FRACTURE MECHANICS OF WELDMENTS: MICROSTRUCTURAL, EXPERIMENTAL AND MECHANICAL ASPECTS

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A brief review is made of the main areas of current research interest on the use of fracture mechanics in welding research. Attention is also given to some aspects/problems of the topics which require further research. Some proposed solutions for toughness testing and defect assessment procedures developed at GKSS Research Center are also covered.

INTRODUCTION

The most reliable, economic and practically feasible design concept is designing against fracture initiation from crack-like defects in weldments for the wide range of structural steels currently used for offshore structures and pressure vessels. This design concept implies that the fracture toughness of all parts of the welded joint must be examined and the lowest toughness region should be identified. Fracture mechanics based toughness testing can then provide a quantifiable material toughness value (corresponding to the individual zones of the welded joint) from which a critical defect size or stress level can be specified for a given operating stress or for a given defect size respectively.

However, a weld joint comprises the weld metal, the heat affected zone (HAZ) and the base metal parts each exposing different properties. The microstructure and mechanical properties of each weld part are closely related to the material as well as the welding process and procedure used. The evaluation of the fracture behaviour of the multipass welds in structural steels presents particular problems due to the heterogeneous nature of the joint and the small width of the HAZ as well as its complexity. Many test methods have been proposed for evaluating the fracture behaviour of welded structures. Soete describes the tests for welded structures in two main categories: 1) scientific tests (fracture mechanics) to determine material properties which are independent of the geometry of the specimen, and 2) technological tests (eg. Charpy, Drop-Weight, Navy tear and wide plate tests etc.) giving tests results depending on the geometry of the specimen (1). Various attempts have been made to correlate the data obtained from these two groups. However, these attempts have met little success.

An increased use of the fracture mechanics concept by designers and concern about the structural significance of the local brittle zones in the HAZs have generally led to the quantitative fracture mechanics based toughness analysis of structural steel welds. The crack tip opening displacement (CTOD) toughness testing practise on weld joints has made good progress and the number of research and industrial groups interested in the fracture problems of welded joints has also grown substantially in recent years.

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The objective of the present paper is therefore to briefly review the state of the art in applying fracture mechanics to structural welded joints. A particular theme of the paper, however, will be concerned with fracture toughness testing and defect assessment procedures developed at GKSS Research Center.

THE MECHANICAL PROPERTIES OF THE WELDS

The development of the commonly used medium strength structural steels to have the optimum strength and fracture properties is based on relatively well-known metallurgical principles. However, establishing optimum weld joint fracture properties is much more complicated because it requires an adequate combination of the base metal type (strength and alloy design) and respective welding procedure. The choice of welding processes and procedures is an essential factor to quality/property control of the welded joint. For example, plates showing good base material mechanical properties and toughness do not necessarily produce high HAZ toughness.

Hardness

The hardness values and their traverse profile across the weld joint usually provide valuable indications about the quality (strength, ductility, hardenability, weld cracking susceptibility and stress corrosion cracking) of a weld/HAZ and about the microstructural constituents. The coarse grained heat affected zone (CGHAZ) adjacent to the fusion line of C-Mn steel welds generally produces a maximum hardness peak which indicates the presence of hard micro-constituents. It is known that HAZ hardness increases with increasing prior austenite grain size (to a lesser degree) and with decreasing cooling rate. The high hardness HAZ indicates very low ductility which can lead to a pop-in event of HAZ cracks. Furthermore, failure analysis information indicates that fracture intiation may develop from points of maximum-hardness HAZ, even in the absence of crack-like defects (2). Recently, the effect of TiN precipitates on HAZ hardness level has been studied by Bowie et al (3). It is evident from their work that the austenite grain size controling effect of fine TiN rich precipitates can also restrict the maximum HAZ hardness by reducing hardenability in C-Mn steels particularly in very low heat input welds where high volume fractions of martensite would normally be anticipated. The presence of TiN in the steels examined in their study reduced the HAZ peak hardness in 0,6 kJ/mm bead on plate welds by up to 59 Vickers hardness points compared to that predicted by the empirical relationship that relates chemical composition and weld condition to peak hardness.

In addition to the CGHAZ embrittlement problem of the commonly used structural steels, low carbon TMCP steels may show HAZ softening problems which can be depicted by a microhardness survey across a welded joint. This zone is usually located at the site of transition from HAZ to unaffected base plate with varying width depending on the heat input level used. Particularly, after flush butt welding, the effect of this low strength soft zone on the transverse weld joint deformation behaviour and the structural significance of this zone yet remains to be clarified.

TOUGHNESS TESTING OF WELDMENTS

Experimental evidence shows that the fracture behaviour of welded joints and also the measured fracture toughness can depend on the selected specimen type/loading mode, the notch orientation (with respect to the weld) and the test temperature.

Charpy-V Notch Impact Tests

The commonly used Charpy-V test is considered by many standards, codes and specifications as a quick and relatively inexpensive quality control test. The vast amount of data accumulated with this test method induces its continued use. However, this test procedure presents some difficulty when considered for quantitative toughness measurement. High Charpy energy which includes the energy to initiate a crack and propagate fracture does not necessarily mean that the weld joint has adequate fracture resistance to brittle fracture. The insensitivity of this test to detect the low toughness CGHAZ region (where weld integrity is generally the most questionable) is inevitably due to its large crack tip radius and propagating crack which both surely sample mixed microstructures. As a result data are susceptible to considerable scatter and interpretation cannot be conducted in a straightforward manner.

For a wide range of structural steel welds no generally applicable correlation between Charpy-V and fracture mechanics toughnesses (CTOD or J) has yet been found. Large amount of Charpy-V and CTOD data on various weld metals of ferritic steels have been reviewed by Dolby (4) and as expected many factors such as specimen thickness, notch location, strain rate and yield strength were found to influence the relationship. The schematic diagram in Fig. 1 shows the influence of yield strength on the Charpy - CTOD relationship.

Due to the limitations on specimen size in thermal simulators, Charpy-V tests have normally been used to measure the HAZ toughness of steels on simulated microstructure. This technique can provide inexpensive and quick toughness comparisons to rank the steels of interest and study the metallurgical factors controlling the HAZ toughness. Yet, there is evidence that HAZ toughness trends between bulk simulated and weld HAZs can be rather different due to the effects of the width of the simulated zone and relative yield strength of the material on either side of the simulated HAZ. In addition to the Charpy tests, however, fracture mechanics tests are also required by a number of codes and specifications to ensure that weld joint toughness is sufficient to avoid brittle fracture. An additional advantage of the fracture mechanics toughness test is that the toughness data can be used for quantitative defect assessment when needed.

CTOD Testing of Welds

The present fracture toughness testing standards and documents BS 5762:1979 (5), ASTM E 1290 (6), EGF P1-90 (7) are specifically developed for the testing of homogeneous metallic materials. At present, unfortunately, there is no available specific standard (or an appendix to present standards) for the fracture toughness testing of welds. Hence, the use of these standards for the testing of weldments requires some modifications on specimen preparation, testing and interpretation of the data. Nevertheless, the situations where CTOD tests are used can generally be divided into two groups;

i) material selection and weld procedure qualification,
ii) defect assessment.

