100</td>
<td>Smooth intergranular</td>
<td>Dimples and cleavage</td>
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<td>Ductile dimples</td>
<td>Dimples</td>
</tr>
<tr>
<td>300</td>
<td>Ductile dimples and cleavage</td>
<td>Cleavage</td>
</tr>
<tr>
<td>400</td>
<td>Dimples</td>
<td>Dimples and cleavage</td>
</tr>
<tr>
<td>450</td>
<td>Dimples</td>
<td>Dimples and shear</td>
</tr>
<tr>
<td>500</td>
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</tr>
<tr>
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</table>

(b) **Effect of crack plane orientation on Comsteel En25 specimens tempered at 500°C**

The fracture morphology of the R-L orientated specimens tempered at 500°C was discussed in the previous section. The fracture features in the R-C and L-R orientations are shown in Figs.6.42, 43 and 6.44, 45 respectively.

Figs.6.42(a) and 42(b) show the stereo fractographs of the overall fracture morphology of the fast fracture region. Similarity with the R-L orientation was observed in that the fracture was mixed mode with shear facets linking between MnS stringers forming a zig-zag type topography. Well formed ductile dimples were observed in flat areas (Fig.6.43(a) ). The fatigue crack was found to be transgranular with stringer troughs aligned in the direction of the crack front as shown in Fig.6.42(c). Steep shear facets (Fig.6.42(c) ) and ductile dimples (Fig.6.42(d) ) were observed at the immediate crack tip. At the shear lips, the predominant fracture mode was shear. Although generally featureless, fine dimples as seen in Fig.6.43(b) were easily discernable. Stereo fractographs in Figs.6.43(c) and (d) show that the typical fracture mode on the shear lip surfaces.
The overall fast fracture region in the L-R orientation did not exhibit terrace or zig-zag type fracture similar to the R-L and R-C orientations. Stereo fractographs in Figs.6.44(a) and (b) show that the fracture was rather flat. Numerous cavities formed around the MnS inclusions were observed. The fatigue region consisted of intergranular fracture with numerous irregular striation like crackings (Fig.6.44(c)). At the immediate crack tip, the fracture morphology was essentially ductile dimple rupture as shown in Fig.6.44(d). Secondary cracks were however frequently observed at the demarcation between the fatigue and the overload regions (Fig.6.45(a)). The entire overload region was dominated by ductile dimple fracture. An example of the well formed dimples is shown in Fig.6.45(b). This fracture feature also persisted on the shear lips, as can be seen from Figs.6.45(c) and (d) where elongated dimples were generally observed at high magnification.

(c) Effect of crack plane orientation on 7075-T6 aluminium alloy specimens

The fracture features of the specimens tested with crack planes in the L-S, S-T and T-L orientations are shown in Figs.6.46 to 6.47, 6.48 to 6.49 and 6.50 to 6.51 respectively. It was observed that for this material in the as-cast conditions, the fracture morphology was not affected by the difference in specimen orientation.

The fatigue crack region, irrespective of the stress intensity used in fatiguing, was dominated by 'cleavage like' fracture as shown in Figs.6.46(a) and 6.48(a) and 6.50(a). At higher magnification, Fig.6.46(b) the smooth flat planes were found to contain river patterns and ridges. Well developed fatigue striations were easily identified at even higher magnifications (Figs.6.46(c) and 6.48(b)). It has been reported that such 'cleavage like' fatigue fracture surfaces are not caused by a true cleavage mechanism.

The demarcation between fatigue and overload fracture was generally not well defined. At the immediate crack tip, the fracture feature appeared to be intergranular as shown in Figs.6.46(d), 6.48(c) and (d) and 6.50(b). This fracture feature persisted in the overload region, as shown in the stereo fractographs in Figs.6.47(a), (b), 6.49(a), (b) and 6.50(c), (d). At low magnification, the intergranular fracture appeared to be flat and smooth but fine shallow dimples were observed on these surfaces at high magnifications. Examples of such dimples are shown in Figs.6.47(c), (d),
6.49(c), (d) and 6.51(a). Closer examination of the dimples showed that they were formed by microvoid coalescence at tiny inclusion sites (see Figs.6.47(d) and 6.49(d)).

In the 45° slant fracture or shear lip regions, the surfaces were observed to be dominated by very fine shallow dimples (Fig.6.51(b) and (c)). These dimples were often found to be elongated, as can be seen in Fig.6.51(d), indicating that the fracture mode was of shear rupture.

6.11 METALLOGRAPHY

Figs. 6.52 (a, b) show the prior austenite grain size of Comsteel En25 austenitized at 850°C. An average grain size of 7μm was obtained. This heat treatment produced a very fine martensite network. The microstructure of the 7075-T6 aluminium alloy is shown in Figs.6.52(c) and (d). The individual grain size was estimated to be approximately 190μm. Coring due to non-equilibrium freezing was generally observed in the specimens that had been etched in 1% HF (see Fig.6.52(c)).

Etching the steel specimens using picric acid/"Teepol" with or without conc. HCl showed that the grain structure in the triangular shear lip region appeared to be identical to that in the undeformed material. Generally, shearing on the shear lip surface was observed to deform a band of only a few grains in width (see Fig.6.53). This is probably due to the rather low strain-hardening properties of the material even in the low yield high toughness 600°C tempered condition.

Standard and modified Fry's reagents (see section 5.14) were used in an attempt to reveal the distribution of plastic strain in the shear lip region. Fig.6.54 shows a typical example of the effect modified Fry's reagent has on a compact tension specimen tempered at 500°C. It was noted that the etchants affected only the areas where thickness contraction had occurred. They were not sensitive enough to detect deformation of lower strains. With an etching time of well over 24 hours, the specimens were not suitable for microscopic observation.
a. Comsteel En25.

b. Comsteel En25. 10 μm

c. 7075-T6 Aluminium.

d. 7075-T6 Aluminium. 100 μm

Fig. 6.52 Grain size of Comsteel En25 and 7075-T6 Aluminium.
Fig. 6.53 Deformation in shear lip formation.

Fig. 6.54 Distribution of plastic strain in shear lip formation.

Fig. 6.55 Micro-hardness test.

a. Before etching.

b. After etching.
6.12 MICROHARDNESS

A typical example of the microhardness indentations produced on the region of the slant fracture is shown in Fig. 6.55. The lengths of the longer diagonals of all the indentations were found to be identical having a value of 0.0745 mm which corresponded to a 136° Diamond Pyramid Hardness numbers of 334 or a Knoop Hardness number of 342. Consequently, no variation in hardness between the shear lip region and the original material could be detected using microhardness testing.

From the etching (section 6.11) and microhardness experiments on the fractured specimens, it can be concluded that the deformation zone in the formation of the shear lip takes place in a thin region on either side of the shear lip surface. Because of the low value of the strain hardening exponent of the material, the thickness of this region was estimated to be only a few grain diameters so that the microhardness testing was not sensitive enough to detect the deformation.
CHAPTER 7

DISCUSSION OF RESULTS
CHAPTER 7

DISCUSSION OF RESULTS

7.1 INTRODUCTION

The chapter is essentially divided into two main sections. The first section deals with the tensile and fracture toughness data. The effects of tempering temperature, specimen size, crack plane orientation and impurities are discussed. The relationship between shear lip size and fracture toughness is examined in the next section. The validity of the fracture toughness values obtained from the ASTM standard analysis is considered. The necessity to employ yielding fracture mechanics approaches to obtain the true plane strain fracture toughness is explained. The chapter concludes with the applications and the possible limitations of the shear lip size/fracture toughness relationship.

7.2 TENSILE PROPERTIES

The effect of tempering temperature on the tensile specimens obtained from the fractured compact tension specimens is shown in Fig. 6.1. The data are typical of low alloy steels in that the yield and tensile stresses decrease with the increase in tempering temperature. The general increase in strength in the steel with composition N most probably resulted from the higher silicon content. At high tempering temperatures, the difference in strength between the two steel composition decreased. As the effect of sulphur and phosphorus on strength remains roughly constant with tempering temperature(74), this latter effect could result from a decreasing effect of silicon in strengthening the material(77). In addition, the difference in strength at the 600°C temper may have been caused by the difference in heat treatment process in that the composition N specimens were tempered in the electric muffle furnace and thus may have been slightly decarburized. The composition O specimens were tempered in a salt bath.

Generally, high strength low alloy steels show an increase in 0.2% proof stress when tempered at 200°C(272). A small increase is observed for the composition N steel and a small decrease for composition
0 (see Fig.6.1). However, in this range of low tempering temperatures, the 0.2% proof stresses could not be determined very accurately since they were obtained from the load versus crosshead displacement records where no distinct yield point was noted. Thus the effect is likely to be hidden by experimental scatter. For material tempered at 300°C and above, a definite yield point was observed in the load versus crosshead displacement record. The 0.2% proof stresses of these materials were thus determined more accurately.

For the tensile strengths, steel of composition N shows an increase in strength and composition O a decrease between the as-quenched and the 100°C temperature. The behaviour follows the same trend as the yield stress.

The tensile data obtained in the present investigation compare favourably with the works of Clark (255) and Logan and Crossland (273), both used Comsteel En25 Steel with very similar compositions. This comparison is shown in Fig.7.1.

The increase in 0.2% proof stress and tensile strength of En25 steel with decrease in testing temperature below ambient (see Table 6.2) is confirmed by Sargisson (256) and Ferguson et al (274). Using the same steel tempered at 350°C, they found a 10% increase in proof stress when the test temperature was lowered from ambient to -196°C.

Fig.7.2 shows the yield strength of the 7075-T6 as-cast aluminium alloy used in the present investigation compared with results of Nelson and Kaufman (275). It can be observed that the trend of the three heat treatments is identical, with the T6 condition sandwiched between the T651 and T7351 conditions.
Fig. 7.1 Comparison of tensile data (Comsteel En25).

Fig. 7.2 Comparison of tensile data (7075-T6 aluminium, below ambient).
7.3 FRACTURE TOUGHNESS DATA

7.3.1 Effect of Tempering Temperature

Fig. 6.5 shows the effect of tempering temperature on the fracture toughness of the 25mm thick Comsteel En25 specimens given the normal commercial hardening treatment of austenitizing at 850°C and quenching in an oil bath. While only a slight increase in fracture toughness was observed at 100°C temper, the tempering at 200°C produced a nearly 30% increase in fracture toughness over that in the as-quenched state. The higher fracture toughness value at 200°C is reflected in the increase in surface relief on the fracture surface of the specimen (see Fig. 6.29(c)). Comparing the fracture surface at 200°C (Fig. 6.29(c)) with that at 100°C (see Figs. 6.27(c) and (d) for example), it can be observed that the increase in surface relief exposes larger quantities of MnS stringers. As the higher fracture toughness gives rise to a larger plastic zone size, the probability of encountering a MnS inclusion at a greater distance from the crack plane as the crack propagates is higher and thus the relief will be more rugged. It appears that the crack front changes its direction by adopting a shearing mechanism to link up bands or layers of MnS inclusions at different elevations. The shear planes were generally observed to be perpendicular to the plane of crack propagation. Consequently, the crack advances in a step-wise manner resulting in a terrace-type fracture surface (see Fig. 7.3).

Fig. 7.3 Schematic diagram of terrace-type fracture.
Due to the orientation of the specimen, this terrace-type fracture was observed to continue into the shear lips at the free surfaces of the specimen.

A significant drop in fracture toughness occurs as the tempering temperature is increased to 300°C (see Fig.6.5). The decrease in toughness value is accompanied by significant changes in fracture morphology, microstructure of the material and the shear lip size (see Fig.6.6). Observation in the scanning electron microscopy indicated a change in fracture mode from ductile dimples in the 200°C temper to extensive cleavage with the facet size being the grain size. At this tempering temperature, a significant increase in the precipitation of cementite occurs in the microstructure. Clark (255) found that the cementite precipitated within the martensite laths and along the lath boundaries. He thus suggested that the tempered martensite embrittlement is the result of the reduction in cohesive strength of the martensite-ferrite interface caused by impurities rejected when the cementite particle grows (276).

Tempering at 400°C produced a considerable drop in the yield stress (see Fig.6.1). However, this decrease is not reflected in a large increase in the fracture toughness value. In the case of the steel with composition N, compared with the 200°C tempered specimen, an increase in toughness of about 10% was observed while for composition O steel, the increase is negligible (64.66 MPa/m at 200°C and 64.68 MPa/m at 400°C). Scanning electron microscopic observation revealed that the fracture surface of the 400°C tempered specimen consisted of a large proportion of cleavage fracture (see Fig.6.33(c)) whereas the predominant fracture mode in the 200°C temperature was ductile dimples. The discrepancy is therefore the consequence of the tempered martensite embrittlement in the 400°C tempered material. Because of the drop in yield stress in the 400°C temper, however, the plastic zone size which is proportional to \((KIC/OYS)^2\) is larger than in the 200°C temper. The larger plastic zone produces greater surface relief. The macroscopic fracture mode is as with the 200°C and 300°C tempers, shear between MnS stringers giving terrace-type fracture. In addition, the ductile dimples formed in the 400°C temper appeared to be deeper and larger (compare Figs.6.33(b) and (d) with Fig.6.29(d) and 6.30(c) and (d)) indicating that the material was more ductile and tougher than that tempered at 200°C.

Further reduction in yield and tensile stresses was observed when the material was tempered at 450°C. But unlike the 400°C tempered specimens, here the fracture toughness was found to increase considerably. Cleavage...
fracture evidenced in the tempered martensitic embrittlement region could no longer be found. The fracture morphology consisted of ductile dimples and shear facets that linked the bands of MnS inclusions on different planes thus forming the terrace-type fracture. It has been observed (255) that at this tempering temperature, there is also a microstructural change in that the cementite in the material changes from a needle-like to a shorter cylindrical-like precipitate with some of the precipitates spheroidizing. At this temper, because of the increased fracture toughness and decreased yield stress, the surface relief was more rugged, but the ductility of the material had increased. An indication of the increased ductility was revealed from the observation of the shear lip formation and the amount of through-thickness contraction occurring at the surfaces of the fractured specimens. Although the shear lip size increased extensively (refer to Fig.6.6), it was not as well formed and distinct as those formed at lower tempering temperatures (compare for example, Fig.6.23(e) with Fig.6.24(a)). The through-thickness or out-of-plane contraction, as shown schematically in Fig.7.4, is the reduction in specimen thickness along the plane of crack propagation.

![Schematic diagram of through-thickness contraction](image)

Fig.7.4 Schematic diagram of through-thickness contraction (277).

This was measured with a microscope by focusing first on the undeformed specimen surface, then on the bottom of the contracted region in the plane of the fracture surface. The through-thickness contraction was taken as the difference of the readings on the micrometer which was attached to the microscope for focusing. Contraction occurs as a result of the loss of constraint by the surrounding elastic material and is an indication that the plastic zone is excessively large in relation to the specimen thickness (15,65). The amount of through-thickness contraction increases with
tempering temperature above 400°C as shown in Fig. 7.5.

Fig. 7.5 Variation of surface depression with tempering temperature.

Tempering at 500°C and higher further increased the value of fracture toughness and surface relief. At the higher tempering range however, the fracture toughness versus tempering temperature curve shown in Fig. 6.5 appears to level off. At these higher tempering temperatures, the fracture surfaces of the specimens contained a lot more MnS stringers (see Figs. 6.37(c) and 6.41(a)). It was the MnS inclusions that appeared to control the fracture characteristics of the material. In the flat areas between stringers, ductile dimples were observed (Fig. 6.37(a) and 6.41(c)). The path of crack propagation was generally one of shear from inclusion band to inclusion band resulting in a
Fig. 7.6 Typical example of terrace-type fracture path.
step-type fracture. A typical example of such fracture path is shown in Fig.7.6 where it can be observed that the shear planes are frequently perpendicular to the direction of crack propagation.

