THEORY

A wealth of experimental data on reinforced plastics is currently available to designers.

Much of this relates to unidirectional reinforced materials. Fatigue in metals and alloys is well understood. It is known how the dislocation substructure is involved in the creation of damaging flaws and how it aids the propagation of surface or internal defects if they are already there. In brief, understanding of the process of accumulation of fatigue damage can be unambiguously characterized. The problem with reinforced plastics is that they are inhomogeneous, anisotropic and rarely behave in a linear elastic fashion.

The utilization of composite materials for primary aerospace applications requires adequate safety margins and lifetime under all anticipated environments.

Since all resin matrix composites are visco-elastic to some extent, one would expect their fatigue lives to depend on loading rate and time spent at load.

Fatigue life of graphite-epoxy laminates is governed by time and the type of load applied. Whether the criterion is of the power law or exponential form is not very clear.

Miner's rule which is an empirical relationship tends to give erroneous answers in many real situations in the field of composite materials. Fatigue failure modes must be identified when presenting fatigue data for composite materials especially in notched carbon-epoxy specimens.

For the modes to be established it is essential to understand the failure mechanisms occuring both in the lamina as well as on the laminate. A ply can be considered to have failed when the load has stressed the ply well beyond the proportional limit of its stress strain curve.

This definition on the other hand is over restrictive as the ply exhibits non-linearity. The full stress strain range should be used instead. As mentioned earlier for a fatigue mode to be established various failure criteria must be established and the localized damage growth must be identified. Fig 1 shows the fracture mechanics approach to failure.

3.1 Failure Criteria

Most of the theories developed are supported and apparently justified by very limited amounts of experimental data. The failure criteria are bounded to static loading. For a longtime static and fatigue loading no failure criteria are proved.

The hope is to achieve strain criteria that are valid also for longtime behaviour. Some of the failure criteria related to the static strength of composites are mentioned below even though their applicability in a fatigue case is questionable.

From Hill's formula we get:

$$\frac{\sigma_{\parallel}}{\sigma_{\parallel t}} = 1, \quad \frac{\sigma_{\parallel}}{\sigma_{\parallel c}} = 1 \quad : \text{Fibre Failure} \quad (3.1.1)$$

$$\frac{\sigma}{\sigma_{\perp t}\sigma_{\perp c}} + \sigma\left(\frac{1}{\sigma_{\perp t}} - \frac{1}{\sigma_{\perp c}}\right) + \frac{\tau_{*}^{2}}{\tau_{*}^{2}} = 1 \quad : \text{Inter Fibre Failure} \quad (3.1.2)$$

Because of the empirical nature of the models used, a large amount of testing is required. In the failure analysis of composite laminates, one of the most serious problems has been the propagation of interlaminar cracking, commonly known as delamination.

Delaminations may be formed during manufacturing due to incomplete curing or the introduction of foreign particles, and they may result from interlaminar stresses existing at discontinuities or at the stress-free edges of a loaded composite structure.

This type of failure is a major cause for the deterioration of laminate structural properties, including its strength, stiffness, reliability and durability.

Recently Reifsnider has measured longitudinal stiffness reductions for various CFRP laminates. He concluded that in fatigue the stiffness falls gradually as the density of the crack increases during cycling and also in fibre dominated laminates the 0^{o} layers are the critical layers. In an attempt to finally characterize the fatigue behaviour Mandell et al[23] noted that in the case of U/D carbon/epoxy specimens the normalized fatigue life equation is given by

$$\frac{\sigma_a}{\sigma_u} = 1 - 0.1 \log n \quad (3.1.3)$$

with σ_a = stress amplitude

σ_u = static strength

n = number of cycles

fits the data with reasonable engineering accuracy.

In general composite materials experience during fatigue a number of failures, such as delamination, cracking of the matrix, debonding and fibre breakage. They do not combine to form a single self-similar crack normally found in metals.

Instead they form a very complex damage state in the laminate.

Conclusions reached by[24] show the following:

- Compared to static loading more delamination and matrix cracks develop under cyclic fatigue loading. The fatigue limit is influenced by the matrix.
- Growing damage causes deformation increase (ie stiffness reduction) at cyclic loading and a loss of static strength. In the case of notched specimens a loss of strength may be preceded by an increase in residual strength.
- Stress concentrations reduce the fatigue strength in the short life region, in the long life region their effects diminish. Thus the fatigue factor decreases with increasing life (in contrast to metal behaviour).