The aim of the first category is to obtain lower bound toughness values (use of deep notched specimens with through thickness notch in full plate thickness: Bx2B, B is thickness, a/W=0,5), Fig. 2. In order to obtain lower bound toughness value for HAZ, the fatigue crack tip should sample a maximum amount of lower toughness zones (local brittle zones, LBZs) of this region. For this reason, it is general practice to use K or 1/2K weld preparations as shown in Fig. 3a. This practise has also been used for characterising hyperbaric repair welds, Fig. 3b (8). If the purpose is to assess the significance of a particular defect in a structure, then the notch position and size (a/W= 0,1-0,5) should simulate the defect of the interest. This is because the measured CTOD toughness value is dependent on the microstructural and stress gradients at the tip/vicinity of the fatigue crack. Fig. 4 shows the effect of the notch position and crack depth on lower bound CTOD transition curves (9).

The CTOD test standards for homogeneous metallic materials recommend to use deep cracked specimens. Therefore, fracture toughness data determined on such specimens are bound to lead to the use of conservative toughness data on material selection, welding qualification and defect assessment procedures. However, in welded structures, defects are often found to be in the form of shallow toe or root cracks. Obviously, the significance of such defects may be assessed in an unduly conservative manner, especially if very low toughness values were used. Selection of the specimen geometry and notch location therefore should depend on the application and objective of the test. For example, the stress state (constraint) at the tip of a short toe crack can be rather different than the deep crack, and hence surface shallow cracked specimen should be used to assess the significance of such defects, although the question of how to measure CTOD on shallow cracked weld specimens must be answered. It has already been shown (10) that the fracture toughness values at the initiation of stable crack growth, δi or Ji are higher for shallow cracks than for deep cracked ones. This implies that the same applied CTOD or J-integral on shallow and deep notched specimens will lead to very different levels of crack tip strains and stresses. Furthermore, the behaviour of a specimen after initiation of tearing will depend on its geometry. However, in the CTOD based defect assessment method, PD 6493 it is recommended (11) to use δc, δu or δm, whichever applies. The appropriate use of the toughness data obtained from shallow notched (a/W=0,1) test pieces in assessing weld defects and characterising weld joints remains yet to be clarified. The fundamental concept of CTOD HAZ testing is obviously to ensure that the fatigue crack tip has been located in the low toughness region. Therefore, the CTOD test report must contain not only a list of toughness values but a metallurgical report on the results of the metallographic mapping/microstructural validation of the region sampled by the fatigue crack tip.

Furthermore, the ASTM, British Standard and ESIS CTOD toughness definitions require the yield strength of a material for the calculation of the small scale yielding portion of the CTOD equation:

$$\delta = \frac{K^2(1-v^2)}{2\sigma_y E} + \frac{r_p(W-a_o)V_p}{r_p(W-a_o)+a_o+z}$$

The choice of such a yield strength value for significantly mismatched weld specimens, particularly with fusion line/HAZ cracks is a problem still to be solved, Fig. 5. The yield strength gradient of the HAZ of multipass welds on StE 460 steel can be experimentally determined (12) by using micro-tensile specimens having 0,5 mm thickness as shown in Fig. 6.

GKSS CTOD (δ_5) Method

The local and direct measurement technique (δ_5 technique) has been developed at GKSS Research Center for determining the CTOD fracture toughness and the crack growth resistance curve. The use of this δ_5 clip gage for testing the HAZ with SENB specimen is shown in Fig. 7 which measures the CTOD from side surfaces of the specimen at the crack tip with 5 mm gage length. The advantage of this measurement concept is that the δ_5 type CTOD can be easily measured on any configuration with a surface breaking crack; no calibration functions are required. There are no geometrical restrictions as with the standardised CTOD techniques which work for SENB and CT specimens only. A further appealing property of the δ_5 technique is that since it is measured locally as a displacement at the location of interest, it does not have to be inferred from remotely measured quantities, like the J-Integral or the standardised

CTOD. This is of particular importance when the specimen is mechanically inhomogeneous, as is the case for mismatched welds. Furthermore, it can be easily estimated as a driving force parameter using the Engineering Treatment Model (ETM) as outlined later in this paper.

The plastic rotation factor r_p used in the CTOD BS formula, can also be experimentally determined in SENB specimens with δ_5 clip gage measurements and the CTOD formula of the BS 5762 standard. This technique clearly assumes the equality of the CTOD values determined by the BS 5762 formula and δ_5 clip gage procedure, Fig. 8 (10). The r_p value found with this technique appears to be specimen geometry (a/W) dependent, Fig. 9. The CTOD (δ_5) measurements are consistent with the calculated CTOD values according to the BS 5762 standard for both deep and shallow cracked specimens (a/W=0,1) if the r_p value of about 0,2 is used in the latter as shown in Fig. 8b (8). This result is of considerable significance with regard to further application of the δ_5 clip gage measurements in the testing of shallow cracked weld test pieces.

The determination of the R-curves of the welded joints is difficult due to their brittle HAZ region. The short brittle crack jumps (pop-ins) cause discontinuity in the R-curves. Nevertheless, so-called CTOD (δ_5) crack resistance curves have been obtained for two weld specimen geometries shown in Fig. 10a and for comparison reason J-R cuves are also included for the same specimens, Fig. 10b. It was attempted to place the fatigue crack tips at the CGHAZ of the welds. For the specimens having a notch perpendicular to the weld length, the pop-in (short arrested brittle crack) occured at lower CTOD and J values compared to the specimens having a notch parallel to the weld. This implies that the specimen with a perpendicular weld is more suitable to sample the CGHAZ since deviation of the fatigue crack from the zone of interest has been avoided (13). After pop-in crack jumps (shown by broken lines), however, the J-R curves indicate somehow lower J values at the point of arrest than at the pop-in initiation. A higher crack arrest toughness level than at the crack initiation is usually expected to stop the running brittle crack. Logically, CTOD (δ_5) measurements indicate higher CTOD at the point of arrest since the crack tip opens more with increasing crack length due to the pop-in crack jump compared to the CTOD level obtained at the point of pop-in initiation.

Numerous studies (13-17) have been carried out to study the structural significance of the pop-ins and distinguish between significant and insignificant pop-ins depending on the crack jump size, compliance change of the specimen and crack arrest capacity of the surrounding tough material. However, there are no generally agreed criteria to assess the structural significance of pop-ins often occuring in CTOD testing of welds. Since pop-ins have generally been shown to cause rather low toughness results in a fracture mechanics test, it seems reasonable to assume that the occurence of a pop-in event in the test specimen will increase the risk of fracture in a real component. Various factors in complex manner control the integrity of a large structure which are difficult to quantify; therefore, the structural significance of pop-ins cannot be soley based on the behaviour of the small-scale specimen, i.e., compliance change of the small CTOD specimen during the pop-in.

Fatigue Precracking

The present fracture toughness testing standards require fatigue precracked specimens with restrictions on the fatigue crack front shape, length and loading conditions (the level of maximum load and stress ratio, R to be used). The through thickness pattern of the welding residual stresses (transverse to the weld length) changes from tensile stresses (+) near the surfaces to a compression (-) as balancing stress at about midthickness. These residual stress components act as additive stresses to the applied stresses during the cyclic loading. Therefore, during the fatigue precracking, compressive stresses (at midthickness region) normal to the crack plane can counteract to the applied cyclic stresses and thereby decrease the magnitude of effective stress intensity range, thus inhibiting crack growth at that region. However, near the side surfaces extensive crack growth occurs due to the summation of the tensile applied stress and the residual stress component in tension. This bimodal fatigue crack shape associated with the pattern of residual stress distribution is shown in Fig. 11. The development of such irregular fatigue cracks does not meet the testing standard requirements for valid CTOD specimen preparation for characterization of welds in the as-welded condition.