While the through-thickness contraction on the specimen surfaces further increased with the tempering temperature (Fig.7.5), the shape of the shear lips deteriorated in that less and less well formed shear lips were observed as shown in Fig.6.24. This implies that at these high tempering temperatures, the size of the plastic zone is no longer negligible compared with the specimen thickness and the crack length, and it is expected that linear elastic fracture mechanics fails to be applicable. This will be discussed further in section 7.4.1.

Fig.7.7 shows a comparison, of the relation between fracture toughness and tempering temperature obtained in the present investigation, with those by Clark (255), Ferguson et al. (274) and Logan et al (273).

Fig.7.7 Comparison of fracture toughness data (Comsteel En25).
All four investigations made use of quenched and tempered Comsteel En25 austenitized at 850°C. The compositions of the steels used are reported in Table 7.1. The specimens employed were all of the same type, being compact tension specimens. The orientations of the specimens differed in that C-L and R-L were used in the present study and C-R and R-C were used by both Clark and Logan et al. The data by Ferguson et al., on the other hand, were obtained with cracked-notched round bar specimens in the L-C and L-R orientation. It is not surprising, therefore, that their fracture toughness data are generally higher especially at tempering temperatures above 300°C.

The trend of the fracture toughness versus tempering temperature curve for the present study compares favourably with those of Clark and Logan et al. Because of the orientation of the specimens, being in the C-L direction which offers the least resistance to crack propagation, the fracture toughness values are consistently lower except at higher tempering temperatures above 500°C. At 600°C for example, the value of fracture toughness was observed to be the highest of the three curves. This difference is explained in section 7.4.1 where the validity of the fracture toughness data is examined.

The data of Ferguson et al.(274) differ considerably from the present results. The discrepancy is due to the size of the specimens used. The specimen size requirement to attain plane strain conditions in a cracked-notched round bar is that the major diameter should exceed \(10\frac{(K_{IC}/\sigma_{YS})^2}{20}\). A specimen of 9.27mm diameter used by Ferguson therefore indicates that fully plane strain conditions were not met in any of the tests except at 350°C, where by testing specimens of different diameters, it was shown(256) that a diameter of 9.27mm was sufficient to obtain a valid plane strain test. As cracked-notched round bar specimens of insufficient diameter have been found to depress the fracture toughness value(171), it is expected that the fracture toughness data of Ferguson is valid at a 350°C temper but should be higher than those exhibited in Fig.7.7 at higher tempering temperatures. Similarly, Ferguson's result at 210°C temper where the fracture toughness has been shown by the results of the present study as well as Clark(255) and Logan et al.(273) to be higher than that at 350°C, is invalid. A diameter of 9.27mm which was just sufficient to determine the fracture toughness value at 350°C is obviously not suitable in the higher toughness temperature range of 200°C but was probably sufficient for the as-quenched and 100°C temper.
<table>
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Table 7.1: Composition of En25 used in various investigations.
7.3.2 Effect of Specimen Thickness

The influence of specimen thickness, \( B \), on the fracture toughness, \( K_c \), of Comsteel En25 tempered at various temperatures is shown in Fig.6.7. At all tempering temperatures, the trend of the \( K_c \) versus \( B \) curves is very similar to that generally observed in the literature (see for example, References 21, 27 and 33) in that \( K_c \) reaches a limiting value at large specimen thickness. With the present data, this limit is not reached for tempers over 450\(^\circ\)C as the curves have not plateaued with thickness. At low tempering temperatures, the limiting fracture toughness is, within a maximum error of less than 5\%, equal to the plane strain fracture toughness, \( K_{IC} \), value. This can also be seen from Fig.7.8 where for these data, \( K_c = K_{IC} \). However, at tempering temperatures higher than 450\(^\circ\)C, the measured \( K_{IC} \) values although they satisfied all the ASTM requirements were not the limiting plateau values.

![Graph showing comparison between \( K_c \) and \( K_{IC} \)]
These values will be somewhat in excess of the limiting value but less than the $K_C$ value as can be seen in Fig.7.8. It will be shown later that this effect results from excess plasticity at the crack tip.

Although the size of the shear lips at the free surfaces of the specimens remained constant along the fracture path and at a given tempering temperature, this was invariant with the specimen size, the proportion of shear lips increased as the specimen thickness is decreased. Since the formation of shear lips and the deformation in the 45° slant fractures require higher energy than that in the flat fracture (278), the total energy required to fracture a particular specimen is accordingly higher in the thinner specimens than the thicker ones (219). Thus the fracture toughness $K_C$ value increases as the specimen thickness decreases as shown in Fig.6.7. The proportion of shear lips in the as-quenched specimens was observed to increase by about 15% when the specimen thickness was decreased from 25mm to 4mm whereas for the 600°C temper, this increase was over 80%. Thus it is not surprising that the ratio of $K_C$ for 4mm specimen over that for 25mm increases from 1.5 in the as-quenched conditions to 2 in the 600°C temper.

Using the present data, comparison is made with the theoretical models of Bluhm (44), Krafft et al. (47), Tetelman et al. (30) and Hahn et al. (49) which are predictions for variation of fracture toughness versus specimen thickness. The observation from the present study, that the size of the shear lips is independent of the specimen thickness, confirms the assumption made in each of the models. Table 7.2 shows the theoretical fracture toughness results (in MPa/m) from Bluhm's model compared with the experimental results. The experimental results appear in parentheses while for each tempering temperature, the two data points used to arbitrarily fit the experimental data to the theoretical curve are marked with asterisks. At the plateau region and also at the rising portion of the curve, the model predicts rather accurately the values of $K_C$ at all tempering temperatures. However, the model fails to predict the levelling off of the $K_C$ versus $B$ curves as $B$ decreases towards the 4mm thickness. Examination of Fig.6.7 shows that at small specimen thicknesses the curves have the tendency to flatten. This tendency has also been observed by Allen (33) using 7075-T6 aluminium alloy although Weiss and Yukawa (23) found otherwise for H11 steel. Bluhm (44) suggested if the mechanism of the shear lip formation was a surface phenomenon rather than a volumetric phenomenon, then the fracture toughness in the region of specimen thicknesses with 100% shear would remain constant.
<table>
<thead>
<tr>
<th>Tempering Temperature (°C)</th>
<th>Nominal thickness of specimen, mm</th>
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</table>
| 140.0                     | 58.9 (52.7) 60.8 (70.2) 62.2 (64.2) 64.2 (65.8) 67.2 (69.8) 72.7 (73.3) 72.7 (84.9)  
| 100                        | 58.8 (55.8) 60.0* (60.2) 61.3 (58.5) 63.2 (61.1) 65.7 (66.1) 69.7 (74.7) 76.5 (73.1) 91.6* (91.6)  
| 200                        | 67.0 (66.7) 68.6* (68.8) 70.7 (71.6) 73.7 (72.0) 77.4 (72.5) 83.5 (82.9) 93.4* (93.4) 114.9 (96.2)  
| 300                        | 60.6 (60.8) 62.5* (62.5) 64.9 (66.6) 68.2 (70.5) 72.5 (71.2) 79.0 (81.1) 89.8* (89.8) 112.6 (94.7)  
| 400                        | 66.6 (64.7) 69.8* (69.7) 73.8 (70.7) 79.3 (74.3) 86.1 (86.6) 96.4* (96.5) 112.9 (102.5) 146.5 (106.3)  
| 500                        | 90.4* (90.4) 94.7 (92.0) 99.9 (100.8) 107.1 (107.1) 116.2 (125.6) 130.0 (160.3) 152.2 (159.1) 197.3 (190.9)  
| 600                        | 105.2 (105.3) 116.0* (112.8) 128.2 (120.5) 144.7 (140.1) 164.6 (193.0) 192.9* (199.0) 235.6 (202.2) 320.6  
| 7075-T6                    | 28.7 (29.9) 29.1* (29.2) 29.8 (29.7) 30.6 (31.5) 31.8 (29.6) 33.7* (33.7) 36.9 (32.4) 43.8 (41.2)  

Table 7.2: Comparison of Bluhm's model with experimental results
However, the data in the present programme cannot confirm this as only the 4mm thick specimens tempered at 550°C and 600°C exhibited 100% shear (see Appendix J) and thinner specimens were not investigated.

The model for the effect of specimen thickness on fracture toughness proposed by Hahn, Hoagland, Rosenfield and Sejnoha

requires the determination of the shear lip depression width, \( \ell \), as shown in Fig.2.10. This parameter was measured on the surfaces of the compact tension specimens after they were fractured apart. To reveal the contracted areas, the specimens were lightly polished so that a contrast between the depressed and the undeformed areas could be observed (see Fig.6.54). \( \ell \) was generally found to remain approximately constant over the centre 15mm of the fracture surface. For a constant tempering temperature, \( \ell \) was roughly independent of the specimen thickness although there was a tendency for it to decrease in the 4mm thick specimens at high tempers. This may have been caused by buckling of the thin specimens when they were loaded to fracture. Fig.7.9 shows the relation between depression width, \( \ell \), and the shear lip size, \( B_{SL} \), as suggested by Hahn et al.\(^{(49)}\). The approximately linear relationship can be expressed as \( \ell = 4.0 B_{SL} \) which compares well with Hahn's experimental result of \( \ell = 3.8 B_{SL} \).

Table 7.3 summarizes the comparison between some of the experimental data, with the theoretical prediction of the fracture energy, \( G_c (\text{Nm/m}^2) \), as a function of specimen thickness using the model of Hahn et al.\(^{(49)}\). Parentheses and asterisks have the same meaning as those used in Table 7.2. Like Bluhm's model, Hahn's predicts the experimental data accurately at low tempering temperature and large specimen thickness but is less accurate at high tempering temperature and small specimen thickness where the experimental curves as shown in Fig.6.7 begin to level off.

The models of Krafft et al.\(^{(47)}\) and Tetelman et al.\(^{(30)}\) involve the use of the strain energy release rate where the fracture occurs entirely by shear, that is, \( G_{c_{max}} \) or \( G_c (45°) \). They are therefore only applicable to the 550 and 600°C tempered specimens where 100% shear was found to occur for the 4mm thick specimens. Assuming that the value of \( G_{IC} \) is achieved at a specimen thickness of 25mm the estimation of the experimental \( G_c \) values using both models is shown in Table 7.4. It can be observed that agreement between theoretical and experimental results is generally very poor.
Fig. 7.9 Relationship between shear lip size and depression width.
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<tr>
<th>Tempering Temperature °C</th>
<th>Nominal thickness of specimen, mm</th>
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<td>550</td>
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<tr>
<td>7075-T6</td>
<td>14.2*</td>
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<tr>
<td>Aluminium Alloy</td>
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Table 7.3: Comparison between experimental data with the theoretical prediction of $G_c$ using Hahn's model\(^{(49)}\)
<table>
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<tr>
<th>Tempering Temperature (°C)</th>
<th>Nominal thickness of specimen, mm</th>
<th>Model</th>
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Table 7.4: Model of Krafft et al. and Tetelman et al. compared with experimental results
Both models underestimate $G_c$ for thin specimens and for thick specimens, the experimental data are sandwiched between the two theoretical estimations with Tetelman's analysis grossly overestimating. The lack of agreement with Tetelman's model is not surprising since a surface mechanism for the shear lip formation is implied in the analysis. Other factors that contributed to the discrepancy of both models were the values of $G_{IC}$ used in the estimation and the proportion of slant fracture corresponding to these $G_{IC}$ values. At these high tempering temperatures, even the thickest specimens were not at the limiting plateau region. The proportion of shear lips under these conditions was found to range from 14 to 18% which could no longer be considered insignificant. In addition, there is the possibility that not the entire central square fracture grows under a plane strain stress state $^{(279)}$. These factors contribute to the larger experimental values of $G_c$ than those predicted in the simplified theoretical analysis.

From the afore-mentioned discussions, it can be concluded that the models proposed by Bluhm $^{(44)}$ and Hahn et al. $^{(49)}$ are sufficiently accurate in predicting the variation of $K_c$ with specimen thickness at all tempering temperatures except in the high tempering temperature, low specimen thickness region. Models proposed by Tetelman et al. $^{(30)}$ and Krafft et al. $^{(47)}$ are however generally less accurate. Nevertheless the models provide a means by which the fracture toughness of a material at any thickness can be estimated once the fracture toughness values of two specimens of different thicknesses in the intermediate thickness range are determined. For the models of Tetelman and Krafft, there is a complication in that one of the specimens must be thick enough to ensure that plane strain fracture toughness can be obtained, while the other specimen has to have thickness equal to $B_0$ (Fig.2.9). The latter thickness can be achieved by measuring the shear lip size on the former specimen and multiplying by two. The accuracy of the model will essentially depend on the justification of the assumptions made. That the size of the shear lips is independent of the specimen thickness is well confirmed by the present data. The formation of shear lips being a volumetric mechanism is a reasonable representation in materials with high work-hardening rate. If the work-hardening rate is low, the deformation occurring in the shear lip formation will tend to be concentrated along the 45° shear planes. The present results show however, that even for the range of values of work-hardening rate between 0.025 and 0.064, the volumetric mechanism yields sufficient accuracy in predicting the variation of $K_c$ with specimen thickness.
7.3.3 Effect of Anisotropy

The effect of specimen orientation on fracture toughness was investigated using 25mm thickness compact tension specimens, which in the case of Comsteel En25 were tempered at 500°C. The results are tabulated in Table 6.4.

Consider first of all, the data for the steel specimens. It is evident that the longitudinal (L - R or L - C) orientation results are considerably higher than those in the transverse (R - C or C - R) and short-transverse (C - L or R - L) directions. Testing under the same conditions, Clark (255) obtained an average fracture toughness value of 86.73 MPa/m in the transverse orientation (see Fig.7.7) which compares well with the value of 87.35 MPa/m found in the present work for the R - C orientation. However, Logan et al. (273) using a specimen thickness of 31.7mm demonstrated that the plane strain fracture toughness in the R - C orientation was 82.7 MPa/m. Bearing in mind that the apparent fracture toughness value determined using compact tension specimens of insufficient size for a valid KIC test is higher than the true KIC values, the lower value of Logan et al. would indicate that a specimen thickness of 25mm was inadequate to obtain a limiting fracture toughness value independent of specimen size. The fracture toughness in the longitudinal direction (150.36 MPa/m) was found to be nearly twice that for the transverse directions. Using 9.27mm diameter cracked-notched round bars tempered at 430°C, Ferguson et al. (274) achieved a toughness value of only 85.2 MPa/m in the longitudinal orientation (see Fig.7.7). This value which is only half of the present value at 500°C temperature was underestimated as a result of misalignment problem (255). Because of the specimen size insufficiency in the present work, it is believed that the true KIC should be slightly lower than 150.36 MPa/m.

The variation of KIC with crack plane orientation can be explained in terms of the distribution of the non-metallic inclusions. The major inclusions (see Figs.6.36(b) and 6.44(d)) were found to be, for all tempering temperatures, Type II inclusions. These inclusions were identified, using energy dispersive X-ray analyses, to be manganese sulphide (MnS). The cylindrical MnS stringers are postulated to be present as interconnecting colonies during the solidification of the ingot (280). When the ingot is subsequently rolled, the colonies are elongated and oriented...
such that their long dimensions lie in the rolling direction. The resulting inclusion distribution remains irrespective of whether or not the material undergoes further heat treatment. It is this directional alignment of the MnS inclusions that gives rise to anisotropic effect in the fracture toughness data. Lowes et al.\(^{(281)}\) postulated that in the case of Charpy testing, the Charpy upper shelf energy is proportional to \(1/NL^2\) where \(N\) is the number of inclusions per unit area and \(L\) is the mean length of the inclusions. This relationship was a modification of the results of Baker et al.\(^{(282)}\) and Brownrigg et al.\(^{(283)}\). Consider the distribution of inclusions in the hot rolled Comsteel En25 used in the present investigation, shown schematically in Fig.7.10, it can easily be seen that fracturing of specimens in the L-C or L-R orientations requires more energy than those in the R-C and R-L orientations.