3.2 Localized Damage Growth

3.2.1 Stiffness reduction due to the presence of a notch

The stiffness reduction due to the presence of a notch in a plate is that in the case of U/D carbon/epoxy specimens the normalized fatigue life equation is given by

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3.2 Localized Damage Growth

3.2.1 Stiffness reduction due to the presence of a notch

The stiffness reduction due to the presence of a notch in a plate is reflected in the extra displacement of the load application points for a given load.

The following deviation is based in the work of Tada, Paris and Irwin (1973)[25].

The total strain energy of a plate of unit thickness may be regarded as being made up of the energy due to applying the load with no crack or notch present, plus that due to introducing the crack whilst holding the load constant thus:

$$U_T = U_{nc} + \int_o^a \frac{\partial U_T}{\partial a}\, da \qquad (3.2.1.1)$$

where U_T is the total strain energy of the plate with a crack, and U_{nc} is the strain energy of the plate with no crack.

As the strain energy release rate G is defined as $\partial U_T/\partial a$, eqn (3.2.1.1) may be rewritten as:

$$U_T = U_{nc} + \int_o^a G da \qquad (3.2.1.2)$$

By making use of Castigliano's theorem

$$\Delta = \frac{\partial U_T}{\partial Q} \qquad (3.2.1.3)$$

By differentiating (3.2.1.2) with respect to load Q, and substituting eqn (3.2.1.3) we get:

$$\Delta = \frac{\partial U_{nc}}{\partial Q} + \frac{\partial}{\partial Q} \int_o^a G da \qquad (3.2.1.4)$$

We may write

$$\frac{\partial U_{nc}}{\partial Q} = \Delta_{nc} \qquad (3.2.1.5)$$

where Δ_{nc} is the displacement with no crack present.

$$\text{Thus } \Delta = \Delta_{nc} + \frac{\partial}{\partial Q} \int_o^a G da \qquad (3.2.1.6)$$

If the gauge length is L, then we define the original stiffness reflected in the extra displacement of the load application points for a given load.

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$$\text{Thus } \Delta = \Delta_{nc} + \frac{\partial}{\partial Q} \int_o^a Gda \qquad (3.2.1.6)$$

If the gauge length is L, then we define the original stiffness

$$E_o = \frac{\sigma L}{\Delta_{nc}} \qquad (3.2.1.7)$$

and the reduced stiffness as

$$E = \frac{\sigma L}{\Delta} \qquad (3.2.1.8)$$

where Δ is calculated from (3.2.1.6)

Use of the isotropic relation with reference to Fracture Mechanics:

$$K_I^2 = E_o G_I \text{ (plane stress)} \qquad (3.2.1.9)$$

where $K_I = \sigma\sqrt{\pi a}$ for a centre crack, length 2a, in an infinite panel leads to:

$$\Delta = \Delta_{nc} + \frac{2\sigma\pi a^2}{E_o W} \qquad (3.2.1.10)$$

by using equation 3.2.1.6, where we have a factor of x2 as there are two crack tips, and W is the width of the plate.

Substituting eqn (3.2.1.10) to (3.2.1.8) and normalizing with respect to E_o, as defined in eqn. (3.2.1.7) we have:

$$\frac{E}{E_o} = \frac{1}{1 + \frac{2\pi a^2}{LW}} \qquad (3.2.1.11)$$

We assume that L is large enough that the points of measurement are outside the region where the stresses are influenced by the crack.

Thus, the closest distance for compliance measurements satisfying this condition appears to be $L/a \cong 3$ (Broek, 1982)[26]

In the case of a finite width panel, or an edge cracked panel, the approach is identical, but rather more involved.

The stress intensity factor K_I is now:

$$K_I = Y\sigma\sqrt{\pi a} \qquad (3.2.1.12)$$

where $Y(a/W)$ is the finite width correction factor.

Substituting eqn (3.1.1.12) to (3.1.1.9) and then eqn (3.1.1.6) yields (Toda, Paris and Irwin, 1973):

$$\frac{E}{E_o} = \frac{1}{1 + \frac{4a}{L} V_1(\frac{2a}{W})} \qquad (3.2.1.13a)$$

for the finite width centre-cracked panel, where:

$$V_1 (\frac{2a}{W}) = - 1.071 + 0.250 (\frac{2a}{W}) - 0.5357 (\frac{2a}{W})^2 + 0.121 (\frac{2a}{W})^3$$

$$- 0.047 (\frac{2a}{W})^4 + 0.008 (\frac{2a}{W})^5 - 1.071 (\frac{W}{2a}) \ln (1- \frac{2a}{W}) \qquad (3.2.1.13b)$$

and

$$\frac{E}{E_o} = \frac{1}{1 + \frac{4a}{L} V_2(\frac{a}{W})} \qquad (3.2.1.14a)$$

for the panel containing an edge crack of length, a, where:

$$V_2 (\frac{a}{w}) = \frac{\frac{a}{w}}{(1 - \frac{a}{w})^2} \; 0.99 - \frac{a}{w} (1 - \frac{a}{w})(1.3 - 1.2 \frac{a}{w} + 0.7 (\frac{a}{w})^2) \qquad (3.2.1.14b)$$