However, to meet the requirements in the CTOD and J- testing of welds (containing residual stresses) is almost impossible if one applies standard precracking procedures without any modifications. Alternative precracking techniques have been proposed and used in various laboratories. Among these techniques, the Local Compression (18), the High R-ratio (19) and Reverse Bending (20) methods are the most commonly known. The modified version of the High R-ratio method as a "Step-Wise High R-ratio (SHR)" was proposed by Koçak et al (21-22). With the help of the SHR technique, improvements on the fatigue crack shape of the as-welded CTOD specimens can be achieved. This technique uses an allowable F_{max} value and simply consists of two R-ratio levels. The basic principles of this technique are schematically shown in Fig. 12

<u>Step I (R=0,1) :</u> This first step can be used to initiate and propagate the fatigue crack to the length of about 1,0 mm. By this step, the initiation period of the fatigue crack from a machined notch will be minimized by using the full range of the applied load. During this step, as expected, a minimum or no crack growth will occur at midthickness of the specimen.

<u>Step II (R=0,7) :</u> In the second step, the R-ratio of the cyclic loading is simply increased to 0,7 by keeping the same allowable maximum load as in the first step. This R-ratio should be used to see the improving effect of the high R-ratio on the crack front shape and to propagate the fatigue crack to the required length ($0,45 \leq a/W \leq 0,55$), Fig. 13. The use of the high R-ratio of 0,7 from the beginning of the precracking will increase the total time of precracking considerably. Therefore, it is proposed to use R=0,1 only for the initiation stage of the fatigue crack at the machine notch tip in order to minimize the total precracking time.

The increased level of mean load $F_m = (F_{max}+F_{min}) / 2$, for a given F_{max} will in fact prevent a possible crack tip closure at the compressive residual stress region of the specimen. The applied static load of F_m should be high enough to keep the crack tip open at about midthickness region of the specimen by balancing the compressive stresses. Experience has shown that in most cases using the R-ratio of 0,5 may not provide a high enough mean load to prevent the retardation of the crack growth at the compressive residual stress region. It is obvious that this technique can easily be applied to any specimen geometry and weld type on any standard testing machine without any extra operation and set-up. This is an important simplification of valid specimen preparation for the fracture mechanics testing of weldments in any section size. The CTOD values obtained from through thickness notched weld specimens precracked with R=0.1 after 1%B local compression are compared with those precracked with SHR method and are shown in Fig. 14. The specimens precracked with the SHR method produced comparable CTOD values with the local compression ones and exhibit reduced scatter. Finally, the limit imposed by fracture toughness testing standards on the use of R ratio of 0,1 can be relaxed in order to prepare valid CTOD specimens from residual stress containing welds in a simple manner.

Machida et al (23) investigated the effect of various fatigue precracking methods (local compression, reverse bending, high R-ratio and normal fatigue) on CTOD and the lowest values were obtained for the locally compressed specimens, Fig. 15. They have concluded that the different precracking methods, including local compression and an irregular crack front shape, have relatively little effect on CTOD of welds, especially for HAZ. Furthermore, they suggested that the restrictions on the fatigue crack front shape can be relaxed and the requirement on the minimum crack length to be proportional to the specimen width may be unnecessary.

Local Brittle Zones (LBZ)

With recent attention on the local brittle zone (LBZ) of weld HAZ, considerable concern has often been expressed as to whether CTOD testing with its deep notched specimens is an appropriate tool for assessing the significance of LBZs in offshore steels. The CTOD test has a potential (compared to Charpy test for example) to pick up (with its sharp fatigue crack) the most brittle and often isolated zone of the HAZ, despite the experimental difficulties. If the CTOD test piece samples the LBZ correctly, the toughness is often found to be extremely low. The occurrence of such low toughness values (even when steel and weldments are sound) is claimed to be unrealistic since many offshore structures with LBZ containing weld joints are still in service and apparently the presence of such brittle zones adjacent to the fusion line (CGHAZ) does not endanger the integrity of the structures. Therefore, it has been argued that low CTOD toughness values may reflect the toughness of the microstructurally brittle zone and not the global fracture behaviour of the welded joints. For this reason, the centre cracked tensile (CCT) panel tests (wide plate tests) with various notch types were extensively used by many investigators to obtain an appropriate answer to the question of the structural significance of LBZs. In order to evaluate the structural significance of the LBZs and their instabilities in various conditions, the major differences between shallow and deep notched CTOD specimens and between the CTOD and tensile panel tests should be taken into account. The obvious conservatism of the deep notched LBZ CTOD specimen results should not be generalized and interpreted as a common fracture behaviour of the LBZs.

For this reason an investigation was performed (24, 25) at GKSS on short and long cracked CTOD and CCT panels, containing two bead on plate welds to determine the effects of crack length (a/W) and loading mode (in bending and tension) on the fracture behaviour of LBZs, Fig. 16. The results of these specimens clearly demonstrate the fundamental differences in the outcomes of the small scale CTOD and tensile panel tests concerning the significance of LBZs. The CTOD tests exhibited pop-ins (initiated from LBZs) for all a/W ratios. In contrast, all tensile panels showed fully ductile failure without triggering a pop-in from existing LBZs at the crack tips, Fig. 17. The apparent CTOD toughness of the LBZs increased with decreasing crack depth to width ratio, (a/W), due to extensive ductile tearing prior to cleavage fracture. The shallow cracked specimen exhibited longer ductile tearing prior to the pop-in initiation than the deep notched one. In the case of deep notched specimens, higher constraint more readily provides the critical condition for cleavage initiation with little or no ductile tearing. Hence, the amount of ductile tearing prior to cleavage fracture appears to be strongly dependent on the a/W ratio of the CTOD specimens. This means that the "low toughness" of the LBZs can not possibly dominate the fracture behaviour of the welded joints on CTOD specimens, if sufficient constraint is not available at the crack tip. It further implies that the LBZs may simply be considered as locations at which brittle microstructural phases exist but "behave well" if the defect depth (constraint) is small.

With regard to the comparison of CTOD and CCT tests, this investigation revealed that the presence of LBZs at the crack tip of the CTOD specimen readily provides a "weakest link" to cause cleavage fracture depending on the a/W ratio - the smaller the a/W ratio the larger the ductile tearing. However, further loss of constraint with CCT specimens leads to a fully ductile failure mode regardless of LBZ presence at the crack tips and a/W ratios. Furthermore, in the light of the observed fully ductile failure mode of the CCT specimens, it can be argued that the LBZs may not be significant and the degree of conservatism of the CTOD results might be large. Consequently instability predictions based on this conservatism might well not be realistic.

Microstructural Aspects of the LBZs

Steel manufacturers have responded positively to the controversy concerning the structural significance of LBZs and tried to produce steel grades with high CGHAZ toughness even if they have to be welded with a high heat input welding process. A new mechanism in a particular offshore steel making practice which uses thermally stable Ti oxide or TiN particles to obtain finer HAZ microstructure and high CTOD toughness will briefly be discussed in this section.