Fig.7.10 Schematic diagram of distribution of inclusions in Comsteel En25.
With the former orientation, the crack must cut across the stringers with decohesion and fracturing of the stringers taking place in the process. But with the latter orientation, the crack encounters a larger projected area of inclusions and thus potential sources of fracture initiation (see later for exact mechanism). Fegredo\cite{284} working with plain C - Mn steel showed that the maximum load required to fracture 3/4" thick wedge open loading specimens in the L - T orientation was nearly twice that in the S - L orientation. Because of the vast difference in load to fracture, most of his specimens in the L - T orientation were found to have crack planes that turned through 90° to progress on the longitudinal section.

Fractographic observation of the fracture morphology using scanning electron microscopy revealed that the fracture mechanisms were different in each of the three specimen orientations. In the R - L orientation (see Figs.6.35(c) and (d)), a terrace-type fracture resulted from shear facets linking bands of MnS stringers at different elevations was observed. Very similar fracture features were seen in the C - R orientation (see Figs.6.42(a) and (b)) although very frequently the terrace-type fracture became zig-zag in nature. High surface relief was observed on both surfaces. On the other hand, the fracture morphology in the L - C orientation contained mainly ductile dimples and was in general flat and featureless (see Fig.6.44(a) and (b)).

The mechanism of the terrace-type fracture, which is similar to lamellar tearing\cite{284,285} appears to be one of shear fracture between the MnS stringers lying along the rolling direction forming "plateaus" or "steps" in the process. As can be seen in Fig.7.6, the length of the plateau is the same as the length of the MnS stringer and banding of these stringers together determines the width of the plateaus (see Figs.6.39(a) and (b)). This type of fracture has been observed previously in steels in the short-transverse orientation\cite{282,287,288} and also in aluminium alloys\cite{289}. Mathematical models that consider coalescence by lateral expansion of the voids and final fracture by a necking down of the intervening matrix are not applicable to the terrace-type fracture observed here. As shown in Fig.6.36(b) for example, fracture in this orientation (R - L) occurs by a highly localized shear mechanism with virtually no lateral expansion of the void. Such phenomenon is a consequence of the work-hardening behaviour of the steel matrix since this determines the level of strain concentration along shear-bands emanating from the crack tip as the fracture
propagates. The mechanism for the formation of the terrace-type fracture can be explained using a model similar to that proposed by Clayton and Knott\(^{(290)}\). In the present material, the steps are associated with the distribution of the major MnS stringers. Free surfaces may have been generated around these inclusions when the material is loaded. Therefore, as the crack front runs past the length of the elongated MnS inclusions, shear bands are formed at the crack tip under the action of the principal tensile stress as shown in Fig.7.11(a).

At the same time, voids around MnS inclusions ahead of the crack tip are formed (Fig.7.11(a)). As the stress increases, when a critical shear strain is reached, the fracture propagates along the band of localized shear joining the crack tip to the stringer ahead as shown in Fig.7.11(c). The fracture then changes its direction to propagate along the length of the inclusion (Fig.7.11(d)). The result of this is a step or terrace. Baker et al.\(^{(282)}\) suggested that the most favourable direction for inclusion linkage was a compromise between the direction of minimum separation of inclusions and the direction of least resistance to matrix fracture. If the strain-hardening capacity of the material is high, the deformation ahead of the crack tip will be diffused since a larger volume of material can participate

![Schematic diagram of terrace-type fracture mechanism.](image-url)
in the fracture process. A less well defined terrace-type fracture will therefore result or may completely be suppressed.

The phenomenon of a more zig-zag type fracture that occurred in the C - R orientation has also been observed previously in high strength steels \(^{(255,291,292)}\) with low work hardening capacity \((n=0.06 - 0.10)\). This type of fracture is, as with the terrace type, predominantly one of shear from inclusion to inclusion (Fig.6.42(c)) with each trough or peak in the fracture surface ending at a MnS stringer, which are now parallel to the crack front, thus giving a characteristic zig-zag appearance. The development of such a fracture can be similarly explained using the same model as the terrace-type fracture above \(^{(292)}\). Rice and Johnson \(^{(194)}\) have suggested that when the critical crack tip fracture strain, \(\varepsilon_f\) in a non-hardening material has a value of about 0.5, the fracture mechanism is by shear linkage probably following the bounding logarithmic spiral slip-lines. For Comsteel En25, using equation 4.8 the critical crack opening displacement, \(\delta_f\), at a 500°C tempering temperature was found to be 37\(\mu\)m. The major inclusion spacing, \(X_o\), has been found to be 130\(\mu\)m \(^{(255)}\). Using the expression \(^{(194)}\),

\[
\varepsilon_f = \frac{\delta_f}{X_o}
\]

the critical crack tip fracture strain can be evaluated to be 0.28 which is less than the required value of 0.5 in order for the strain field to be large enough to envelop a major MnS inclusion according to the blunting crack tip model \(^{(194)}\). It therefore appears that fracture in the C - R orientation occurs before sufficient crack tip blunting has taken place. And because of the rather steep shear planes in the shear linkage fracture (nearly perpendicular to the general plane of crack propagation), it is believed that logarithmic spiral slip-lines are not followed \(^{(255)}\). However, it has been suggested \(^{(290,293)}\) that when the critical crack tip fracture strain is less than 0.5, the process of crack growth is one of shear decohesion along straight slip-lines forming zig-zag or terrace-type fracture. This is substantiated by the observations in the present work. From the above discussion, the fracture toughness of a material in the R - C and R - L orientations can therefore be increased by, for example, reducing the sulphur content of the steel or by employing high-temperature homogenization treatment to spheroidize the inclusions \(^{(282)}\).
In the L - C orientation, the MnS stringers are perpendicular to the plane of crack propagation and therefore an entirely different fracture morphology was observed. No shear facets could be found and the surface consisted only of ductile dimples (Figs.6.44(a), (b) and (d) ). Neal and Doig (294) observed that when the MnS inclusions are stressed across the major axis as in the case of short-transverse or R - L orientation, cracking occurs along the inclusion matrix boundary but when the inclusions are stressed along their major axis, the inclusions fracture before the matrix. It can therefore be concluded that in the L - C orientation, fracture occurs mainly by the nucleation of voids at inclusion particles. Some secondary crackings along bands of inclusions in the direction of the applied stress (Figs.6.44(a), (b) and 6.45(a) ) were observed in the fracture surface. This cracking is thought to give rise to an increase in fracture toughness (288). The macroscopic fracture surface viewed with the naked eye appears similar to that obtained by Beachem and Yoder (295) and Yoder (296) in that periodic ridges and valleys (i.e. the fracture has a zig-zag appearance) are observed (see Fig.6.13). A rationale for this fracture feature in terms of a simple tension-compression model within a constrained plastic zone is postulated in Reference (295). Essentially, as the crack propagates along a shear plane, it does not continue indefinitely because of the tension and compression system that is set up at the crack tip. The crack therefore changes its direction to travel along a plane in the more favourable tensile region giving rise to a characteristic zig-zagging topography of ridges and valleys.

From the discussion above, it can be seen that the differences in fractographic features, as well as the values of fracture toughness for the three specimen orientations were the direct consequence of the non-uniform distribution of inclusions and the anisotropy in their shape. The R - L or short-transverse fracture plane orientation suffers the most as a result of the alignment of the elongated MnS inclusions in a direction parallel to the crack propagation. The effect of anisotropy on fracture toughness is expected to be minimized in materials that are not rolled. This is revealed in the data obtained from the As-Cast 7075-T6 aluminium alloy. As shown in Table 6.4, the fracture toughness values in the S - T and L - S orientations are very similar (29.60 and 29.67 respectively). Even in the T - L orientation the value of $K_{IC}$ ranges from 24.2 to 30.8 (see Appendix K) giving an average result of 27.67 MPa/m. Comparing with the data for 7075-T651 alloy given by Hertzberg (100) where in the L - T, T - L and S - L orientations, the fracture toughness values are respectively 27 to 30,
25 to 28 and 16 to 21 MPa/m, it can be concluded that within experimental error, no orientation effect can be detected in the as-cast material. This is confirmed by the fractographic observations of the fracture surfaces of the specimens (Figs. 6.47(a), (b), 6.49(a), (b) and 6.50(c), (d)). The fracture features of all three orientations tested were nearly identical.

7.3.4 Effect of Alloy Elements

The slight difference in steel compositions, as tabulated in Table 7.1, was the result of different bar stocks being obtained from the manufacturer at different times. The major variation in composition occurred in the silicon, nickel, sulphur and phosphorus contents. In the steel with composition N, the silicon and nickel contents were higher than those in the composition O while the other two elements were lower. The silicon content in composition N was higher than the generally allowed range of 0.10 - 0.35 for this material, while the nickel content in composition O was too low (2.30 - 2.80)\(^{(297)}\).

The variation of fracture toughness with tempering temperature for both steel compositions is shown in Fig. 6.5. For comparison purposes, only the data for the 25mm thickness specimens were plotted. It can be observed that the curve for composition N is generally slightly higher than that for composition O over the entire range of tempering temperatures investigated. The increase in toughness as a result of alloying effect can be considered as follows. It has been observed that the addition of silicon tends to improve fracture toughness slightly at tempering temperatures below 200\(^\circ\)C, but at higher temperatures, the toughness is greatly reduced\(^{(77)}\). On the other hand, the increase in sulphur and phosphorus contents, as shown by the results of Cottrell\(^{(74)}\) and Birkle et al.\(^{(73)}\), lowers the value of fracture toughness considerably at all tempering temperatures. The beneficial role of nickel in improving toughness has long been recognised. All these effects therefore result in a higher fracture toughness in the steel with composition N at least at low tempering temperatures. As the temperature increases, the increase in fracture toughness due to lower sulphur and phosphorus contents increases\(^{(74,73)}\) even though at this temperature range the higher silicon content gives rise to deleterious effect on the toughness. The net effect as observed in Fig. 6.5, is further increased in fracture toughness value in the steel with composition N. The increase in toughness at 600 and 650\(^\circ\)C for composition N steel may also have resulted from the differences in heat treatment processes,
as mentioned in section 7.2. However, it is not possible to separate the latter heat treatment effect from the former alloying effect.

### 7.3.5 Errors

Systematic errors due to uncertainties in measurements, as shown in section 6.4.2, are of the order of 2%, arose chiefly from the determination of the load in the fracture toughness calculation. Examination of Appendices J and K however reveals that even for the specimens having the same thickness and same heat treatment, a variation of larger than 2% in the fracture toughness values was observed. In certain cases, the three specimens which were given identical heat treatment at the same time gave a variation in fracture toughness of about 3.5%. The variation under such circumstances must be partly due to the inhomogeneities within the steel, particularly the distribution of the MnS stringers at the immediate crack tip. The other important factor that contributed to the variation may be the slight variation in tempering temperature due to inadequate control over the heat treatment process. The latter was expected to have a larger effect on the specimens tempered in the high temperature range. This factor was magnified when specimens were heat treated at different times throughout the investigation. The third set of specimens tempered at 300° and 500°C, that is, specimens 3A5...3A3...etc., for example were heat treated at different time from the others. It can be seen that these specimens showed a slightly higher fracture toughness value. Likewise, the scatter of results for composition O steel at tempering temperatures above 450°C as shown in Fig.6.5 was caused by the 450°C and 550°C tempered specimens being heat treated at a different time from the rest of the curve. Hence, the variation in fracture toughness values was attributed mainly to the slight variation in the heat treatment procedure rather than to errors arising from the fracture toughness testing and analysis. The discrepancy, however, did not affect significantly the 0.2% proof stress of the material. A variation of only about 2% was observed in specimens tempered at different times.
The correlation of fracture toughness with the size of the shear lip is not new. As early as 1961, Krafft, Sullivan and Boyle\(^{(47)}\) postulated that the shear lip is a measure of the size of the plastic zone formed at the free surfaces of a plate specimen and in turn is related to the characteristic dimension \((K/\sigma_y)^2\). Comparing for example Fig. 6.5 with Fig. 6.6 in the present study, the variations of shear lip size and fracture toughness with tempering temperature are very similar. As discussed in detail in section 2.6, the shear lip size/fracture toughness relationship may be expressed in the form

\[
B_{\text{SL}} = C(K/\sigma_y)^2
\]

where \(C\) is a constant.

Before discussing the experimental verification of equation 7.2, it may be worthwhile to restate here the findings of the present investigation in relation to the size of the shear lips. It was found that,

a) the size of the shear lips once fully developed remained approximately constant along the free surfaces of the specimen,

b) the size of the shear lips in a mixed mode fracture, that is, beyond the critical thickness \(B_0\), at a specific tempering temperature was independent of the thickness of the specimen (see Fig. 6.9).

Fig. 7.12 shows a linear plot of shear lip size versus the relative toughness, \(K_c/\sigma_y\), at constant specimen thickness for Comsteel En25 over the entire range of tempering temperatures studied. It can be seen that for a particular thickness, the \(B_{\text{SL}}\) versus \(K_c/\sigma_y\) relationship initially follows a similar trend to the theoretical plastic zone size estimations of Rice\(^{(158)}\) and Irwin\(^{(19)}\). At large specimen thicknesses, the experimental curves tend towards Rice's result while small thicknesses tend towards Irwin's. This implies that the value of \(C\) in equation 7.2 is specimen size dependent, larger in the 25mm thickness A specimens and smaller in the 4mm H specimens. This however cannot be true since \(C\) is a constant dependent only on whether the state of stress is plane strain or plane stress. Indeed, a closer examination of equation 7.2 reveals that for a given tempering temperature, since \(C\) is a constant, \(\sigma_y\) is a material property and \(B_{\text{SL}}\) remains invariant.
with specimen size, the parameter $K$ must necessarily be constant with respect to specimen thickness. It can therefore be concluded that equation 7.2 is valid only if the parameter $K$ has a value independent of thickness. Such a value is $K_{IC}^*$ which is a material constant in the same sense as the yield stress, $\sigma_{YS}$. This conclusion is substantiated by the work of Krafft and Sullivan (97) who observed that the estimation of the plastic zone size, $r_y$, based on the plane strain fracture toughness, $K_{IC}^*$, appeared to match the experimental shear lip size better than the estimation based on the plane stress value, $K_c$. Hence, equation 7.2 can be rewritten as

$$B_{SL} = C \left( \frac{K_{IC}^*}{\sigma_{YS}} \right)^2 \hspace{1cm} 7.3$$

**Fig. 7.12** Relationship between $B_{SL}$ and $(K_c/\sigma_{YS})$. 
Equation 7.3 is plotted on linear scales in Fig. 6.14. The $K_{IC}$ value at a given tempering temperature was the average of all the $K_Q$ values that were valid according to the ASTM Standard E399-74 (118). Also plotted in Fig. 6.14 are the theoretical estimations of the plastic zone size due to Rice (158) and Irwin (19). At tempering temperatures below $450^\circ C$, the experimental shear lip size follows closely the plastic zone size predicted by Rice's estimation. However, at higher tempering temperatures (see Fig. 7.12), the experimental data deviates from the theoretical analysis. The deviation increases with the increasing tempering temperature such that the increase in shear lip size does not match the rapid increase in the $(K_{IC}/\sigma_{YS})$ ratio.