It can be seen that eqn (3.2.13) tends to take the form of eqn (3.2.1.11) as $(2a/w)$ tends to zero.

3.2.2 The Orthotropic case

The above deviation is valid only for an isotropic panel. The case of an infinitely wide orthotropic plate is given as follows:

The relation between the stress intensity factor K_I and the strain energy release rate G_I for an orthotropic plate, may be written (Sih, Paris and Irwin, 1965)[27] as:

$$K_I^2 = E_{eff} G_I \text{ (plane stress)} \qquad (3.2.2.1)$$

where

$$E_{eff} = 2E_T \left(\frac{1}{E_A E_T} - \frac{A}{E_A} + \frac{1}{2G_A}\right)^{-1/2} \quad (3.2.2.2)$$

where E_A, E_T and G_A are the axial, transverse and shear moduli respectively, and $_A$ is the major Poisson's ratio . Furthermore, Sih and co-workers showed that the relation:

$$K_I = \sigma \sqrt{\pi a} \quad (3.2.2.3)$$

is also valid for the case of an orthotropic panel. Hence the use of eqn (3.2.2.1) instead of eqn (3.2.1.9)

leads to $$\frac{E}{E_o} = \frac{1}{1 + 2 \left(\frac{E_o}{E_{eff}}\right) \frac{\pi a}{LW}} \quad (3.2.2.4)$$

for the stiffness reduction of an infinetely wide, centre-cracked, orthotropic panel.

To calculate the stiffness reduction of a finite width panel, it would, in principle, follow the same procedure as for the isotropic case. However, the stress intensity factor finite width correction factor Y is a function of the elastic constants for the orthotropic case (Bowie and Freese, 1970)[28], and it not generally available in the literature. It is therefore, suggested to resort to empirical methods.

Sih further concluded that for a given notch length, the stiffness reduction of the orthotropic specimens is far greater than the isotropic specimens and that the stiffness reduction depends on the type and size of notch.

There are no analytical expressions, and therefore results indicate that for edge notched specimens the stiffness reduction for the orthotropic material is smaller than the isotropic one for a given notch length. But limited data could only be misleading and therefore various empirical formulae established should be checked.

Finally Rosen (1982)[30] has suggested four main reasons, apart from the lack of understanding on local damage and load interaction effects, which may help to explain why predicted crack propagation rates are inaccurate.

They are as follows:

i) uncertainty in the local stress level

ii) uncertainty in the stress intensity factor calculation

iii) insufficient knowledge of the load spectrum

iv) possible environmental effects

Though these reasons are formulated for metals, they are equally applicable, if only by analogy, to composite materials.

CHAPTER 4 **Page**

Discussion and Conclusions

DISCUSSION

4.1 ULTIMATE TENSILE STRENGTH TEST

A wide range of UTS values were obtained during testing of the specimens. Graph 1 shows the wide variation in the results.

This variation may be attributed to variables described such as defects, laminate porosity, ply gap defects, ply waviness, machinery used and surface modules[29]

In the case of laminate porosity the matrix dominated properties are degraded by:

- 5% strength reduction for 1% porosity
- 50% life reduction for 1% porosity

Fibre dominated properties are not affected but delamination growth is affected. In the case of ply gap defects an 8,7% strength reduction can be recorded. Up to 25% reduction in strength could be recorded due to ply waviness.

Finally in the case of surface notches, static strength is being reduced up to 50% and local delamination occurs at the root of the notch[29]

A photomicrographic technique similar to the one described in Appendix A was employed in order to record void content. The content was found to be up to 4%. Acid digestion technique suggested by ASTM D 3171-76 proved to be unsatisfactory since resin was still attached to the fibres and it was proved difficult to clear them from the surface of the fibres. From the values recorded during testing between the various batches there is a rather high variation in the UTS. Between both [A] and [D] the variation is as high as 50%. This comparison does not include Batches [E] and [A'] since their UTS values are extraordinary high.