It is evident that the weld thermal cycle with a peak temperature of about 1300 °C experienced by the microstructure adjacent to the molten weld metal can lead to a pronounced austenite grain growth, particle dissolution and the formation of hardened transformation products during cooling and hence results in rather low toughness which is susceptible to brittle fracture initiation. In order to prevent embrittlement in this region (caused mainly by the excessive grain coarsening with bainitic microstructure often containing M-A-C constituents at the prior austenite grain boundaries, Fig. 18 (8)) steel manufacturers have made an attempt to restrict the austenite grain growth by introducing finely dispersed stable particles, such as TiN and Ti-oxides into the various steel grades. Various studies have already indicated that the decomposition of the M-A constituents in CGHAZ into ferrite and cementite aggregate can improve the toughness of the CGHAZ. According to the Amano et al results (26), the decrease of the Si content to 0,1% or below, M-A constituents were decomposed at the third thermal cycle, 450 °C and thus improved the toughness of the ICCGHAZ. The surveyed literature (27, 28) indicates that efforts have been made to achieve a fine grained HAZ in high heat input welds by using Ti-microalloyed steels. Titanium is mainly being used by virtue of its ability to form stable nitrides and oxides even at high temperatures as well as forming various other types of particles such as TiC, TiN, Ti_2O, TiO and TiO_2. Additionally, TiN precipitates in complex compositions, e.g. (Ti,V)N and (Ti,Nb)N also depending on the presence of other alloying elements in the steel. The expected role of Ti can be complicated if the steel contains (in addition to Ti) other microalloying elements also, e.g. Nb and V. In this case, complex carbo-nitrides may precipitate and this may influence the grain growth and precipitation hardening behaviours of the steel as well as possibly affecting (reducing) the solubility temperature of the particle. It may further cause a deterioration of weldability and HAZ toughness properties if the interrelationship between the elements is not finely balanced. Therefore, the effect of Nb and V presence in Ti-microalloyed steels still requires further attention. An investigation (28) carried out at GKSS showed the possibility to pin the austenite grain boundaries and prevent excessive grain growth by both utilization of the optimum size distribution of the TiN precipitates and by finely balancing the alloy design, thus improving the CTOD fracture toughness of the CGHAZ/LBZ of StE 355 offshore steel grade, Fig. 19.

Fig. 20 presents the effect of the CGHAZ percentage on CTOD fracture toughness values for multipass welds on three steels. It clearly indicates the non-existence of any relationship; contrary to common assumptions no or very poor relationship in so far as CTOD should have decreased with an increasing portion of the LBZs at the crack tip was found (28). Furthermore, the distance (D) between fusion line and fracture plane was measured at 12 locations of the post sectioned CTOD specimens. The minimum and average values of these measurements were plotted against respective CTOD values in Fig. 21 which again surprisingly indicated no correlation. Therefore, it can be suggested that the post test validation procedure for HAZ toughness tests may involve some other analysis rather than the single 15% CGHAZ requirement of the API 2Z document (29). However, such an analysis, i.e generally agreed probabilistic fracture analysis has not yet been developed; hence, the structural significance of the LBZ based on its low CTOD toughness cannot be quantified exactly at this time. Some of the existing statistical models (30-32) to determine the probability of interaction between fatigue crack and LBZ which can lead to a brittle fracture require detailed information on; i) the toughness distributions for the LBZ and for the surrounding matrix, ii) the distribution/length of the LBZ obtained from metallographic sections, iii) the location/depth/path of the exsisting fatigue crack, iv) the constraint etc. However, Denys (33) suggests to use the wide-plate panel test results to substantiate conservative conclusions drawn from CTOD tests with regard to the engineering significance of LBZ.

Wide-Plate Testing

Various aspects of the testing of weld joints by using flat wide-plates have extensively been discussed by Denys (33-35) and hence this topic will only be covered here shortly. It is known that the wide-plate tests in some cases can simulate the loading conditions of the structural components more readily than small-scale fracture mechanics specimens tested normally under bending mode. Fracture of the flat wide-plates, however, occurs under low constraint condition whereas in standardised small scale fracture mechanics specimens fracture usually takes place in a high constraint situation. It is therefore not surprising that significant differences in fracture performance of weld joints can be observed between wide-plate and small scale fracture mechanics specimens as dicussed briefly (see Fig. 17) in the LBZ section of this paper, Fig. 22 shows a further example of this discrepancy in the performance of welds in CTOD and wide-plate tests and differences in their sensitivities to microstructural changes of the weld metal due to the variation of the nitrogen content (36). The CTOD values (corresponding to the maximum load) presented in this figure were measured with δ_5 clip gages on center cracked wide-plates with transverse weld metal (with varying total nitrogen content) and SENB CTOD specimens having both a/W ratio of 0,5. The results of the Charpy-V results (37) and fracture mechanics specimens (38, 39) clearly show that nitrogen has a definite embrittling effect on weld metal toughness in both as-welded (AW) and stress relieved (SR) conditions. However, the wide-plate test results do not show a similar degree and mode of sensitivity to nitrogen content.

The differences in the outcomes of the small-scale and structurally relevant wide-plate tests present considerable technical difficulty and remains a controversial issue to the engineering community at the present time.

MISMATCHING

Both strength and toughness properties of the defective region (weld joint) of any structure will clearly control the structural performance. The failure behaviour of the structure associated with this defect will certainly be ***influenced*** by the strength levels of the neighbouring zones. Substantial differences in strength properties (mismatching) of the base, weld metal and HAZ may often occur in welded structures. It is common practice to deposit weld metals which have higher strength (***over-matching***) than the steels used in offshore structures. The weld metal is usually being considered a potential site for defects or cracks to be present or develop in welded structures. In this case, the higher strength of the weld metal (which is likely to be the defective region) compared to base metal may provide an optimum weld joint performance by shielding a crack from applied strains. According to the results of Machida et al (40), the benefit of weld metal overmatching can be small for structural components having a higher stress concentration (such as the tubular joints of the offshore structures) compared to flat wide-plate specimens because higher stress concentration causes much more plastic strain concentration in a limited local area. In such a case, they suggest that the improved CTOD toughness is much more important for the structural integrity than overmatching in the welded joint. It should be noted that overmatching can only be effective if the adequate level of toughness of the weld is maintained.

On the other hand, lower weld metal yield strength than the base metal (*undermatching*) will cause a concentration of the applied strain in the weld metal. In this case undermatched weld metals require higher fracture initiation resistance (toughness) to prevent the risk of unstable fracture. The effects of the under- and overmatching weld metals and a/W ratio on "apparent" HAZ CTOD toughness (measured fracture toughness) are shown in Fig. 23 (41). The presence of high strength overmatching (OM) weld metal near the tip of the shallow crack (a/W=0,1) can create a high constraint similar to that in deep notch specimens. Therefore it is possible to observe very low apparent CGHAZ toughness values even with shallow notched specimens if highly overmatched weld metal is present at the vicinity of the crack tip. The apparent toughness values of the same CGHAZ for shallow notches are found to be higher than deep notched specimens for matching (M) and undermatching (UM) welds. Finally, the shallow cracked specimen results further indicate a trend of decreasing CGHAZ apparent toughness values with increasing weld metal yield strength.