To evaluate the value of the constant C in equation 7.3, the data in Fig. 6.14 are replotted in log-log scales in Fig. 6.15 for both the 7075-T6 aluminium alloy and the two compositions of Comsteel En25. Once again, it is observed that the data follow approximately a linear relationship at tempering temperatures below about $450^\circ C$, while at high tempering temperatures, deviation from linearity is evident. Using regression analysis, it was shown that the $B_{SL}/K_{IC}$ relationship below $450^\circ C$ temper be expressed by

$$B_{SL} = 0.41 (K_{IC}/\sigma_{YS})^{2.02}$$

Referring to Table 3.2, it can be seen that the experimental result compares favourably with the theoretical estimations of the maximum extent of the plane stress plastic zone, particularly that due to Rice (158), namely, $\gamma_Y = 0.40(K/\sigma_{YS})^2$. This agreement indicates that the size of the shear lip is not only a representation of the plane stress plastic zone at the free surfaces but is also a measure of its size.

For tempering temperatures higher than $450^\circ C$, Fig. 6.15 shows that the experimental results do not conform to the linear relationship of equation 6.2, but deviate from it in such a way that the deviation appears to increase with tempering temperature. Vosikovsky (298) also observed a similar phenomenon when attempting to map the extent of through-thickness notch contraction at the vicinity of the crack tip in plates of different thicknesses. His results showed that the extent of the notch contraction (plastic zone) in the direction of crack propagation deviated from the theoretical analyses at a stress level of $\sigma_N/\sigma_{YS} \approx 0.7$ where $\sigma_N$ was the net section stress.
7.4.1 Validity of the fracture toughness data

Consider Fig. 7.7 where comparison between the experimental data of the present work and those of Clark (255), Logan and Crossland (273) and Ferguson and Sargisson (274) is made. It was observed that the trend of the variation of fracture toughness with tempering temperature correlated well with the results of Clark and Logan et al. except at about 450°C where a cross over occurred and the present results became consistently higher. The composition of the steels used, as shown in Table 7.1, varied only very slightly and the variation in composition was not expected to have any significant effect on the fracture toughness values. However, with both Logan et al. and Clark's compact tension specimens being tested in the C - R and R - C orientation, it was not surprising that the \( K_{IC} \) was as a whole higher than the present results obtained in the R - L or C - L orientation (see section 7.3.3). The present \( K_{IC} \) data in the R - L orientation should therefore be lower at all tempering temperatures and not just for tempers below 450°C. The thickness of the specimens used by Clark and in the present work was 25mm while 31.7mm was used by Logan et al. As specimens of larger thickness give better estimations of true \( K_{IC} \) values, it is to be expected that the present \( K_{IC} \)/tempering temperature curve should be below that of Logan et al. The implication of this consideration is that the fracture toughness values at tempering temperatures above about 450°C have been overestimated perhaps as a consequence of insufficient specimen thickness or crack length.

In section 2.7, it was stated that in order for the value of \( K_Q \) to qualify as a valid \( K_{IC} \), the ASTM Standard E399-74 (118) requires the specimen thickness and crack length to be greater than 2.5 \( (K_Q/\sigma_{YS})^2 \) and the ratio of \( P_{\text{max}}/P_Q \) to be less than 1.10. While these restrictions are generally applicable to most materials, recent data have shown that in some materials of relatively high toughness they are inadequate to assure a \( K_{IC} \) value that is independent of specimen thickness (124 - 128). The thickness requirement, plotted as a short vertical line for each temper in Fig. 6.7 did not intersect on the flat portions of the \( K_C \) versus B curves. Moreover, the thickness requirement could not be used to discriminate valid from the invalid \( K_Q \) values as the values were found to vary only slightly with the change in specimen thickness (Refer to Appendices J and K).

Generally in fracture toughness testing, a relatively good indication of resistance to fracture is the net section stress, \( \sigma_N \) at failure. Gerberich (173, 174) has shown that when the ratio of net section stress to 0.2%
proof stress, \( \sigma_{N}/\sigma_{YS} \), exceeds a certain value, the plastic energy release rate becomes excessive and the plane strain fracture toughness values so obtained are probably quantitatively incorrect. The necessity of keeping \( \sigma_{N}/\sigma_{YS} \) below a certain level in fracture testing has long been recognised (171 - 175). In the past, a \( \sigma_{N}/\sigma_{YS} \) ratio of 0.8 had been suggested but recent studies have demonstrated the importance of keeping the ratio to below two thirds (177) to obtain size independent fracture toughness values even within the general plane strain region. On the other hand, Heald et al. (183) have commented that the use of linear elastic fracture mechanics, on which the standard fracture toughness testing analyses are based, is only useful at stresses up to approximately one half the general yield stress. Although the exact \( \sigma_{N}/\sigma_{YS} \) ratio is controversial, a value of 0.67 appears to be recommended by Broek (15) and confirmed analytically by Feddersen (176).

The net section stress for compact tension specimens has been formulated (see Appendix 0) to be (299,300),

\[
\sigma_{N} = \left[ 1 + \frac{3(W + a)}{(W - a)^2} \right] \frac{P_{\text{max}}}{b(W - a)} \quad \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdots \cdOTS
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TABLE 7.5: Variation of $\eta / \eta_0$ with tempering temperatures and specimen size.
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**TABLE 7.6:** Variation of $P_{\text{max}}/P_q$ with tempering temperature and specimen size.
Table 7.7 shows the comparison between a and the parameter $5(K_{IC}/\sigma_{ys})^2$.

The shaded area covers those specimens with a $<5(K_{IC}/\sigma_{ys})^2$. Comparing with Table 7.5, the shaded areas are found to envelop identical specimens. Hence, to obtain a valid $K_{IC}$ value at tempering temperatures higher than 450°C, it is necessary that the crack length of the compact tension specimen be longer than $5(K_{IC}/\sigma_{ys})^2$.

The purpose of the various restrictions in the standard plane strain fracture toughness tests (118,119) is to avoid complications due to plane stress and to restrict the plastic zone size. To obtain valid $K_{IC}$ values of a material, a state of plane strain must be maintained across the specimen. However, it is difficult to predict when the effect of plane stress becomes significant. Andersson (301) has experimentally shown that when the value of $P_{max}/W_0$ is less than 0.03, the state of stress is in plane strain but failure occurs in plane stress conditions when $P_{max}/W_0 > 0.03$. Some support for the validity of such a criterion has been demonstrated by Chell and Spink (302). In an effort to assess the significance of this restriction, the parameter $P_{max}/W_0$ for Comsteel En25 with Composition O is tabulated in Table 7.8. The criterion appears to be more conservative than the $\sigma_N/\sigma_{ys} < 0.66$ restriction in that the effect of plane stress becomes influential at tempering temperatures of 450°C and above (compare Table 7.8 with Table 7.5). Although the factor 0.03 may not be precise (303), the parameter serves as an indication that the values of $K_{IC}'$, obtained at tempering temperatures above 450°C cannot be qualified as true plane strain fracture toughness, $K_{IC}$ values.

Before attempting to determine the true $K_{IC}$ values at tempering temperatures above 450°C using yielding fracture mechanics approaches, it may be worthwhile to reconsider Fig.6.15. Clark (255) has shown that a plot of relative toughness, $K_{IC}/\sigma_{ys}$, against $\sigma_{ys}$ provided a linear relationship for Comsteel En25. This observation may be used as a preliminary means to investigate the validity of equation 6.2 for tempering temperatures above 450°C. The plots of $K_{IC}/\sigma_{ys}$ versus $\sigma_{ys}$ for both composition O and N steels are shown in Fig.6.16. Although the relationship is linear, a change in gradient is observed at tempering temperatures above about 400°C. This is expected since the fracture toughness at these tempers were not true $K_{IC}$ values but had been overestimated due to yielding effect. The true $K_{IC}$ values may be extrapolated from the valid data obtained at temperatures below 450°C, as shown in Fig.7.13 by extending the linear relationship from the low to the high temperature range. If the relative toughness values so obtained are then plotted against the size of the shear lips as
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**TABLE 7.7:** Comparison between $a$ and $5(K/\sigma_{ys})^2$. 
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**TABLE 7.8:** Comparison between $P_{\text{max}}/B_{\text{w0}}$ and 0.03.
**Fig. 7.13** Extrapolation of $K_{IC}/\sigma_{YS}$ values from tensile data.

**Fig. 7.14** Relationship between shear lip size and fracture toughness using extrapolated data.
shown in Fig. 7.14, it can be seen that the data fall on the $\frac{K_{IC}}{\sigma_{ys}}$ line (of equation 6.2) within ±6%. This indicates that provided true plane strain fracture toughness data are used, the relationship between $\frac{K_{IC}}{\sigma_{ys}}$ and the size of the shear lip, $B_{SL}$, expressed as

$$B_{SL} = 0.41 \left(\frac{K_{IC}}{\sigma_{ys}}\right)^{2.02},$$

is valid even at tempering temperatures beyond 450°C.
7.4.2 Yielding Fracture Mechanics Data

Owing to the compatibility between the $K_{IC}$ parameter and the various yielding fracture mechanics parameters, an attempt was made to obtain a valid $K_{IC}$ value from the load/displacement records using yielding fracture mechanics approaches. The values of fracture toughness for 25mm thickness, composition O steel, computed using equation 4.6, $K^E$ \textsuperscript{COD}, equation 4.8, $K^T$ \textsuperscript{COD}, equation 4.15, $K_J$, and equation 4.19, $K_{EE}$ are tabulated in Table 6.9 and presented graphically in Fig.6.22.

Consider the crack opening displacement approach. Irrespective of whether equation 4.6 or 4.8 was used, a value of $\lambda = 1$ in equation 4.5 was employed to convert $\delta$ to the fracture toughness $K$ values. $\lambda$ is theoretically taken as 1 in the plane stress conditions and 2 in the plane strain conditions. As such, taking $\lambda = 1$ in the calculations was not strictly accurate especially in the low tempering temperature range where the fracture was one dominated with plane strain square fracture. However, it has been shown\textsuperscript{200} that the theoretical results of $\lambda$ do not agree with those evaluated from experimental methods which favour $\lambda = 1$ in the correlation between $K$ and $\delta$.

As shown in Fig.6.22, $K^E$ \textsuperscript{COD} grossly overestimated the values of $K_c$ and $K_{IC}$ over the entire range of tempering temperatures investigated. The discrepancy can be attributed to three factors: the use of a constant rotational factor, $r$, evidence of crack extension before maximum load was attained, and the rather large contribution from the elastic component of the load/displacement record (see Fig.6.8) to the value of $V_g$. Elliott et al\textsuperscript{(215)} suggested that one of the limitations of the use of the crack opening displacement approach occurs at $V_g$ values below about 0.5mm or correspondingly, $\delta$ values below about 0.09mm. Below this lower limit, the rotational factor has been found experimentally to vary with $V_g$. Using $K_{IC}$ determined from the ASTM analysis, $\delta$ was calculated to have a value in the range 0.0074mm in the as-quenched state to 0.043mm at a 600°C tempering temperature. The use of equation 4.6 to determine $\delta$ from $V_g$ is only justified in cases where fracture initiates well after general yielding\textsuperscript{(233)}. Even in the 600°C tempered specimens, crack extension was found (see section 7.4.4) to occur before the attainment of the maximum load. Robinson and Tuck\textsuperscript{(304)} reasoned that when the elastic portion of the load/displacement record was large compared with the plastic portion, the determination of $\delta$ should be based on the plastic
component only. The analysis of Wells (210), equation 4.8, appears to take into account the contribution due to the elastic component through the parameter \( \gamma \). This explains why \( K_{\text{COD}}^T \) has a slightly better correlation with \( K_c \) and \( K_{IC} \) as shown in Fig.6.22. In addition, equation 4.8 is generally used when there is evidence of crack extension before general yield (293). Since the C.O.D. method is basically a yielding fracture mechanics analysis, it is expected to apply better at high tempering temperatures where the specimens approached a state of general yielding. This is reflected in Fig.6.22.

The overestimation of \( K_{IC} \) using the crack opening displacement approach is substantiated by Liebowitz et al. (305) using compact tension specimens for several aluminium alloys. Because of the general lack of agreement with \( K_{IC} \), the uncertainty of the rotational factor and of the position where \( V_g \) was to be measured, it was concluded that \( K_{\text{COD}}^T \) and \( K_e \) could not be used to accurately determined the valid value of fracture toughness at high tempering temperatures. Thus the more unambiguous parameter, \( \delta_i \), was measured. This determination will be discussed in sections 7.4.3 and 7.4.4.

It has been recognised that the width of the stretch zone can be related to the crack tip opening displacement. The stretch zone is a region of local strain at the crack tip which is formed on loading prior to incipient cracking as a result of crack tip blunting (306 - 309). Many attempts have been made to correlate the stretch zone width, \( t_{sz} \), with \( \delta \). Spitzig (310), Brothers et al. (307) and Pelloux (311) have reported \( t_{sz} \) to be nearly equal to \( \delta \) whereas other investigators suggested different relationships (312 - 315). Recent experimental data (316) obtained using silicone rubber castings in notched HY100 and HY130 steel specimens have shown that the stretch zone width is equal to half the value of \( \delta \). This is very similar to the findings of Harris et al. (317) and the simple model of Green et al. (318).

\( t_{sz} \) was measured in the present work by means of the scanning electron microscope operating at a magnification of 600 to 2000 times. The width was generally uniform although there was sometimes irregularity resulting from MnS inclusions. Measurements were made at locations where the zones were distinct and definite, and consequently no stretch zone size could be measured at tempering temperatures below 300°C. At the low
temperature range, the stretch zones were usually very small and less well defined. However, it is believed that stretch zones still occurred in the specimens (see for example, Fig.6.29(d)). The values of $t_{sz}$, as shown in Table 6.7, varied from about 6μm in the 400°C temper to about 30μm at 650°C.

Assuming the relationship between $t_{sz}$ and $\delta$ discussed above, namely, $t_{sz} = \frac{1}{2} \delta$, the values of the fracture toughness, $K_{sz}'$, determined with equation 4.5 and $\lambda = 1$ are tabulated in Table 7.9. $K_{sz}$ values were slightly lower than the ASTM $K_{IC}$ values. However, as the tempering temperature decreases, it can be seen that $K_{sz}$ approaches the value of $K_{IC}$. The difference between $K_{IC}$ and $K_{sz}$ at 600°C temper was about 17% while at 400°C, it has steadily dropped to only about 6%. Green, Smith and Knott (318) have suggested the use of stretch zones to denote crack tip ductility as an alternative method for estimating the crack opening displacement at initiation, $\delta_i$, from a fractured specimen.

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Table 7.9: Comparison between $K_{IC}$ and $K_{SZ}$

As such, it would be expected that $K_{SZ}$ values are very close to the true value of $K_{IC}$ at the high tempering temperature range. Since the overestimation of the true $K_{IC}$ values increases with temperature, the difference between $K_{IC}$ and $K_{SZ}$ would necessarily increase with temperature. Compare with Logan and Crossland's (273) results in Fig.7.7, it is noted that $K_{SZ}$ values are lower indicating perhaps the trend of true $K_{IC}$.
The fracture toughness computed using the equivalent energy approach, $K_{EE}$, although it predicts accurately the values of $K_C$ over the whole range of tempering temperatures, consistently overestimates the $K_{IC}$ values, as shown in Fig.6.22. This is surprising since it contradicts Witt's idea that it is a lower bound [247]. However, this inconsistency has also been observed by Robinson and Tetelman [251], Chell [186] and Chell and Worthington [246]. Although good results have been reported using this analysis, the validity of the equivalent energy method is as yet unverified and in the case of the present study, the poor estimation of $K_{IC}$ indicates that the adequacy of the method to yield accurately the true $K_{IC}$ in the high tempering temperature range is dubious.