It seems also that a more experienced laminator and a better batch of materials should be used. During UTS testing, the type of failure was observed to be a fibre breakage. Initially the specimens failed near the grips, but those results were rejected.

After further examination under the microscope it was found that specimens from the batches [C] and [D] suffered severe shear failure. This was attributed to the fact that a slight missalignment of the ply of up to 7^{o} degrees was established between the first and the successive layers of the laminates.

In order to understand the failure occuring during a tensile testing of carbon/epoxy specimens the following explanation is given:

Failure mechanisms are generally studied as they affect the whole, or macroscopic part. Furthermore, the ways a large specimen can be loaded are similar to loadings found within the microstructure of the specimen. Here the failure characterization will be related to carbon fraction in order to understand the damage mechanism occuring:

4.2 FAILURE CHARACTERIZATION

Consider the case where all carbon fibres are in line with the direction of the tensile load. As the structure is loaded, both the carbon and the matrix will elongate, producing equivalent strain in both elements. However, since the carbon modulus is much higher, it will be less stressed.

Whether failure is indicated in the carbon or matrix depends mostly upon the resin properties. If the resin cracks first, then all the load is transferred to the carbon. Still having the ability to carry the entire load, the structure will stay together and multiple resin fracturing will take place. However, if the carbon cannot carry the load, then the entire structure fails at the point where the resin reaches its ultimate strain.

Another point is an idealised stress-strain curve of a structure where the matrix yields and the carbon fibres carry the load. Multiple dispersed microcracks are similar to yielding. The matrix and fibres both act together to give some initial modulus. When the fibres alone carry the load, a second modulus is introduced. Stress-strain curves from actual tests seldom look the same because shear forces detour the load around the resin cracks and transfers the load from the carbon fibres adjacent to the next resin crack. Th resin is carrying part of the load, hence an early straight line, followed by a curve, is more indicative of what is happening in an actual sample.

4.3 THE MACROSCOPIC VIEWPOINT - FAILURE PROGRESSION

In order to understand observations made on a microscopic scale, it is necessary to analyse what is happening within the specimen. In the following paragraphs examination of early damage mechanisms, damage stopping mechanisms and later damage mechanisms are discussed.

4.3.1 Early Damage Mechanisms

Micro damage can be caused by localized stress concentrations, such as fibre interfaces, debonds, voids, pre-existing cracks and close adjacent fibres.

4.3.1.1 Fibre Interfaces

There is usually a stress concentration around the fibre because of the discontinuity in the material. Since the resin elongates more than the fibre, a stress concentration of about 1.5 or more times the average stress exists at the edges of the fibre.

4.3.1.2 Debonds and Porosity

A cylindrical hole such as a debond will create a 3 to 1 stress concentration. A spherical hole, such as a bubble, would cause a stress concentration of approximately 2 to 1.

4.3.1.3 Cracks

All cracks may be modelled by progressively tighter elliptical cracks. The amount of stress concentration around an elliptical crack, depends on the length and the radius of the crack. The longer the crack and smaller the radius, the higher the stress concentration.

4.3.1.4 Close Adjacent Fibres

Microcracks between adjacent fibres can occur very early in the stress/strain curve but may not propagate until much later.

Symmetry suggests that fibre centres would move with the average strain of the composite.

If the fibres are considered totally rigid, all the strain over the distance between the fibre centre (L) would be concentrated in the matrix between the fibres, which is initially (ℓ) or (L) minus the fibre diameter (D). Hence, a strain concentration of L/ℓ is generated.

As the distance between fibres (L) decreases, the stress concentration increases rapidly. Although, when the fibres touch, ℓ is equal to zero, the stress concentration cannot equal infinity because the fibres are not totally rigid.

4.4 Later Damage Mechanisms

As loading increases, later damage mechanisms are usually linked to the reinforcement and its load transfer with the matrix. These include progressive brittle phase cracking, parallel fibre debonding and fibre breakage.

4.4.1 Fibre breakage

When the resin is rubbery and the carbon fraction low, multiple fibre breakage can occur. When the length of the fibre is short enough that the shear required to load the centre of the fibre to failure exceeds the shear strength of the interface, the fibre will debond.

When transverse cracks criss-cross or debond the parallel fibres a weep can occur. The load causing a transverse crack is redirected to uncracked phases, by shear. Microscopically the loading is drastically changed by cracks. The viscoelastic matrix takes time to readjust and can cause debonding due to high shear at the interface near the crack.