As expected, the crack path direction can also be strongly influenced by the heterogeneity of the crack tip vicinity as shown schematically in Figs. 24a and b for three point bend specimens with shallow and deep cracks. Fig. 24c shows the fracture initiation in the overmatched weld metal which extends as brittle fracture predominantly at the fusion line without any deviation, as schematically shown in Fig. 24b. The effect of the specimen design on the apparent toughness is schematically shown in Fig. 25. The specimen having a weld joint parallel to the notch shows the pronounced influence of the mismatching weld metal on CTOD compared to the test piece having a weld joint perpendicular to the notch (13, 41, 42).

The stress-strain or toughness characteristics of the undermatched welds may dominate the fracture performance of the transversely loaded structures. However, this also depends on the size and location of the defect. The results of the tests on mismatched welds carried out at GKSS Research Center (13, 41-44) indicated that the tensile panels with transverse undermatched weld metal showed net section yielding and gross section yielding modes of deformations by simply changing the location of the short crack (a/W=0,1) from weld metal to HAZ region, Fig. 26. It is evident that even a rather small weld metal crack suffers from strain concentration in the undermatched weld metal and causes NSY of the specimen. Therefore, the interaction between weld metal strength, crack size, apparent and intrinsic fracture toughnesses of the weld zones *should be systematically evaluated* to avoid any overestimation of the fracture resistance of mismatched welds.

The fracture assessment of such welds for structural integrity requires detailed information on the effects and/or interactions of each part on the overall fracture behaviour. Therefore, the effect of relative difference (***mismatching***) of the yield strengths of the base, weld metal and heat affected zone on the actual toughness value of the material at the tip of the defect must be determined. Commonly used fracture toughness parameters, however, may not be applied in a straightforward manner to the testing of mismatching welds. It is therefore important to clarify the possible effect of the over- and undermatched weld metal on the fracture toughness parameters, currently an intensely investigated topic in various institutes world wide.

The present fracture toughness testing procedures (even the currently formulated BS, ASTM and IIW draft documents for fracture toughness testing of weldments) do not give a clear description for the CTOD or J-Integral toughness testing of mismatched, bi-material weld joints. At present, it is not even clear as to whether CTOD or J estimations with standard SENB specimen geometries (BxB or Bx2B, B=thickness) are suitable for the fracture toughness determination of mismatched or bi-material weld joints. If so, the question remains which one is more suitable for over- and undermatched joints to be used at different regimes of the ductile-to-brittle transition curve.

Use of J-Integral on Mismatched Joints

The problem of measuring a meaningful J-integral value on a mismatched weld specimen is very hard to solve, since it is not a simple task to distinguish between the contributions from the weld metal at the vicinity of the crack tip and from the base material to the remotely measured load line displacement (LLD) usually used in J estimation. Lee and Luxmoore (45) have investigated the fracture behaviour of undermatched double-V welds with shallow defects in tension and bending using finite element and experimental techniques. It was found that the FE- and experimental J values for matched welds were much higher than the values obtained for 26% undermatched welds, although all plastic strain shoud be concentrated in the weld metal, and hence higher J-values were expected (i.e. undermatching lead also to a shielding effect on J). This surprising fracture behaviour was attributed to the complex yield pattern; severe strain build-up occured at the fusion boundary, leaving the outer central region of the weld unstrained. The FE study of Zhang et al (46) revealed that the mismatch effect on J is fully developed if the width of the weld metal (h) is greater than the remaining ligament width (c). With decreasing weld metal width increasing interaction of the weld metal plasticity with the surrounding base plate suppresses the effect of the weld metal mismatch on J and hence the specimen behaves as if made entirely of base material. The present authors (44) have also made an attempt to show the effects of mismatching as well as a/W and h/c ratios on J and CTOD(δ_5) values obtained on center cracked tensile panels containing austenitic, ferritic and martensitic transverse welds. Increasing a/W and h/c ratios increases the effect of the mismatching on CTOD and to a lesser extent on J values as shown in Figs. 27 and 28 respectively. Kirk and Dodds (47) have recently investigated the effect of weld strength mismatch on J estimation formulas for SENB specimens for various joint geometries and weld widths. Their results suggest that the CMOD based J estimates in SENB specimens are considerably more accurate than LLD based J estimates for cases of extreme overmatch (50% to 100%). However, for highly overmatched welds, CMOD measurements can still be considered as a remote quantity due to the plastic work at the lower strength base plate parts of the specimen. This effect can be particularly extreme on shallow cracked specimens. Therefore it would be ideal, if the critical crack tip characterising parameter can be *locally quantified or measured* on mismatched or bi-material joints and not be inferred from remotely measured quantities, like J-Integral and standardised CTOD.

MECHANICAL ASPECTS AND APPLICATION TO COMPONENTS

The previous sections shortly discussed some features of the welds, problems and developments in the CTOD fracture toughness testing of welded joints. Subsequent sections mainly discuss the CTOD based fracture assessment procedure developed at GKSS which takes the mismatching aspect of the weld joints into account. Discussions on the generally known R6 and PD 6493 fracture assessment methods are beyond the scope of the present communication. The assessment of the severity of a crack in a component is conducted by comparing the material's fracture resistance (= fracture toughness) with the crack driving force present in the component. Failure is thus given by; **Driving Force = Fracture Toughness.** This implies that i) the component exhibits the same fracture toughness as determined on the laboratory specimen (transferability), ii) the driving force can be determined appropriately. Both issues will be dealt with in the following, though with the main emphasis on the latter.

Transferability

Fracture toughness (in particular in the ductile-to-brittle transition of steels and their weldments) exhibits tremendous scatter and is sensitive to the local constraint at the crack tip. An example of the former is given in Fig. 29 (48), the constraint effect is shown schematically in Fig. 30. Scatter presents a problem insofar as statistical treatment of the data should be applied. Methods for planning of tests in order to obtain statistically relevant data with a minimum of effort as well as for statistical description of experimental results are under development, see e.g. (49). To date, often the lowest value obtained in three tests is used (50), thus limiting of course the experimental effort at the expense of a more precise statistical view. It should be mentioned that the British Standard PD 6493 Document provides no advice on the scatter of the CTOD data.

Ideally, the component should exhibit the same toughness behaviour as the laboratory specimen. However, the constraint conditions may be different, thus resulting in different toughness values (or distributions). In most cases the constraint conditions in a component are unknown. The usual escape out of this dilemma is using standard bend or compact specimens which are supposed to produce highest constraint; if the component exhibits less constraint, then the test answer is on the conservative side. The size of the specimen to be used depends on the fracture mode; if fracture occurs in a ductile manner, a specific minimum size (for reaching plane strain constraint) can be deduced from the existing test methods. In the case of cleavage type failure, the specimen thickness should equal the component thickness. If, however, the stress conditions in the component can be characterized as prevailing membrane stresses, a standard bend or compact specimen may be unduly conservative. In this case the component maybe modelled by a tensile panel. Fig. 17 shows for a specific double cross bond weldment that the high constraint bend specimens fail at very low CTOD levels whereas the tensile panels exhibit a completely ductile failure mode, resulting in stable crack growth.