The dependence of the fracture toughness values determined at maximum load using the J-integral approach upon tempering temperature follows a nearly identical trend with that of ASTM $K_{IC}$ values. However, an underestimation of about 15% was observed at all tempering temperatures (Fig.6.22). The lack of agreement may be due to the slow crack growth that occurred before the maximum load was attained, especially in specimens tempered at high tempering temperatures. As a result of the crack growth, unloading of the material behind the crack tip occurs. This invalidates the path independent characteristic of the J-integral. Therefore, the criterion must necessarily be restricted to crack initiation. This is the $J_{IC}$ approach which is discussed in section 7.4.4. Similar discrepancy has been observed by Robinson and Tetelman [251] using the estimation formula in equation 4.15. By using compliance measurements and the corrected crack initiation J values instead of the values at maximum load, they have found $K_J$ to be in good agreement with the ASTM $K_{IC}$ values.

In summary, it has been observed above that the fracture toughness values determined at maximum load using the various yielding fracture mechanics approaches did not generally correlate favourably with the ASTM $K_{IC}$ values over the entire range of tempering temperatures studied. The only promising approach for the determination of true $K_{IC}$ values at high tempering temperatures appears to be the stretch zone width method. However, further confirmation of $K_{SZ}$ is required. As the above criteria were usually invalidated by the occurrence of slow crack growth before the onset of fracture, fracture toughness at crack initiation must be determined in order to obtain true $K_{IC}$ values at tempering temperatures above 450°C. In the following section, the crack opening displacement at initiation, $\delta_i$, obtained using side-grooved compact tension specimens is discussed.
7.4.3 Side-Grooved Specimen Data

It is reported in section 2.3 that the onset of maximum load is generally associated with the deformation, necking and finally fracture of the ligaments (shear lips) at the surfaces of the specimen. This occurs because the plane stress fracture of the ligaments required a higher energy than the plane strain fracture at the centre. If side-grooves of sufficient depth are introduced, the effects of the ligaments and therefore the relaxation of the through-the-thickness stress will be prevented. Under such circumstances, plane strain fracture will occupy the full thickness of the specimen immediately after the attainment of \( \delta_i \), and maximum load instability will be coincident with the initiation of fracture such that \( \delta \) at maximum load is equal to \( \delta \) at initiation.

In the present study, side-grooves were introduced into three specimens that were found to contain quench cracks. It was thought that a groove depth of 2 to 3mm would be sufficient to eliminate the quench cracks since they had been generally found to occur only at shallow depth and along the surfaces of the specimen in the direction of the machined notch. Unfortunately, two specimens were rendered useless because of too severe quench cracks. The results of the successful specimen are presented in Table 6.8. As shown in Fig.6.21, the load/displacement records for the grooved and non-grooved specimens are different in that the former resembles, refer to Fig.2.6, a type C test record while the latter, type A. Determining the fracture toughness using the crack opening displacement and J integral approaches, \( K_{COD} \) and \( J \) were found to be 81.87 and 80.55 MPa\(\sqrt{m} \) respectively. The two values are very close and in good agreement with \( K_{SZ} \) of 78.53 MPa\(\sqrt{m} \).

Although in side-grooved specimens \( \delta \) at maximum load has been reported \((220,279)\) to coincide with \( \delta \) at crack initiation, it is not clear if this is indeed so for the present material.

7.4.4 \( \delta_i \) and \( J_{IC} \) Data

The recommended procedure for determining \( \delta_i \) is the extrapolation ("R" curve) technique \((293)\) which involves marking the amount of fibrous crack extension after the specimen has been unloaded from a certain displacement along the load/displacement curve. The procedure has the advantage that both the crack opening displacement at crack initiation, \( \delta_i \), as well as the elastic-plastic fracture toughness, \( J_{IC} \), can be obtained from the same set of specimens.
Displacement, mm.

Fig. 7.15 Load/displacement record showing positions of unloading.

The values of $\delta_1$ and $J_{IC}$ were determined only on specimens tempered at 600°C. The positions from which the specimens were unloaded are shown in Fig. 7.15. The maximum load occurred between specimens 11A6N and 8A6N (refer to Appendix M), or points 3 and 4 in Fig. 7.15. It was observed that before maximum load was reached, slow crack growth had occurred. Points 1 and 2 had crack extension of 0.10 and 0.35 mm respectively. The ASTM 5% secant line intersects the load/displacement curve at a point between 2 and 3. This is not surprising since a 2% increment of crack extension is used in the ASTM standard (118).
The variation of $\delta$ and $J$ with crack extension, $\Delta a$, is shown in Fig. 6.19 and 6.20 respectively. The linear relationship in Fig. 6.20 is similar to those obtained by Logsden (319) and Clarke et al. (320). The value of $\delta_1$ was found to be 0.039mm which corresponded to a true plane strain fracture toughness, $K_{IC}$, value of 84.64 MPa/m. $J_{IC}$ was found to be 28.5 Nmm/mm$^2$. This gave a value for the true $K_{IC}$ at 600°C temper of 80.13 MPa/m. Table 7.10 shows the comparison between the values of $K_{IC}$ obtained in this section and those obtained using side-grooved specimen and the stretch zone width.

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<th>Experimental Procedure</th>
<th>$K_{IC}$, MPa/m</th>
</tr>
</thead>
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<tr>
<td>Stretch zone width measurement</td>
<td>$K_{SZ}$</td>
</tr>
<tr>
<td>Side-grooved specimen: C.O.D.</td>
<td>$K_8$</td>
</tr>
<tr>
<td>J-Integral</td>
<td>$K_J$</td>
</tr>
<tr>
<td>R-curve procedure: C.O.D.</td>
<td>$K_{8i}$</td>
</tr>
<tr>
<td>J-Integral</td>
<td>$K_{JIC}$</td>
</tr>
</tbody>
</table>

Table 7.10: $K_{IC}$ of Comsteel En25 at 600°C temper using different experimental procedures.

The agreement in the $K_{IC}$ values obtained is obvious. The slightly larger values in the C.O.D. analyses may be due to the factor $\lambda$ whose value was uncertain. Robinson and Tetelman (200) obtained an experimental value of $\lambda$ equal to 0.97. In the present study, $\lambda$ was taken as unity.

From Table 7.10 it can be concluded that the true plane strain fracture toughness, $K_{IC}$, for Comsteel En25 tempered at 600°C is 80.13 MPa/m (taking $K_{JIC}$ of R-curve analysis only).

Now reconsider the relationship between fracture toughness and the size of the shear lips as depicted in Fig. 6.15. The data below 450°C are replotted in Fig. 7.16 with the 600°C data included. It can be seen that the 600°C data, instead of deviating from the linear relationship as in Fig. 6.15, now lies within a ± 6% scatter band of the relationship showing conclusively that the plane strain fracture toughness is related to the size of the shear lips through equation 6.2 even when excess yielding is present.
Fig. 7.16 Relationship between fracture toughness and shear lip size (with $J_{IC}$ value included).

Thus the apparent deviation from the straight line behaviour for tempering temperatures above 450°C (see Fig. 6.15) was due to the ASTM $K_Q$ values being classed as true $K_{IC}$ values. In this respect, if true $K_{IC}$ values are to be determined by the ASTM testing method, the standard will have to be restricted in terms of the allowable $\sigma_p/\sigma_s$ ratio or some similar requirement.

7.4.5 Below Ambient Temperature Testing Data

To extend the range of applicability of the plane strain fracture toughness/shear lip size relationship, attempt was made to decrease the size of the shear lips by lowering the testing temperature. This was done using 25mm thickness specimens tempered at 600°C. The results of such tests are
Fig. 7.17 Overload fracture morphology of 7075-T6 aluminium specimen tested at -196°C.
presented in Table 6.5. In spite of the large temperature drop, from 0 to -196°C, the values of $K_I$ remained virtually the same. For the Comsteel En25 specimens, this is contradictory to the findings of Ferguson and Sargisson (274). Using cracked-notched round bars of the same material tempered at 350°C, they demonstrated, like for most high strength steels (321,322), that the fracture toughness decreased with the decrease in testing temperature. The drop in toughness is generally associated with the change in microscopic fracture behaviour, from ductile dimples to quasi-cleavage or 100% cleavage (323). Ferguson et al. (274) found that the fracture surface changed from a mixture of tear dimples and intergranular cleavage at room temperature to a predominantly quasi-cleavage at liquid nitrogen temperature (256). However, in the present study, scanning electron microscopic observation of the fracture surfaces did not reveal any significant change in the fracture mechanism, ductile dimples remained from room temperature down to -196°C. The initial high toughness of the 600°C tempered material did not seem to be reduced by the decrease in testing temperature. Consequently, the fracture toughness remained approximately constant.

For 7075 aluminium alloy in the T651 and T7351 conditions, Nelson and Kaufman (275) found a slight but steady increase in fracture toughness as the testing temperature was lowered from room temperature to -196°C (or -320°F). This increase was also observed in the present investigation with 7075-T6 alloy from 0°C to -78°C. However, a significant drop in $K_{IC}$ was observed at -196°C. Macro-fracture and electron micro-fractographic observations showed that the fracture appearance of the 0°C and -78°C specimens was quite different from that at -196°C. While the former specimen exhibited a ductile fracture mode similar to that in the as-cast material at room temperature (discussed in section 6.10.2c), the latter specimen (as shown in Fig. 7.17) was observed to have been rolled rather than cast. This could have happened as the material was supplied in machined blocks of size 70mm x 160mm x 160mm for which the history was unknown. From the fibre direction observed on the fracture surface, the crack plane was found to be in either the T-L or S-L orientation. As noted in section 7.3.3, this orientation offers the least resistance to fracture and therefore, the lower fracture toughness value obtained at -196°C was to be expected.

The variation of plane strain fracture toughness obtained at testing temperatures below ambient with the size of the shear lips is shown in Fig. 7.18.
along with the previous data. It can be seen that the aluminium alloy data follow very closely the previous trend expressed by equation 6.2.

\[ B_{SL} = 0.41 \left( \frac{K_{IC}}{\sigma_{YS}} \right)^{2.02} \]

Fig. 7.18 Relationship between fracture toughness and shear lip size (Below room temperature testing data).

The data for the Comsteel En25 specimens were not shown in Fig. 7.17 because they were not regarded as valid plane strain fracture toughness values by the same criteria used in section 7.4.1. It was found that,

a) the net section stress to 0.2% proof stress ratio for specimen 14A6N and 15A6N was 0.81 and 0.77 respectively, both larger than two thirds,
b) \( \frac{P_{\text{max}}}{BW} \) for 14A6N and 15A6N was 0.038 and 0.035 respectively which exceeded 0.030 for the effect of plane stress to be negligible.

Although the range in which the data fall in Fig. 7.18 was by no means extended by lowering the testing temperature, the data do support the previous conclusion that the plane strain fracture toughness can be related to the size of the shear lips through equation 6.2.

7.4.6 Application

For the determination of the plane strain fracture toughness, \( K_{IC} \), the ASTM Standard [118] and the British Standard [119] impose severe restrictions on the conditions of testing to ensure that the amount of plasticity occurring in the test is small in comparison to the dimensions of the specimen. As a result, the presently accepted testing methods for \( K_{IC} \) are difficult, time consuming and expensive for general use. Several parameters have been proposed to overcome the problem. One of these involves the measurement of the stretch zone width [307, 315] which can be linearly correlated to the relative toughness of the material. The fracture toughness can thus be determined by measuring the stretch zone width at the interface between the fatigue crack and the overload fracture. However, the measurement of this region requires the use of the electron microscope. Additionally, it is difficult to decide the position where fatigue crack ends and where the stretch zone begins. These make the method impractical. Measurement of the size of the shear lips at the free surfaces of a specimen on the other hand does not involve methods of great complexity. As the size remains fairly constant along the free surfaces, this quantity can be determined with greater accuracy. The present study has demonstrated conclusively the relationship between the shear lip size, \( B_{SL} \), and the plane strain fracture toughness of the material. It is therefore evident that \( B_{SL} \) is a promising parameter to estimate \( K_{IC} \). The \( K_{IC}/B_{SL} \) relationship accordingly has considerable importance in the analysis of service failures. Service failures often exhibit distinct shear lips which are easily measured and using the relationship a quick estimate of the fracture toughness of the material can be obtained. In addition, the relationship can be used as a preliminary estimation of the fracture toughness value to investigate the specimen thickness requirement for standard fracture test. From the design point of view,
the relationship has the added advantage that it provides directly the conservative plane strain fracture toughness value. Before outlining a procedure by which \( K_{IC} \) can be predicted using shear lip size, it is however important to investigate the applicability of the relationship and to evaluate its limitations if any.

Since the \( K_{IC}/B_{SL} \) relationship expressed in equation 6.2 was evaluated in this work using only Comsteel En25 and 7075-T6 aluminium alloy, it is necessary to examine first of all whether the relationship is applicable to other materials. Available information on the size of the shear lip and the corresponding plane strain fracture toughness is very limited. In Fig.7.19, available shear lip size and reported plane strain fracture toughness data of other investigators \((90,128,255,324)\) is plotted for comparison.
Some scatter in the results is evident but it can be seen that the shear lip size/fracture toughness relationship expressed in equation 6.2 is obeyed reasonably well. The scatter was from the inaccuracy in the shear lip size measurements; in References (90), (128) and (324) the investigators measured the percentage shear lip only as an indication of the change in fracture appearance in which accuracy is not essential. Fig.7.19 indicates that the shear lip size/fracture toughness relationship is not only applicable to Comsteel En25 and 7075-T6 aluminium alloy, but to other materials as well.

With the established relationship between fracture toughness and shear lip size discussed above, it is now possible to predict the plane strain fracture toughness of a material without having to conduct the inconvenient standard recommended procedure. First of all, the average size of the shear lips at the free surfaces has to be measured. This may be obtained, in the case of a preliminary determination of specimen size with a standard test. In service failure analysis, this may be obtained from the fractured component or article. To determine the 0.2% proof stress of the material, it is required to perform a tensile test. Alternatively however, a rough estimation of the flow stress (approximately half way between yield stress and ultimate stress) of the material can be obtained from a hardness test since the hardness of plastically deforming materials, taken as force per unit area is approximately 3.2 times the flow stress$. The plane strain fracture toughness of the material can finally be evaluated using equation 6.2. To illustrate the procedure, a numerical example is given as follows:

Using H-11 steel tempered at $1000^\circ F (538^\circ C)$, Steigerwald$^{(322)}$ found that the plane strain fracture toughness at $200^\circ F (93^\circ C)$ testing temperature was about 50.1 Ksi/in (estimated from Fig.1 of Reference $^{(322)}$), the 0.2% proof stress at the corresponding testing temperature, 231.6 Ksi. At this test temperature, the fracture surface of the 0.085in thickness centre-precracked sheet specimen was observed to contain 50% shear fracture. From the data, therefore,

$$B_{SL} = 0.021 \text{ in (0.54 mm)}$$

$$\sigma = 231.6 \text{ Ksi (1597 MPa)}$$
Using equation 6.2,

\[ B_{SL} = 0.41 \left( \frac{K_{IC}}{\sigma_{YS}} \right)^{2.02} \]

\( K_{IC} \) can be calculated to be 53.2 Ksi/in (57.5 MPa/m). It is noted that the calculated value compares well with the experimental value of 50.1 Ksi/in (54.2 MPa/m).

Equation 6.2 correlates the shear lip size with the plane strain fracture toughness and the 0.2% proof stress. As such, the equation always gives a conservative fracture toughness value in medium thickness specimens, that is, region B in Fig. 2.9 where the amount of shear lips in the mixed mode fracture is compatible to the amount of flat fracture. When the specimen thickness is large and the proportion of shear lips is less than 15%, region C of Fig. 2.9, the fracture toughness calculated from equation 6.2 is equal to the plane strain \( K_{IC} \) of the material.