4.4.2 Debonding of parallel fibres

The work to debond a fibre is related to its surface area, the shear strength of the bond or resin, and the compliance of the fibre and the matrix. The tensile load in the fibre is the integral of the shear forces from the edge of the fibre towards the centre.

4.5 Damage Stopping Mechanisms

The mechanisms that can prevent early damage from progressing are: increased elasticity or compliance, reduced stress concentration, crack tip blunting and redirecting, plasticity, pullout friction, high strength barriers and load rerouting.

4.5.1 High strength barrier

When a crack grows the stress concentration increases as the square root of twice the length divided by the crack tip radius. As this crack grows larger the stress concentration increases and crack growth accelerates. As the crack nears a set of fibres, an opposite phenomenon occurs. The fibres are reinforcing the resin and reducing resin-strain near the fibre surface. Therefore, stress concentrations which are trying to cause a high tensile stress at the tip of the crack in the resin, are transferred through shear, to the carbon.

Since the matrix is bonded to the carbon it will not allow the crack to progress unless the crack can overcome the fibre restriction. Therefore in many cases a crack stops just before contacting the fibres because the load is not concentrated directly at the crack tip but is spread to the bonded fibres by matrix shear.

This shear distribution causes a large effective crack tip and reduces the crack stress concentration. As the loading continues, the stress on the fibre grows. If the crack reaches the fibre, the resin may debond or the fibre may fracture.

4.5.2 **The critical carbon fraction** - describes a situation where there are enough fibres to stop a crack, a complex result of the size and stress concentration of the crack, the strength of the resin and fibres, and the orientation and amount of fibres. A few simple approximations can be made relative to this important effect.

(a) The running matrix crack tip stress should equal the matrix strength

(b) The crack typically is very large in comparison to a single fibre in its path. Hence the average strength of the fibre barrier is directly proportional to the average fibre volume through which the crack must propagate

(c) in most reinforcement structures about half of the carbon fibres are typically 50 - 100 times the strength of the resin.

4.5.3 **Pull out** is another important way a crack can be retarded especially when the fibre still bridges the crack. Pulling debonded fibres out of the matrix can involve a substantial amount of energy. If these fibres are not clearly debonded they will have small amounts of resin attached to them which will exert considerable friction on the matrix during the pullout. The energy required to pull the fibres out increases with friction. The friction will depend on the material's coefficient of friction and poisson ratios. The poisson ratios influence the tightness of fit of the debonded fibre with the matrix.

During tensile loading a composite shrinks laterally according to its poisson ratio. When a fibre is unloaded by a debond, the poisson ratio of the fibre dictates its lateral expansion. Further damage is retarded by the high friction of crack bridging fibres with more energy being required to expand the damage.

4.6 TENSILE FAILURE MODEL

As a complement to the abovementioned experimental predictions D Valentin and A Bunsell[32] introduced a probabilistic model where the total failure of the specimen under tensile testing is the result of the successive failure of the fibres.

Zweben[31] considered the effect of the load being transferred primarily onto intact fibres in the neighbourhood of the break. He further considered a unidirectional composite of length L_0 containing N_0 fibres for which the strength probability distribution as a function of applied stress, σ, and fibre length, L, is given by:

$$F(\sigma,L) = 1 - e^{-L\left(\frac{\sigma}{\sigma_0}\right)^{\beta}} \qquad (4.6.1)$$

The length of fibre in the composite which should be taken, δ, must be related to the critical load transfer length. In this case the composite can be considered to consist of N_0 separate chains having L_0/δ links. In such an arrangement the number of breaks recorded for a given applied stress on the fibre, σ_f, without taking into account any redistribution of stress due to broken fibres will be:

$$N_1 = N_o \frac{L_\Omega}{\delta} F(\sigma,\delta) \qquad (4.6.2)$$

Rosen gave the following expression for δ

$$\delta = \frac{1}{2} d \left[\frac{E_f}{G_m} \frac{1 - V_f^{1/2}}{V_f^{1/2}} \right]^{1/2} \text{Cosh}^{-1} \left[\frac{1 + (1 + (1-\Phi)^2}{2(1 - \Phi)} \right] \qquad (4.6.3)$$

where d is the fibre diameter and Φ is defined from the relation, $\sigma(\delta) = \sigma\Phi$, $\sigma(\delta)$ is the stress in the fibre at the distance δ from the break and σ is the stress of the unbroken fibre.