Weldments represent a further transferability problem: the crack configuration used in the test must represent the defect expected or detected in the component. A further point of interest is related to the residual stresses in a weldment. It is known that in the low stress fracture range, the weld residual stress has a significant effect on fracture strength. However, in the plastic range the fracture process is apparently not affected by the residual stress. Cutting the specimen blank out of a weldment will result in some stress relief. The remaining residual stresses in small CTOD specimens may be removed mechanically by the local compression technique (50). Residual stresses in the structure can then be accounted for on the driving force side. This may provide a practical solution but ideally fracture toughness testing should be carried out in the "as-welded" condition (i.e without any extra treatment to the test pieces before the testing) and therefore possible effects of the mechanical treatments prior to the testing should be taken into account. For example, by pre-loading (overstressing) the Tee joint wide plate specimens, residual stresses were mostly removed but compressive stress remained at the tip of the surface notch (51). In this case, the transition curve shifted about 100 ^{o}C to lower temperature after pre-loading as shown in Fig. 31.

DRIVING FORCE

Crack driving force considerations at a weldment present two additional problems as compared to a homogenous material: strength mismatch and residual stresses. It is common practice to superimpose residual stresses with the stresses due to the applied loads in a simplistic manner.

Strength Mismatch

A number of defect assessment methods have been developed which are supposed to cover the assessment of the severity of crack-like defects in welded joints. However, these methods are based on homogeneous materials and generally assume that defects are located in material of **uniform** mechanical/ microstructural properties and it is normal practice to use the tensile properties of the material in which the defect is located. However, in reality the mechanical heterogeneity (differences in tensile properties between the different zones of weld joint) will influence the plastic zone development process at defects and hence affect the relationship between crack driving force and applied loading. Welds are often made with substantial strength mismatch between base material and weld metal. However, the implications of the strength mismatch with respect to the performance of such welds is not known quantitatively. The reasons are; i) crack driving force formulations for mismatched configurations are not available in a straightforward manner, ii) suitable fracture mechanics test techniques have not yet been developed. The test methods presently in use for weldments do not account for strength mismatch. Since modern design and maintenance require quantitative assessment of the performance of welded structures, the problem of mismatch is attracting increasing interest.

As long as the welded joint is in a purely linear elastic condition, it causes no complications in the determination of the driving force. If, however, the weldment becomes fully plastic, the inhomogeneity of the deformation properties across the weld gives rise to variations of the crack driving force accordingly. This is so because in the fully plastic condition the driving force for a given geometry depends on the stress-strain properties of the material. If one looks at the problem very closely, it reveals an extremely complex pattern of stress and strain distributions, see for example (33, 45, 52). In general, experimental and finite element investigations (33, 34, 45, 52-55) demonstrate that overmatched weld metals exhibit a shielding effect due to their higher yield strength, i.e. they attract less strain than the base material, and that strain concentration occurs in undermatched weld metals, both with effects on the driving force acting on a weld metal crack accordingly. Therfore, we propose δ_5 as a parameter characterising the crack tip behaviour. It has the following advantages:
i) δ_5 is a quantity which is locally measured; the measurement is independent of the global behaviour of the specimen; some experimental details have already been shown in the previous section. ii) δ_5 is directly measured as a displacement; no calibration function is needed. iii) It is of particular interest for mismatch welds and for interface cracks. iv) It can be easily estimated as a driving force parameter using the ETM as shown below.
For engineering assessment purposes, a simplified view has been developed which is based on the Engineering Treatment Model (ETM). With this model, the driving force for fully plastic conditions can be expressed in a size and geometry independent formulation (56)

$$\frac{\delta_5}{\delta_Y}=\left[\frac{F}{F_Y}\right]^{1/n}=\frac{\varepsilon}{\varepsilon_Y}=\left[\frac{J}{J_Y}\right]^{\frac{1}{1+n}} \qquad (1)$$

where

$$\delta_Y=\frac{K_{pl}^2}{E\sigma_Y} \quad for \quad F=F_Y \qquad (2)$$

F_Y denotes the yield load,

$$J_Y=\frac{K_{pl}^2}{E} \quad for \quad F=F_Y \qquad (3)$$

$$\frac{\sigma}{\sigma_Y}=\left[\frac{\varepsilon}{\varepsilon_Y}\right]^n, \quad for \quad \sigma\geq\sigma_Y \qquad (4)$$

σ_Y, ε_Y, and n are defined by the power law

This formalism can be applied to a weld metal crack in a weldment with substantial strength mismatch (56-58). For the case of a wide plate with vanishing a/W, loaded transversely to the weld, and for the assumption of plane stress, the CTOD at a weld metal crack, δ_W, is given by

$$\delta_W = \frac{1.5\pi a \sigma_{YB}}{E} M^{\left(1-\frac{1}{n_W}\right)} \cdot \left[\frac{\varepsilon}{\varepsilon_{YB}}\right]^{n_B/n_W} \qquad (5)$$

Here the CTOD is in the spirit of δ_5 (see Fig.7), the subscript B refers to the base material, the subscript W refers to the weld metal, and

$$M = \frac{\sigma_{YW}}{\sigma_{YB}} \qquad (6)$$

is the mismatch factor, i.e. the ratio of weld metal yield strength to base material yield strength; M<1 characterises undermatching, M>1 refers to overmatching. Furthermore, the stress-strain curves of the weld metal and of the base material are represented by power laws of the type shown in Eq (4). Eq (5) is valid for both base material and weld metal being beyond their respective yield points. If only one of these two materials is plastic, other relationships are valid, for details see Refs. (56-58). Fig. 32 shows an example. It can be clearly seen how strongly the local conditions in the weld metal affect the driving force; the strain concentration in the weld metal for that specific case of undermatching leads to a dramatic increase of the CTOD as compared to the base material (M=1), which increases even further during straining. From Eq (5) critical crack lengths can be determined

$$a_c = \frac{\delta_{cW} \bullet E}{\pi \sigma_{YW}} \frac{1}{\delta^*_W} \qquad (7)$$

with δcW being the critical CTOD measured for the weld metal and

$$\delta_W^* = \delta_W \frac{E}{\pi a \sigma_{YW}} \qquad (8)$$

As an example, the maximum crack length at the attainment of full base material plasticity ($\varepsilon/\varepsilon_{YB}$ = 1) is plotted in Fig. 33. Again, the strong effect of mismatch is obvious. The beneficial shielding effect of overmatching is of particular interest. From the formulas and diagrams it is obvious that not only the mismatch factor, M, governs the effect of mismatch on the driving force, but that the hardening behaviour plays also an important role. The above considerations represent a first attempt to quantifity the mismatch effects in analytical form. They are based on simplifying assumptions and will be validated by means of experiments and finite element calculations. An important element in any elastic-plastic and fully plastic fracture mechanics assessment is given by the yield load, F_Y. For example, the kinks in the curves of Fig. 32 are related to the yield loads of the base material and the weld metal, respectively. Finite element work suggests that the yield load of the cracked weld metal can be influenced by the surrounding base material. Thus, further work in this area should be conducted in order to derive analytical yield load expressions taking into account mismatch effects.

CONCLUSIONS

In the present paper we have tried to highlight some important aspects regarding the fracture toughness properties of weldments and their behaviour in structural components. Since this matter has already been treated at length in the open literature, we have presented mainly our laboratory's view.