7.4.7 Possible Limitations of the KIC/BSL Relationship

In section 7.4.6, it was shown that the relationship between plane strain fracture toughness and shear lip size is valid for several materials for which data were obtained using plate specimens, namely, compact tension and 3-point bend specimens. In order to generalize the relationship, it is desirable to examine the situation where data are obtained with Charpy impact and tensile specimens.

Charpy impact data for Comsteel En25 of the following composition (weight percent): 0.27C, 0.30Si, 0.63Mn, 2.63Ni, 0.70Cr, 0.55Mo, 0.021S and 0.023P were obtained by Ferguson, Clark and Watson \(^{(326)}\) using standard Charpy V-notched specimens and fatigue-cracked impact specimens of similar dimensions. The shear lip size data were obtained by Watson \(^{(327)}\). Their results are presented in Fig. 7.20. Although the data points lie in the vicinity of equation 6.2, a considerable amount of scatter in results is also apparent. This scatter may be due to the measurements in shear lip size, the fracture toughness as well as the yield strength of the material. It was mentioned in section 2.5.1 that fracture toughness is influenced by the rate of loading. The variation of \( K_{IC} \) with strain rate is still not well established \(^{(328)}\). For instance, using the same A302-B steel, Ripling and Crosley \(^{(328)}\) found little difference between static and dynamic fracture
toughness values while Bush (328) found that dynamic $K_{IC}$ to be lower than the static results.

\[
BSL = 0.41 \left( \frac{K_{IC}}{\sigma_{ys}} \right)^{2.02}
\]

Fig. 7.20: Relationship between shear lip size and fracture toughness using data from Charpy tests.

Hahn et al. (49, 329) reported a considerable increase in dynamic $K_{IC}$ over the static with SAE 4340 steel. Therefore without performing the dynamic fracture toughness tests, it is not possible to discern the effect of strain rate on $K_{IC}$ and $\sigma_{ys}$ for Comsteel En25. In addition, the formation of the
shear lips is generally suppressed in the impact test. Because of the small specimen size, the shear lips that begin to form at the two free surfaces of the specimen are rapidly influenced by the effect of the free back surface. To investigate the relationship between $K_{IC}$ and $B_{SL}$ in impact testing, it may be necessary to increase the unfractured ligament, $(W - a)$, of the test specimen so that the formation of the shear lip is not affected by the back surface. It is thus concluded that Fig.7.20 is not meaningful and that further investigation is needed to substantiate any correlation between shear lip size and fracture toughness for Charpy impact tests.

Fig.7.21 shows a plot of the shear lip thickness for the tensile specimens which were machined from the fractured compact tension specimens against the corresponding relative toughness. The shear lips were measured by viewing the tensile specimens at 20 times magnification under the Nikon profile projector. Equation 6.2 appears to fit the data below a 400°C tempering temperature. However, in this temperature range, the size of the shear lips was not compatible with those in the corresponding compact tension specimens. Comparing the data with those in Fig.6.15, it can be seen that the shear lips on the tensile specimens are slightly larger than those in the compact tension specimens. No data above a tempering temperature of 400°C was presented in Fig.7.21 because the specimens fractured by full shear in this temperature range rather than by cup and cone fracture. This is to be expected as the diameter of the tensile specimens was about 3.56mm while the size of the shear lips on the compact tension specimens tempered at, say, 450°C averaged 1.49mm (Table 6.6).

It was thought that the shear lip size is influenced by the size of the tensile specimens used. To investigate the size effect, a series of tensile specimens of different diameters in the as-quenched condition, were tested. The specimens, designed in accordance with the British Standard requirements, were machined from 1 in. diameter bar of Comsteel En25. The axis of the specimen was in the direction of rolling. The specimens were austenitized at 850°C in a salt bath for one hour and quenched in an oil bath. The tensile tests were carried out on a 1000 KN capacity Avery testing machine. The results of the tests are summarised in Table 7.11. The hardness measurements have been corrected for round sections.
It can be seen that the change in specimen diameter does not produce a significant change in tensile strength. The slight variation in strength may have been resulted from the lack of alignment device in the testing machine. While the percentage reduction in area decreases with increasing bar diameter, the size of the shear lips on the other hand increases with bar diameter. Similar trend has also been observed by DeSisto et al. (86)
<table>
<thead>
<tr>
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<th>1</th>
<th>2</th>
<th>3</th>
<th>4</th>
<th>5</th>
</tr>
</thead>
<tbody>
<tr>
<td>Diameter, mm</td>
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<td>11.28</td>
<td>13.83</td>
<td>15.98</td>
<td>19.52</td>
</tr>
<tr>
<td>$\sigma_{UTS}$ MPa</td>
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<td>1871</td>
<td>1864</td>
<td>1845</td>
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<tr>
<td>% area reduction</td>
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<td>40</td>
<td>30</td>
<td>32</td>
<td>26</td>
</tr>
<tr>
<td>shear lip size (including thickness contraction), mm</td>
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<td>2.7</td>
<td>3.4</td>
<td>3.7</td>
<td>4.3</td>
</tr>
<tr>
<td>shear lip size (excluding thickness contraction), mm</td>
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<td>1.4</td>
<td>2.3</td>
<td>2.2</td>
<td>2.9</td>
</tr>
<tr>
<td>hardness, Rc</td>
<td>51</td>
<td>50</td>
<td>51</td>
<td>52</td>
<td>51</td>
</tr>
</tbody>
</table>

Table 7.11: Tensile data from tensile specimens of varying diameters
with 7075-T6 aluminium, 6Al-4V-2Sn titanium and 4340 steel. The results appear to be in contrast with the finding of the present work using compact tension specimens, that shear lip size is independent of the specimen size. The cause of such discrepancy is uncertain. It is however, thought that the mechanism of shear lip formation in the compact tension specimens may be different from that in the tensile specimens. Further research is therefore necessary to investigate any correlation between shear lip size and fracture toughness.

It is observed that in general the fracture surfaces of most materials exhibit shear lips at the free surfaces of the specimen or component. However, particularly in glassy materials and in low temperature testings, this is not always the case. Completely flat or square fracture, with no trace of shear lips, is also a common phenomenon. The reasons for the absence of shear lips in such situation is not clear but may be due to the dynamics and speed of the crack propagation, the mode of fracture being cleavage or to the incapability of the material to undergo plane stress yielding. It is believed that the present fracture toughness/shear lip size relationship is not applicable to such materials.

7.5 SUMMARISING DISCUSSION

To briefly summarise the discussion in this chapter, it was observed that the material austenitized at 850°C and oil quenched produced a comparatively low fracture toughness. The low toughness was accompanied by considerable cleavage fracture and rather flat dimples. Tempering at 100°C did not produce a significant change in either the fracture toughness or the tensile properties of the material. At the 200°C temper, the tensile strength was observed to be slightly reduced but significant improvements occurred in the fracture toughness and strain-hardening exponent. Scanning electron microscopic observation showed that the increased fracture toughness was caused by a higher material ductility which was evident from the fracture surface. The dimples formed at this temper were deeper and better formed than those in the as-quenched or 100°C tempered materials. Moreover, only few cleavage areas were noted. Tempering at 300°C produced a marked change in fracture toughness as well as the fracture mechanism. The drop in fracture toughness, strain-hardening exponent and percent elongation caused by the tempered martensitic embrittlement at this tempering temperature was accompanied by a change in
fracture morphology, from dimples to predominantly cleavage. The presence of cleavage fracture continued onto the 400°C tempered materials where the yield and tensile stresses dropped considerably but the fracture toughness increased only slightly. At tempering temperatures above 450°C, further reduction in yield and tensile stresses occurred. The fracture toughness, strain-hardening exponent, reduction in area and elongation were all observed to rise steadily. Fractographically, the fracture surfaces consisted of ductile dimples which as with the previous tempers, the fracture mechanism was shear linkage between major bands of MnS stringers at different elevations forming more marked terrace-type fracture. The fracture toughness of the material was controlled by the major MnS stringer population. In the L - C crack plane orientation where the inclusions were aligned normal to the crack plane, the fracture toughness was found to be two times that in the R - L orientation where the MnS stringers were aligned in the direction of crack propagation. In the 7075-T6 aluminium alloy where the material was in the as-cast condition, no anisotropic effect was observed. The fracture was macroscopically brittle but microscopically ductile.

The relationship between plane strain fracture toughness and shear lip size at tempering temperatures below 450°C was linear and could be expressed using equation 6.2. At temperatures higher than 450°C, the relationship was not linear. The deviation was the result of excessive yielding which was reflected in the increasing amount of through-thickness contraction. The ASTM standard testing procedure was proven to be not sensitive enough to detect the invalidity of the specimens. In this temperature range, it was found that $\frac{\sigma_N}{\sigma_{YS}}$ was greater than two-thirds and that $\frac{P_{\text{max}}}{B W \sigma_{YS}}$ was greater than 0.03. Thus there was excessive plasticity which tended to overestimate the $K_{IC}$ values. To show that equation 6.2 was valid for specimens tempered at temperatures higher than 450°C, other methods of calculating $K_{IC}$ were employed. Except when using stretch zone width measurements, the various methods which used the maximum load were found to be inaccurate because of slow crack growth prior to the attainment of the maximum load. The true $K_{IC}$ at 600°C temper was determined using the value of crack opening displacement at crack initiation obtained from side-grooved compact tension specimen and the "R-curve" method. The resulting true $K_{IC}$ compared favourably with that obtained using stretch zone width and the true $K_{IC}$ at 600°C temper was
further confirmed using the elastic-plastic $J_{IC}$ parameter. When the latter value was plotted on the $B_{SL}$ versus $K_{IC}/\sigma_{YS}$ curve, it was found that the data point fell within the scatter band of equation 6.2 showing that the plane strain fracture toughness/shear lip size relationship was valid even at high tempering temperatures provided that the true $K_{IC}$ value was used. The relationship was additionally found to be valid using data of different materials obtained from the literature. Consequently it can be satisfactorily used in service failure analysis (with the possible restrictions expressed above) as well as in preliminary study of specimen size effect in standard plane strain fracture toughness testings.
CHAPTER 8

CONCLUSIONS
CHAPTER 8

CONCLUSIONS

On the basis of the data obtained in the present investigation, the following conclusions were drawn.

(A) Fracture toughness:

1. The average as-quenched plane strain fracture toughness in the R - L crack plane orientation for Comsteel En25 was found to be 48 MPa/m.

2. The fracture toughness of the steel specimens was considerably affected by the orientation of the crack plane. At 500°C temper, fracture toughness in the L - C orientation was twice that in the R - L orientation.

3. The fracture toughness of the steel could be increased significantly by reducing the total impurity content, especially phosphorus and sulphur, of the material.

4. The fracture mode tested in the R - L and C - R orientation was controlled by the major MnS stringer population. A shear linkage mechanism giving rise to terrace-type fracture was observed in the above orientations. In the C - R orientation, the terrace-type fracture was often found to change to a more zig-zag nature. Beachem and Yoder (295) type of fracture topography was macroscopically observed in the L - C orientation.

5. At tempering temperatures below 450°C, the KIC obtained from the ASTM E399-74 5% secant offset analysis was found to be the true plane strain fracture toughness of the material at that tempering temperature.
6. The values of $K_e$ obtained at tempering temperatures above $450^\circ C$, although passed the ASTM standard requirements, were not the true plane strain fracture toughness due to excessive yielding. It was observed that $\sigma / \sigma_{YS}$ was greater than 0.66 and that $P_{max} / B W_{YS}$ was greater than 0.03.

7. Fracture toughness evaluated at maximum load using the yielding fracture mechanics approaches, namely, the C.O.D., $J$ integral and the equivalent energy concept, did not correlate well with those obtained from the ASTM standard because of the occurrence of fibrous crack growth before the attainment of maximum load.

8. The true plane strain fracture toughness of Comsteel En25 tempered at $600^\circ C$ using the elastic-plastic $J_{IC}$ approach was found to be 80 MPa$m$$. This correlated favourably with those obtained from $\delta_1$ using the "R-curve" method, the side-grooved compact tension specimen, and the stretch zone width.

9. The plane strain fracture toughness of 7075-T6 aluminium alloy in the as-cast condition was found to be 27.7 MPa$m$.

10. The fracture toughness of the aluminium alloy was not significantly affected by the orientation of the crack plane. This is because in the as-cast condition, no rolling operation has been performed on the material to align the inclusions.

11. The fracture toughness, $K_c$, values of both the steel and the aluminium alloy were dependent on the specimen thickness. $K_c$ decreased with the increase in specimen thickness until a limiting value was reached. At low tempering temperatures and with maximum specimen thickness, $K_c = K_{IC}$. 
(B) Shear lips:

1. The size of the shear lips on the free surfaces of the specimens tested in the R - L orientation was constant along the path of crack propagation.

2. The size of the shear lips once fully developed, that is, in a mixed mode fracture beyond the critical specimen thickness where $K_c$ has a maximum value, was independent of the specimen thickness.

3. The dependence of shear lip size upon tempering temperature was nearly identical to that of plane strain fracture toughness upon tempering temperature.

4. Due to the low strain-hardening capacity of the material, the deformation associated with the shear lip formation was confined within a band of a few grains in width on the $45^\circ$ surfaces of the shear lips only. Consequently, microhardness testings could not reveal the extent of the deformation.

(C) Relationship between fracture toughness and shear lip size:

1. The size of the shear lips was found to be better related to the plane strain fracture toughness, $K_{IC}$, rather than $K_c$.

2. The relationship between the plane strain fracture toughness, $K_{IC}$, and the size of the shear lip, $B_{SL}$, was conclusively shown to be

$$B_{SL} = 0.41 \left( \frac{K_{IC}}{\sigma_y} \right)^{2.02}.$$ 

3. The relationship between $K_{IC}$ and $B_{SL}$ was nearly identical to the theoretical prediction of Rice (158) on the size of the plane stress plastic zone using the Von Mises yield criterion. The shear lip size was therefore a good measure of the size of the plastic zone at the surfaces of a plate specimen.
4. The $\frac{K_{IC}}{B_{SL}}$ relationship expressed as

$$B_{SL} = 0.41 \left( \frac{K_{IC}}{\sigma_{YS}} \right)^{2.02}$$

was also found to be applicable to other materials like 4340 steel, maraging steel and 2219-T851 aluminium alloy. It may not be applicable, however, to data obtained from Charpy impact and tensile specimens.
CHAPTER 9

SUGGESTIONS FOR FUTURE INVESTIGATIONS
CHAPTER 9

SUGGESTIONS FOR FUTURE INVESTIGATIONS

1. To determine the relationship between shear lip size and fracture toughness using tensile and Charpy impact specimens.

   It has been suggested that the relationship between shear lip size and fracture toughness found in the present study may be limited to plate specimens only. Additional investigation is necessary to extend the use of the relationship to tensile and Charpy specimens. Comparison between the mechanisms in shear lip formation may be made with specimens of different geometry. In the case of the tensile specimen, analytical approach may be used to relate the amount of necking to the size of the shear lips. To eliminate the effect due to the back surface in Charpy impact test, the unfractured ligament of the specimen may have to be increased.

2. To determine the effect of fracture mechanism on shear lip formation.

   Using a material with well defined transition temperature, the relationship between shear lip size and fracture mechanism may be investigated. The \( K_{IC}/B_{SL} \) relationship may also be examined over the entire range of testing temperature.

3. To determine the effect of back stress on shear lip size.

   Photoelasticity method may be employed to study the effect of compressive stress, due to the free back surface of the compact tension specimen, on the size of the shear lip at various applied stress levels. Does the compressive stress suppress the growth of the shear lip?