In the case where no stress concentration occurs the stress supported by the intact fibres is given by:

$$\sigma_f^* = \frac{\sigma_f}{1 - F(\delta,\sigma_f)} \qquad (4.6.4)$$

where

$$\sigma_f = \frac{\sigma_\Omega}{V_f} \left[1 - \frac{E_m}{E_f}(1 - V_f)\right] \qquad (4.6.5)$$

and the total number of broken fibres will be

$$N_T = N_o \frac{L_\Omega}{\delta} F(\sigma_f^*,\delta) \qquad (4.6.6)$$

It can be seen that the total failure is the result of the broken fibres and the controlling parameter in composite failure. This is justified from the results obtained after tensile testing and after careful microscopic analysis. See Figs. 4 and 5.

4.7 FATIGUE

Referring to a number of observations that were made during fatigue testing of the specimens and after careful examination under the microscope matrix cracking was the main fatigue failure (See Figs 2, 3, 4 and 5). Figs 2, 3 indicate crack initiation and propogation successively while Figs 3, 4 show the final crack running along the fibres up to the grips of the specimen.

It was mentioned earlier that various types of failures could occur during fatigue testing.

The failure in the form of a longitudinal cracking was caused by the shear stresses due to an abrupt load transfer around the semi circular notch was recorded. According to Valentin and Bunsel, carbon/epoxy laminates, exhibit a 'size effect' when perforated with cut outs, such as holes or notches.

The effective strength reduction factor, R, and R is the ratio of the average stress at failure across the effective length of the notch specimen to the average stress at failure for the intact specimen and is a measure of the size effect. They further concluded that for a brittle laminate such as carbon/epoxy the laminate will exhibit an R value close to 1/K where K is the notch stress concentration shape factor, and sharp notches will reduce K to very small values.

After a number of tests a concentration factor of (7-8,5) was found. It will be premature at this stage to indicate that the value recorded was a very accurate one since results[40] have indicated that the stress concentration factor depends on the type and size and anisotropy ratio and could be as high as 10 for CFRP.

The scatter in the results (see graph 2) may be attributed to inapropriate lamination, poor quality of machining as well as high stress concentration factors and variation in the UTS among the batches.

Graph 2 shows the scatter for specimens from batches [A], [B], [C] and [D]. It could be rather misleading to draw a conclusion on the fatigue behaviour of the specimens of various batches made from unidirectional carbon tape.

It is therefore impossible to draw a graph of each individual UTS versus the number of cycles. The percentage of UTS has been used instead for convenience. More points are required for a particular percentage UTS to predict the scatter band more accurately, although five points are recognised as the least number. In addition more data is required at low stresses and long life.

These tests are, however, time consuming since they must be conducted at a low enough frequency to avoid heating of the specimens. An S-N curve based on percentage of UTS is shown in Graphs 3 and 4. The stress level R = 0. A least square straight line is fitted through the data points. See graphs 5 and 6.

The dramatic variation in UTS between the various batches affected the fatigue life of the specimens and therefore the scatter between graphs 3 and 4. The normalised standard error was found to be 0,72 and 0.84 from graphs 5 and 6. These do not, therefore, indicate any linear trend.

As indicated earlier in the theory an empirical formula similar to Mandel's formula of the form S = UTS-Blogn was tried to be established where B is the number of stress reduction per decade of cycles.

B was found to be 6.78 and 3.34 accordingly. A best fit parabola is proposed and indicated in graph 7, the normalized standard error is 0.59 which shows a considerable improvement. Linear and parabolic limits are proposed (Graphs 8,9,10,11). However, the parabolic limits seem to bound the data points closely (Graph 12).

In this instance, the possibility of a parabolic degradation model is proposed, although below 50% of the UTS further testing is required. Whichever model is used a rough safe line area is suggested for design purposes (Graph 13). This area is obtained from the intersection of the area below the line of maximum stress at 70 percent UTS and the straight line S = 100-15 logn. This safe line region is only verified up to 50 percent UTS.

It is worth noticing that the best fit lines for graph 9 varies to graphs 10 and 11. In those cases the best fit lines using best fit lines were found to be parabolas and are convex upwards in 9 and concave upwards in 10 and 11. This is contraditory to the existing shapes for curves found during fatigue testing. It is rather difficult at this stage to establish the shape for the parabola except the fact that is of the form, $S = A\log n^2 + B\log n + C$.

4.7.1 **Trends in Fatigue Results**

Analysing the results on fatigue a least square fit was done on data-points and plotted on graph 5. The formula and error of this line are:

S = 92,31-6,78 logn