LBZ: In order to evaluate the structural significance of the LBZs and their instabilities in various conditions, the major differences between shallow and deep notched CTOD specimens and between the CTOD and wide plate tests should be taken into account (constraint effect). The obvious conservatism of the deep notched LBZ CTOD specimen results should not be generalized and interpreted as a common fracture behaviour of the LBZs. This implies that the LBZs can be considered as isolated locations at which brittle microstructural phases exist but they can "behave well" if the constraint (stress state at the crack tip) is small.

Mismatching: The interaction between the weld metal strength, crack size, apparent and intrinsic fracture toughnesses of the weld zones should be systematically evaluated to avoid any overestimation of the fracture resistance of mismatched welds. Therefore, the fracture toughness testing procedures for weldments should give a clear description for the CTOD or J-Integral toughness testing of mismatched, bi-material weld joints. Further refinement of the handling of the effects of mismatch on the driving force is needed. As an example, yield load solutions should be reviewed for mismatch effects. Assessment methods such as PD 6493, R6 method and the basic ETM can be used for structural assessments. A first step has been undertaken to adjust the ETM to the strength mismatch of weldments.

Fatigue precracking/Residual stresses: The limit imposed by fracture toughness testing standards for homogeneous materials on the use of R ratio of 0,1 can be relaxed in order to prepare valid CTOD specimens from residual stress containing welds in a simple manner. It is important to have a better quantitative understanding of residual stresses. This includes a better knowledge of the magnitude of stresses present - both under as-welded and PWHT conditions - and the superposition with stresses from applied loads.

CTOD (δ_5) technique: Based on extensive experience, we propose δ_5 as a parameter characterising the crack tip behaviour also in weldments due to the following advantages: i) δ_5 is a quantity which is locally measured; the measurement is independent of the global behaviour of the specimen. ii) δ_5 is directly measured as a displacement; no calibration function is needed. iii) It is of particular interest for mismatch welds and for interface cracks. iv) It can be easily estimated as a driving force parameter using the ETM.

Finally, the appropriate use of the toughness data obtained from shallow notched (a/W=0,1) test pieces in assessment of weld defects and weld joint characterization is yet to be clarified. Furthermore, there are no generally agreed criteria to assess the significance of pop-ins often occuring in CTOD testing of welds.

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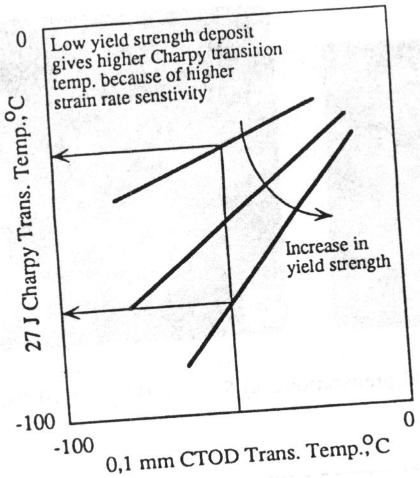


Fig. 1 Schematic diagram showing the effect of yield strength on Charpy-CTOD correlation (4).

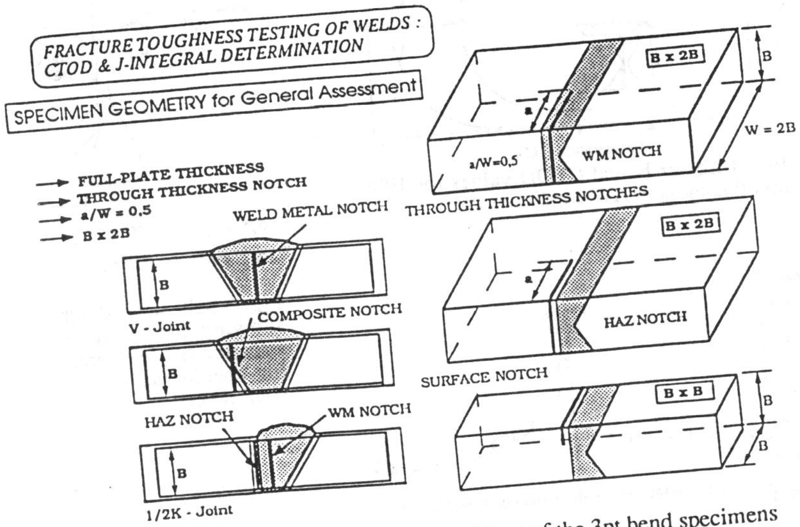


Fig. 2 Schematic showing the notch positions of the 3pt bend specimens for general assessment purposes.

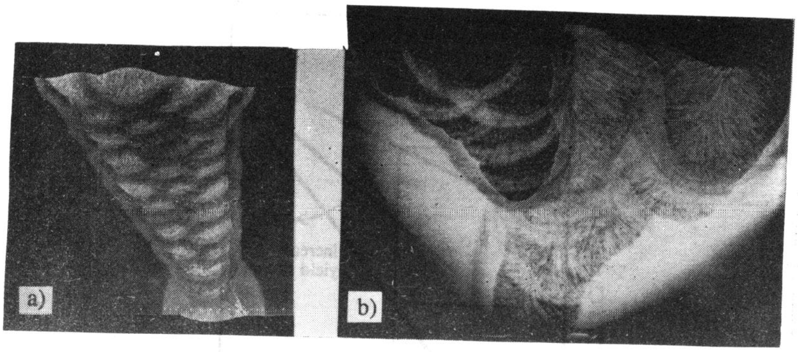


Fig. 3 The 1/2K weld preparations, a) SAW weld joint, b) SAW weld joint repair weld deposit.

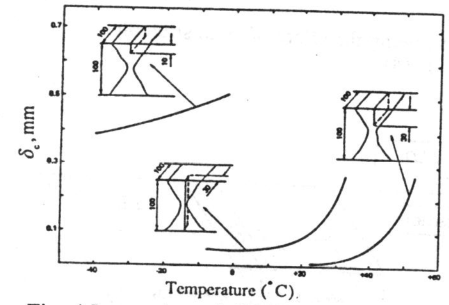


Fig. 4 Lower bound CTOD values for 100 mm thickness in AW conditions (9)

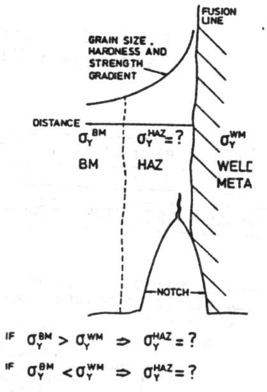


Fig. 5 Schematic diagra showing the problem selection of yield streng for specimens having HAZ notch.