4. To determine further appropriate restrictions in the ASTM standard plane strain fracture toughness testing method.

   Using the fracture toughness results obtained in the present study, investigation may be made to propose further restrictions to limit the effect of plasticity in the standard test. The net section stress to yield
stress ratio may have to be kept below two thirds or the crack length of the specimen be increased to $5 \left( \frac{K_{IC}}{\sigma_{YS}} \right)^2$, depending on whichever is appropriate.

5. To determine the effect of major inclusion spacing on short-transverse fracture mechanism.

For specimens obtained in the R-L and C-L orientations, the fracture mechanism was step or terrace type, each crack growth increment being determined by the major MnS stringer population. The effect of increasing the spacing between MnS stringers has on the fracture mechanism may be investigated. Is the fracture topography still terrace in nature? The effect in altering the shape of the MnS stringers, perhaps by adding rare earth elements such as cerium, has on fracture mechanism may also be investigated.

6. To determine the factors that suppress the formation of shear lips in materials that fail by completely flat fracture.

Shear lips are formed as a result of the plane stress stress state that exists at the surfaces of a specimen. The shear lip size is controlled by the plane strain fracture toughness of the material. The existence of plane stress stress state and the values of the fracture toughness in some brittle or glassy materials that fail by completely flat fracture may be investigated using compact tension specimens.
APPENDIX A

Derivation of Crack Tip Stress Field

Consider the stress components acting on an element of dimension $dx \, dy$. In the case of a two-dimensional problem, Fig.A.1 shows such an element with stresses resolved in the $x$-direction.

![Stress components diagram](image)

**Fig.A.1 Stress components acting on a material element.**

The equation of equilibrium for forces in the $x$ direction is

$$
\frac{\partial \sigma_x}{\partial x} + \frac{\partial \tau_{xy}}{\partial y} = 0 \quad \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots A.1(a)
$$

Similarly, in the $y$ direction,

$$
\frac{\partial \sigma_y}{\partial y} + \frac{\partial \tau_{xy}}{\partial x} = 0 \quad \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots A.1(b)
$$

If $U$ and $V$ are the displacements in the $x$ and $y$ direction respectively, the three strain components can be expressed as
\[
\varepsilon_x = \frac{\partial U}{\partial x}
\]
\[
\varepsilon_y = \frac{\partial V}{\partial y}
\]
\[
\gamma_{xy} = \frac{\partial U}{\partial y} + \frac{\partial V}{\partial x}
\]

With appropriate differentiations, equation A.2 can be combined to give

\[
\frac{\partial^2 \varepsilon_x}{\partial y^2} + \frac{\partial^2 \varepsilon_y}{\partial x^2} = \frac{\partial^2 \gamma_{xy}}{\partial x \partial y} \quad \text{A.3}
\]

This differential equation is the condition of compatibility. By using Hooke's law, equation A.3 may be transformed into a relation between the components of stress.

\[
\frac{\partial^2}{\partial y^2} (\sigma_x - \nu \sigma_y) + \frac{\partial^2}{\partial x^2} (\sigma_y - \nu \sigma_x) = 2 (1 + \nu) \frac{\partial^2 \tau_{xy}}{\partial x \partial y}
\]

\[\text{............. A.4}\]

The equilibrium condition expressed in equations A.1 will be automatically satisfied if

\[
\sigma_x = \frac{\partial^2 \psi}{\partial y^2}, \quad \sigma_y = \frac{\partial^2 \psi}{\partial x^2}, \quad \tau_{xy} = -\frac{\partial^2 \psi}{\partial x \partial y} \quad \text{............. A.5}
\]

where the function \( \psi \) is called the Airy stress function.
With appropriate substitutions, the compatibility equation can be reduced to,

\[ \frac{\partial^4 \psi}{\partial x^4} + 2 \frac{\partial^4 \psi}{\partial x^2 \partial y^2} + \frac{\partial^4 \psi}{\partial y^4} = 0 \quad \text{A.6(a)} \]

or,

\[ \nabla^2 (\nabla^2 \psi) = 0 \quad \text{A.6(b)} \]

The solution of a plane extensional problem in linear elasticity involves therefore finding the stress function \( \psi \) in equation A.6b that satisfies the boundary conditions of the problem.

In the case of Model I crack, Westergaard\(^{(12)}\) proposed a complex function

\[ \psi = \text{Re}Z + y \text{Im}Z \quad \text{A.7} \]

where \( Z(z) = \text{Re}Z + i \text{Im}Z \) with \( z = x + iy \), \( Z \) and \( \dot{Z} \) are given by

\[ \frac{dZ}{dz} = \dot{Z}, \quad \frac{\dot{Z}}{dz} = Z \]

Using Cauchy-Riemann conditions, it can be seen that

\[ \nabla^2 \text{Re}Z = \nabla^2 \text{Im}Z = 0 \quad \text{A.8} \]

Consider the problem of a crack of length \( 2a \) in an infinite plate under biaxial stress shown in Fig.A.2. The stress function which satisfies both the compatibility and boundary conditions is

\[ Z = \sigma / \sqrt{1 - a^2/z^2} \quad \text{where} \ z = x + iy \quad \text{A.9} \]

The displacement \( V \) in the \( y \) direction can be shown to be given by
\[ v = \frac{2\sigma}{E} (1 - \nu^2) \sqrt{a^2 - x^2} \] .......................... A.10

which is rearranged to give

\[ \frac{v^2}{4\sigma^2 (1 - \nu^2)^2} + x^2 = a^2 \] .......................... A.11

Equation A.11 shows that the shape of the crack is elliptical.

\[ \text{Fig. A.2 Crack under biaxial tension} \]

For convenience, the coordinate system shown in Fig. A.2 is converted with the origin at the tip of the crack by replacing \( Z \) by \((z + a)\). \( Z \) then becomes

\[ Z = \frac{f(z)}{\sqrt{z}} \] .......................... A.12

where \( f(z) \) is real and a constant at the crack tip. The required value of \( f(z) \) is given the notation \( K \),

\[ Z_{[z=0]} = \frac{K}{\sqrt{2\pi}} \] .......................... A.13

Writing \( z = r e^{i\theta} \) as shown in Fig. A.3, the stresses at the crack tip can be evaluated to be as follows:
Fig.A.3: Coordinates for stresses at crack tip

\[
\begin{pmatrix}
\sigma_x \\
\sigma_y \\
\tau_{xy}
\end{pmatrix} = \frac{K}{\sqrt{2\pi r}} \begin{pmatrix}
\left(1 - \sin \frac{\theta}{2} \sin \frac{\theta}{2}\right) \\
\left(1 + \sin \frac{\theta}{2} \sin \frac{\theta}{2}\right) \\
\left(\sin \frac{\theta}{2} \cos \frac{\theta}{2}\right)
\end{pmatrix}
\]

\[
\sigma_z = 0 \quad \text{A.15}
\]

while for plane strain,

\[
\sigma_z = \nu(\sigma_x + \sigma_y) \quad \text{A.16}
\]

It is noted that the analysis above dealt with biaxial tensile stress system. The uniaxial case can be simply obtained by superimposing a stress of \(-\sigma\) in the x direction. The superimposition of such a stress however, does not alter the crack tip stresses given in equation A.14.
### APPENDIX B

Weight Percent Composition Analysis of Comsteel En25

<table>
<thead>
<tr>
<th>Elements</th>
<th>Composition O</th>
<th>Composition N</th>
<th>Accuracy (+)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Carbon</td>
<td>0.31</td>
<td>0.33</td>
<td>0.02</td>
</tr>
<tr>
<td>Silicon</td>
<td>0.26</td>
<td>0.42</td>
<td>0.02</td>
</tr>
<tr>
<td>Manganese</td>
<td>0.58</td>
<td>0.58</td>
<td>0.02</td>
</tr>
<tr>
<td>Phosphorus</td>
<td>0.025</td>
<td>0.018</td>
<td>0.003</td>
</tr>
<tr>
<td>Sulphur</td>
<td>0.032</td>
<td>0.02</td>
<td>0.0003</td>
</tr>
<tr>
<td>Nickel</td>
<td>2.22</td>
<td>2.45</td>
<td>0.05</td>
</tr>
<tr>
<td>Chromium</td>
<td>0.59</td>
<td>0.60</td>
<td>0.02</td>
</tr>
<tr>
<td>Molybdenum</td>
<td>0.53</td>
<td>0.51</td>
<td>0.02</td>
</tr>
<tr>
<td>Aluminium</td>
<td>0.028</td>
<td>0.040</td>
<td>0.002</td>
</tr>
<tr>
<td>Tin</td>
<td>0.008</td>
<td>0.006</td>
<td>0.004</td>
</tr>
<tr>
<td>Copper</td>
<td>0.05</td>
<td>0.05</td>
<td>0.01</td>
</tr>
<tr>
<td>Vanadium</td>
<td>&lt;0.005</td>
<td>0.006</td>
<td>0.002</td>
</tr>
<tr>
<td>Titanium</td>
<td>&lt;0.002</td>
<td>&lt;0.002</td>
<td>0.004</td>
</tr>
<tr>
<td>Boron</td>
<td>0.0004</td>
<td>0.0004</td>
<td>0.0002</td>
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<td>Arsenic</td>
<td>0.010</td>
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</tr>
<tr>
<td>Antimony</td>
<td>0.005</td>
<td>0.007</td>
<td>0.002</td>
</tr>
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</table>

Table B-1: Weight Percent Composition Analysis of Comsteel En25
APPENDIX C

Weight Percent Composition Analysis of 7075-T6 Aluminium Alloy

<table>
<thead>
<tr>
<th>Element</th>
<th>Percent</th>
</tr>
</thead>
<tbody>
<tr>
<td>Silicon</td>
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<tr>
<td>Iron</td>
<td>0.16</td>
</tr>
<tr>
<td>Copper</td>
<td>1.46</td>
</tr>
<tr>
<td>Magnesium</td>
<td>2.51</td>
</tr>
<tr>
<td>Zinc</td>
<td>5.76</td>
</tr>
<tr>
<td>Nickel</td>
<td>0.005</td>
</tr>
<tr>
<td>Manganese</td>
<td>0.007</td>
</tr>
<tr>
<td>Lead</td>
<td>&lt;0.001</td>
</tr>
<tr>
<td>Chromium</td>
<td>0.19</td>
</tr>
<tr>
<td>Titanium</td>
<td>0.14</td>
</tr>
<tr>
<td>Tin</td>
<td>&lt;0.001</td>
</tr>
<tr>
<td>Calcium</td>
<td>&lt;0.001</td>
</tr>
<tr>
<td>Sodium</td>
<td>&lt;0.001</td>
</tr>
<tr>
<td>Vanadium</td>
<td>0.005</td>
</tr>
<tr>
<td>Gallium</td>
<td>0.008</td>
</tr>
<tr>
<td>Boron</td>
<td>0.002</td>
</tr>
<tr>
<td>Lithium</td>
<td>&lt;0.001</td>
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<tr>
<td>Beryllium</td>
<td>0.001</td>
</tr>
<tr>
<td>Bismuth</td>
<td>&lt;0.001</td>
</tr>
<tr>
<td>Zirconium</td>
<td>&lt;0.001</td>
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<tr>
<td>Strontium</td>
<td>0.001</td>
</tr>
</tbody>
</table>

Table B-1: Weight Percent Composition Analysis of 7075-T6 Aluminium.
APPENDIX D

Compliance Coefficient for Compact Tension Specimen

\[ f\left(\frac{a}{W}\right) = 29.6 \left(\frac{a}{W}\right)^{1/2} - 185.5 \left(\frac{a}{W}\right)^{3/2} + 655.7 \left(\frac{a}{W}\right)^{5/2} - 1017.0 \left(\frac{a}{W}\right)^{7/2} + 638.9 \left(\frac{a}{W}\right)^{9/2} \]

**Fig. D-1:** Compliance coefficient for compact tension specimen
Appendix E

Clip Gauge Calibration Result

Excitation Voltage \[ 3.00 \text{ V} \]

Strain Gauge Bridge Resistance \[ 120 \text{ Ohm nominal} \]

<table>
<thead>
<tr>
<th>Slip Gauge in.</th>
<th>Output Voltage, mV</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>1</td>
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<tr>
<td>0.200</td>
<td>0.000</td>
</tr>
<tr>
<td>0.205</td>
<td>0.182</td>
</tr>
<tr>
<td>0.210</td>
<td>0.363</td>
</tr>
<tr>
<td>0.215</td>
<td>0.547</td>
</tr>
<tr>
<td>0.220</td>
<td>0.721</td>
</tr>
<tr>
<td>0.225</td>
<td>0.895</td>
</tr>
<tr>
<td>0.230</td>
<td>1.073</td>
</tr>
<tr>
<td>0.235</td>
<td>1.253</td>
</tr>
<tr>
<td>0.240</td>
<td>1.431</td>
</tr>
<tr>
<td>0.245</td>
<td>1.606</td>
</tr>
<tr>
<td>0.250</td>
<td>1.786</td>
</tr>
<tr>
<td>0.255</td>
<td>1.964</td>
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<tr>
<td>0.260</td>
<td>2.140</td>
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</tbody>
</table>

Table E.1: Results of three consecutive calibration run.
APPENDIX F

Elastic Crack-Edge Displacement

For Compact Tension Specimen (H/W = 0.6)

TABLE F.1

<table>
<thead>
<tr>
<th>a/W</th>
<th>At load line EBV/P</th>
<th>At knife edges EBV/P</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.30</td>
<td>14.48</td>
<td>26.33</td>
</tr>
<tr>
<td>0.35</td>
<td>18.41</td>
<td>31.46</td>
</tr>
<tr>
<td>0.40</td>
<td>23.22</td>
<td>37.48</td>
</tr>
<tr>
<td>0.45</td>
<td>29.27</td>
<td>45.04</td>
</tr>
<tr>
<td>0.50</td>
<td>37.20</td>
<td>54.83</td>
</tr>
<tr>
<td>0.55</td>
<td>47.90</td>
<td>68.55</td>
</tr>
<tr>
<td>0.60</td>
<td>63.20</td>
<td>88.42</td>
</tr>
<tr>
<td>0.65</td>
<td>85.80</td>
<td>118.83</td>
</tr>
<tr>
<td>0.70</td>
<td>119.80</td>
<td>167.79</td>
</tr>
</tbody>
</table>

Fig. F.1: Variation of BEV/P with a/W.
APPENDIX G

Determination of Strain Hardening Exponent, n

The Strain Hardening Law

The true stress-strain relationship for many materials can be expressed mathematically as

\[ \sigma = A \varepsilon^n_P \]  \hspace{1cm} G.1

where

- \( \sigma \) = true stress,
- \( A \) = constant - the strength coefficient,
- \( \varepsilon_P \) = plastic component of true strain,
- \( n \) = strain hardening exponent

The equation is valid only from the commencement of plastic flow to the maximum load at which the specimen begins to neck.

Assuming the validity of the equation, then necking or maximum load occurs when

\[ \frac{d\sigma}{d\varepsilon_P} = \sigma \] \hspace{1cm} G.2

Therefore

\[ A \varepsilon^n_P = nA \varepsilon^{n-1}_P \]

or

\[ \varepsilon_{pn} = n \] \hspace{1cm} G.3

where \( \varepsilon_{pn} \) is the plastic component of the true strain at necking. The true strain at this point is found from the load-extension curve using the relationship

\[ \varepsilon_n = \ln \frac{L_n}{L_0} \] \hspace{1cm} G.4

where

- \( L_n = \) gauge length at necking,
- \( L_0 = \) initial gauge length

thus

\[ \varepsilon_{pn} = \varepsilon_n - \varepsilon_e \] \hspace{1cm} G.5

where \( \varepsilon_e \) = elastic component of the true strain
### Appendix H

Room temperature tensile properties of Comsteel En 25 and 7075-T6 Aluminium Alloy

<table>
<thead>
<tr>
<th>Material</th>
<th>Tempering Temperature °C</th>
<th>0.2% proof stress MPa</th>
<th>ultimate tensile stress MPa</th>
<th>Elongation Percent on 17.8mm</th>
<th>Reduction in area Percent</th>
<th>Hardness Rc</th>
</tr>
</thead>
<tbody>
<tr>
<td>Comsteel En25</td>
<td>100</td>
<td>1503</td>
<td>0.012</td>
<td>1829</td>
<td>0.002</td>
<td>14</td>
</tr>
<tr>
<td>Composition O</td>
<td>200</td>
<td>1463</td>
<td>0.008</td>
<td>1'37</td>
<td>0.010</td>
<td>14</td>
</tr>
<tr>
<td>R - L Orientation</td>
<td>300</td>
<td>1370</td>
<td>0.021</td>
<td>1534</td>
<td>0.036</td>
<td>11</td>
</tr>
<tr>
<td></td>
<td>400</td>
<td>1270</td>
<td>0.000</td>
<td>1350</td>
<td>0.010</td>
<td>12</td>
</tr>
<tr>
<td></td>
<td>450</td>
<td>1174</td>
<td>0.007</td>
<td>1248</td>
<td>0.000</td>
<td>14</td>
</tr>
<tr>
<td></td>
<td>500</td>
<td>1117</td>
<td>0.020</td>
<td>1181</td>
<td>0.024</td>
<td>16</td>
</tr>
<tr>
<td></td>
<td>550</td>
<td>1012</td>
<td>0.001</td>
<td>1082</td>
<td>0.001</td>
<td>17</td>
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<tr>
<td></td>
<td>600</td>
<td>940</td>
<td>0.015</td>
<td>1020</td>
<td>0.014</td>
<td>18</td>
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<tr>
<td>Comsteel En25</td>
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<td>1592</td>
<td>0.012</td>
<td>1977</td>
<td>0.032</td>
<td>13</td>
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<tr>
<td>Composition N</td>
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<td>0.015</td>
<td>2004</td>
<td>0.036</td>
<td>13</td>
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<tr>
<td>R - L Orientation</td>
<td>300</td>
<td>1589</td>
<td>0.015</td>
<td>1876</td>
<td>0.003</td>
<td>14</td>
</tr>
</tbody>
</table>

Note: The table provides the tensile properties of Comsteel En 25 and 7075-T6 Aluminium Alloy under different tempering conditions. The properties include 0.2% proof stress, ultimate tensile stress, elongation, reduction in area, and hardness.
A P P E N D I X H (Cont.)

<table>
<thead>
<tr>
<th></th>
<th>400</th>
<th>500</th>
<th>600</th>
<th>650</th>
<th>12</th>
<th>25</th>
<th>43.3</th>
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<tbody>
<tr>
<td></td>
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<td>896</td>
<td>831</td>
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<td>0.001</td>
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<td>985</td>
<td>926</td>
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<td></td>
<td>40</td>
</tr>
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<td></td>
<td></td>
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<td>12</td>
<td></td>
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</table>

Comsteel En25

<table>
<thead>
<tr>
<th>Composition N</th>
<th>L-R Orientation</th>
<th>R-C Orientation</th>
<th>7075-T6 Aluminium</th>
<th>77.7*</th>
</tr>
</thead>
<tbody>
<tr>
<td>500</td>
<td>1221</td>
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<tr>
<td>482</td>
<td>547</td>
<td>-</td>
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<td>77.7*</td>
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<td>0.000</td>
<td>0.008</td>
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<tr>
<td>6</td>
<td>5</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

L-S Orientation

| -              | 547             | 597             | 10                | 73.3* |
|                | 0.012           | 0.010           |                   |       |
|                | 16              | 10              |                   |       |

* Hardness in $R_B$
APPENDIX I

Strain Hardening Exponent, n

And Modulus of Elasticity, E

Table H.1.

<table>
<thead>
<tr>
<th>Material</th>
<th>Tempering Temperature °C</th>
<th>n</th>
<th>E (MPa)</th>
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<tbody>
<tr>
<td>En25 As-Quenched</td>
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<td>0.025</td>
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<td></td>
<td>200</td>
<td>0.040</td>
<td>207000</td>
</tr>
<tr>
<td></td>
<td>300</td>
<td>0.037</td>
<td>205000</td>
</tr>
<tr>
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<td>400</td>
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<td>207000</td>
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<td></td>
<td>550</td>
<td>0.057</td>
<td>205000</td>
</tr>
<tr>
<td></td>
<td>600</td>
<td>0.064</td>
<td>214000</td>
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<tr>
<td>En25 As-Quenched</td>
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<td>0.020</td>
<td>188000</td>
</tr>
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<td>188000</td>
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<td></td>
<td>300</td>
<td>0.039</td>
<td>204000</td>
</tr>
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<td></td>
<td>400</td>
<td>0.026</td>
<td>196000</td>
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<td>500</td>
<td>0.032</td>
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<td>63000</td>
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<tr>
<td>Aluminium</td>
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<td></td>
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Cell contents represent specific values for strain hardening exponent (n) and modulus of elasticity (E) under different tempering temperature conditions for various materials.
# Appendix J

Table J.1 Room Temperature Data from Compact Tension Specimens using Comsteel En25

<table>
<thead>
<tr>
<th>Specimen</th>
<th>5% offset Secant load P&lt;sub&gt;max&lt;/sub&gt; P&lt;sub&gt;sec&lt;/sub&gt;</th>
<th>ASTM fracture toughness K&lt;sub&gt;Q&lt;/sub&gt; MPa/m</th>
<th>Max. load fracture toughness K&lt;sub&gt;C&lt;/sub&gt; MPa/m</th>
<th>2.5(ε&lt;sub&gt;YS&lt;/sub&gt;)&lt;sup&gt;2&lt;/sup&gt;</th>
<th>Shear lip size B&lt;sub&gt;SL&lt;/sub&gt; mm</th>
<th>σ&lt;sub&gt;N&lt;/sub&gt; ε&lt;sub&gt;YS&lt;/sub&gt;</th>
</tr>
</thead>
<tbody>
<tr>
<td>1A65N</td>
<td>37.90</td>
<td>110.13</td>
<td>-</td>
<td>43.9</td>
<td>1.80</td>
<td>1.59</td>
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<td>2A65N</td>
<td>56.05</td>
<td>107.17</td>
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* 100% shear fracture

**TABLE J.1 Cont'd**
# APPENDIX K

## Table K.1 Room Temperature Data from Compact Tension Specimens Using 7075-T6 Aluminium Alloy

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<th>Max. load fracture toughness</th>
<th>$K_Q / K_{eq}$</th>
<th>2.5($\sigma_{ys}^2$)</th>
<th>Shear lip size</th>
<th>$\sigma_N / \sigma_{ys}$</th>
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*Note: $P_Q$ is in KN.*
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APPENDIX L

Systematic Error in $K_{IC}$ Calculation

The error propagation formula can be expressed as

$$\left[ \sigma_{F(x)} \right]^2 = \sum_i \left[ \frac{\partial F(x)}{\partial x_i} \sigma_{x_i} \right]^2 \quad \text{(L.1)}$$

where $\sigma_{F(x)}$ is the error in $F(x, x_2, x_3 \ldots)$ and $\sigma_{x_i}$ is the error in independent variable $x_i$.

For compact tension specimen where the fracture toughness is calculated using the equation,

$$K_{IC} = \frac{P}{BW^{1/2}} f \left( \frac{a}{W} \right) \quad \text{(L.2)}$$

the error propagation formula gives

$$\begin{align*}
\left[ \frac{\partial K_{IC}}{K_{IC}} \right]^2 &= \left[ \frac{\sigma_P}{P} \right]^2 + \left[ \frac{\sigma_B}{B^{1/2}} f \left( \frac{a}{W} \right) \right]^2 + \left[ \frac{\sigma_W^{1/2}}{W}\right]^2 \left[ \frac{1}{2} \frac{\sigma_W}{BW^{3/2}} f \left( \frac{a}{W} \right) \right]^2 \\
&+ \left[ \frac{\sigma_f \left( \frac{a}{W} \right)}{BW^{1/2}} \right]^2 \quad \text{(L.3)}
\end{align*}$$

Dividing through by $K_{IC}$, thus,

$$\frac{\sigma_{K_{IC}}}{K_{IC}} = \left( \frac{\sigma_P}{P} \right)^2 + \left( \frac{\sigma_B}{B^{1/2}} f \left( \frac{a}{W} \right) \right)^2 + 0.25 \left( \frac{\sigma_W}{W^{1/2}} \right)^2 + \left( \frac{\sigma_f \left( \frac{a}{W} \right)}{f \left( \frac{a}{W} \right) \left( \frac{a}{W} \right)^{1/2}} \right)^2 \quad \text{(L.4)}$$

from

$$f \left( \frac{a}{W} \right) = 29.6 \left( \frac{a}{W} \right)^{1/2} - 185.5 \left( \frac{a}{W} \right)^{3/2} + 655.7 \left( \frac{a}{W} \right)^{5/2} - 1017.0 \left( \frac{a}{W} \right)^{7/2}$$

$$+ 638.9 \left( \frac{a}{W} \right)^{9/2} \quad \text{(L.5)}$$
the error propagation formula yields,

\[ \sigma_{f(a/W)} = 14.31 \left[ \frac{\sigma(a/W)}{f(a/W)} \right] \]  \quad (L.6)

for \( a/W = 0.5 \), equation (L.6) gives

\[ \frac{\sigma_{f(a/W)}}{f(a/W)} = 1.49 \left[ \frac{\sigma(a/W)}{f(a/W)} \right] \]  \quad (L.7)

From equation (L.4), hence,

\[ \left( \frac{\sigma_{KIC}}{KIC} \right)^2 = \left( \frac{p}{P} \right)^2 + \left( \frac{b_0}{B} \right)^2 + 0.25 \left( \frac{W}{W_b} \right)^2 + 2.22 \left( \frac{a}{W} \right)^2 \]  \quad (L.8)
### APPENDIX M

Determination of C.O.D. at Crack Initiation and $J_{IC}$

**Using Compact Tension Specimen Tempered at 600°C**

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Width W (mm)</th>
<th>Thickness B (mm)</th>
<th>Crack Length a (mm)</th>
<th>Crack Growth $\Delta a$ (mm)</th>
<th>C.O.D. $\delta$ (N/mm²)</th>
<th>$J$ (N/mm²)</th>
<th>$K_{IC}$ MPa/m</th>
</tr>
</thead>
<tbody>
<tr>
<td>10A6N</td>
<td>49.94</td>
<td>24.62</td>
<td>24.30</td>
<td>2.82</td>
<td>0.062</td>
<td>47.85</td>
<td>103.45</td>
</tr>
<tr>
<td>9A6N</td>
<td>50.02</td>
<td>24.61</td>
<td>24.82</td>
<td>2.53</td>
<td>0.053</td>
<td>38.78</td>
<td>98.25</td>
</tr>
<tr>
<td>8A6N</td>
<td>50.09</td>
<td>24.62</td>
<td>24.61</td>
<td>1.60</td>
<td>0.051</td>
<td>41.46</td>
<td>105.48</td>
</tr>
<tr>
<td>11A6N</td>
<td>50.26</td>
<td>25.03</td>
<td>24.50</td>
<td>1.13</td>
<td>0.047</td>
<td>35.67</td>
<td>102.82</td>
</tr>
<tr>
<td>6A6N</td>
<td>49.26</td>
<td>25.03</td>
<td>23.64</td>
<td>0.35</td>
<td>0.042</td>
<td>34.53</td>
<td>-</td>
</tr>
<tr>
<td>7A6N</td>
<td>50.29</td>
<td>25.04</td>
<td>23.83</td>
<td>0.10</td>
<td>0.041</td>
<td>27.81</td>
<td>-</td>
</tr>
</tbody>
</table>

Table K.1: Determination of $\delta$ and $J_{IC}$.
Fracture Modes

As the interpretation of electron fractography has been discussed in detail in References (115,268,269 and 270), only a very brief description of the principle fracture modes is presented here.

Irrespective of whether the fracture occurs by intergranular (along the grain boundaries) or transgranular (through the grains), there are essentially four fracture modes, namely

a) dimple rupture
b) cleavage
c) fatigue
d) decohesive rupture

a) Dimple Rupture

This is the consequence of the growth and coalescence of microvoids that are usually nucleated at areas of high plastic deformation, for example, second phase particles, inclusions and grain boundaries. As the strain increases, the microvoids grow until the material between the two voids thins down and finally ruptures forming cup-like depressions known as dimples. A dimple is half of a microvoid through which the fracture has occurred. The size of the dimple is dependent on the size and density of the nucleation sites and the relative plasticity of the material.

b) Cleavage

Cleavage is defined as the fracture of a crystal or grain along specific crystallographic planes. Due to the presence of imperfections within the grain, the cleaved planes are usually not completely flat and featureless. In fact, these imperfections together with the changes in crystal lattice orientation give rise to such features as cleavage tongues, steps and ridges, river patterns and feather markings.

When cleavage does not occur along true crystalline cleavage planes but on unspecified planes, the fracture mode is termed as quasi-cleavage. This fracture surface feature resembles cleavage in that it contains river patterns radiating from the initiation sites.
c) Fatigue

Fatigue fractures are caused by damage from cyclic or repetitive stress. Generally, the fracture occurs in two separate stages: the Stage I fatigue is the initial propagation of the fatigue cracking process while the Stage II fatigue begins with the formation of arrest marks commonly called striations. Each striation is believed to be the result of one single stress cycle. The spacing between striations is not only affected by the magnitude of the alternating stress, the cyclic frequency but also by the difference in material property and environment. Although by and large the fatigue striations tend to be aligned in the direction of the crack front, locally this direction may not be so because of material discontinuities and local stresses.

d) Decohesive Rupture

Fracture of material that occurs along path of weakness usually caused by precipitation of second phase particles, low strength grain boundary phase or environmental factors is often known as decohesive rupture. The more common environmental factors include stress corrosion cracking and hydrogen embrittlement which generally produce either an intergranular or transgranular fracture topography.
APPENDIX O

Derivation of Net Section Stress, $\sigma_N$

Consider the free body diagram of a compact tension specimen shown in Fig.N.1. The nominal stress, $\sigma$, of the specimen ligament near the crack front is given by

$$\sigma = \left( \frac{P}{A_{\text{net}}} \right) + \left( \frac{M}{I} \right)$$

where $P$ is the applied load, $A_{\text{net}}$ is the area of the uncracked ligament given by $B(W-a)$, $M$ is the bending moment applied at the centre of the uncracked ligament given by $P\left(\frac{W}{2} - \frac{1}{2}(W-a)\right)$, and the ratio $I/C$ is the section modulus of the cross section.

Fig.N.1: Free body diagram of Compact tension specimen.

The section modulus can be shown to be,

$$I/C = \frac{1}{6} B(W-a)^2$$

By substitution, the net section stress of the compact tension specimen can be evaluated as

$$\sigma_N = \frac{P}{B(W-a)} + \frac{3P}{B} \frac{(W+a)}{(W-a)^2}$$

which can be simplified to give

$$\sigma_N = \frac{P}{B(W-a)} \left[ 1 + \frac{3(W+a)}{(W-a)} \right]$$.
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REFERENCES

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<table>
<thead>
<tr>
<th>Number</th>
<th>Author(s)</th>
<th>Title and Reference</th>
</tr>
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