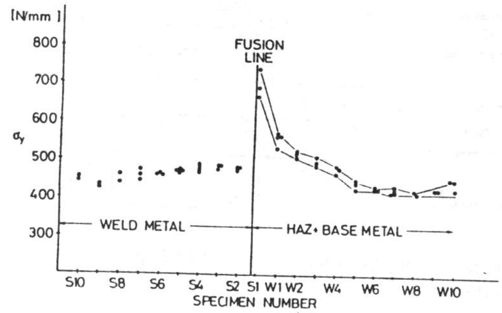


Fig. 6 The yield streng distribution across tl weld joint (12)

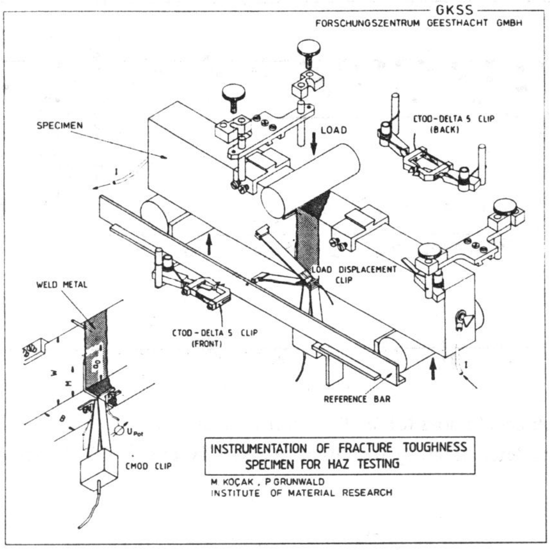


Fig. 7 Instrumentation of the HAZ notched specimen for CTOD and J-Integral testing. Two δ_5 CTOD clip gages are used for direct CTOD measurements.

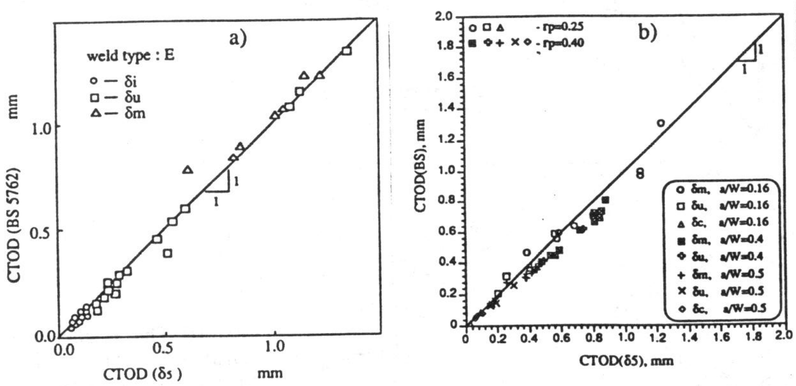


Fig. 8 Comparison of δ_{BS} and CTOD (δ_5) values for multipass weld joints (8, 10)

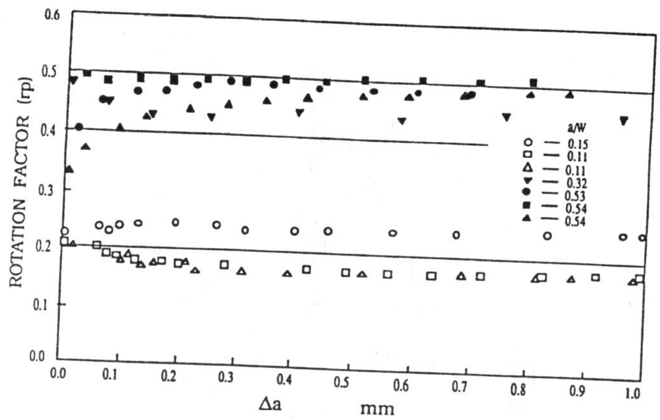
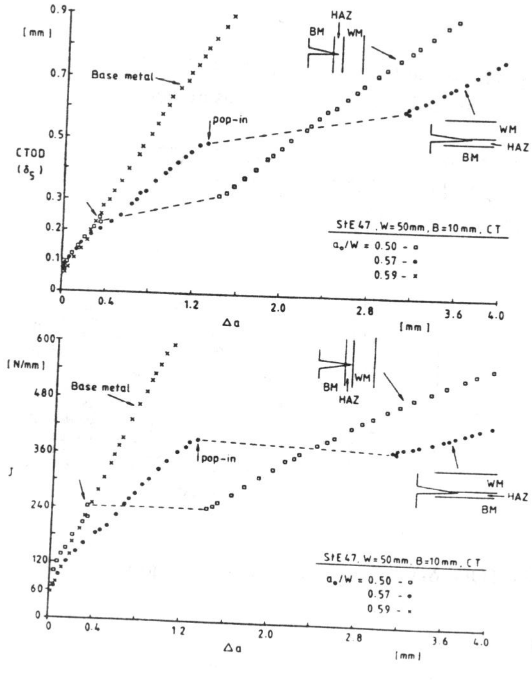


Fig. 9 Rotation factors for shallow and deep cracked weld metal CTOD specimens determined by δ_5 clip gage and CTOD BS 5762 formula [8].



a)

b)

Fig. 10 The R-curves obtained on CT specimens with a weld parellel and perpendicular to the HAZ notch,

a) CTOD (δ_5) R-curves,
b) J-R curves.

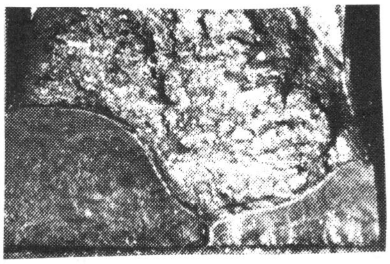


Fig. 11 The bi-modal fatigue crack front development of the 10 mm thick CT specimen due to the welding residual stresses.

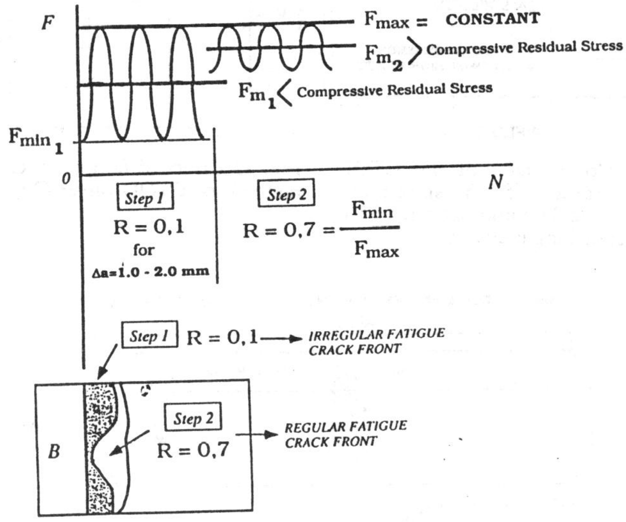


Fig. 12 Schematic diagram showing the SHR fatigue precracking method.

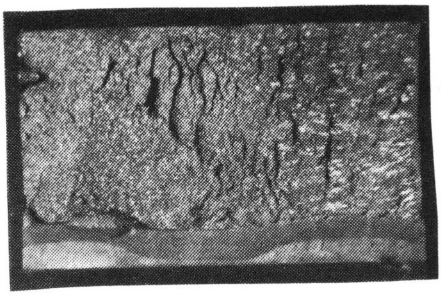


Fig. 13 The improvement of the fatigue crack front by using SHR method.

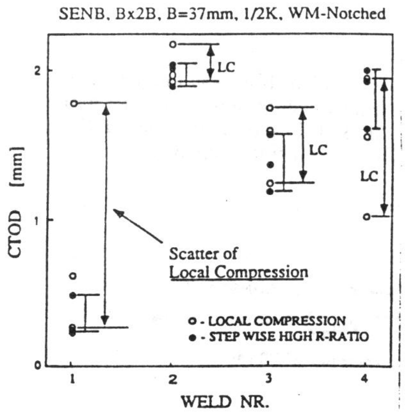


Fig. 14 Comparison of the CTOD values obtained from specimens prepared by local compression and SHR fatigue precracking methods.

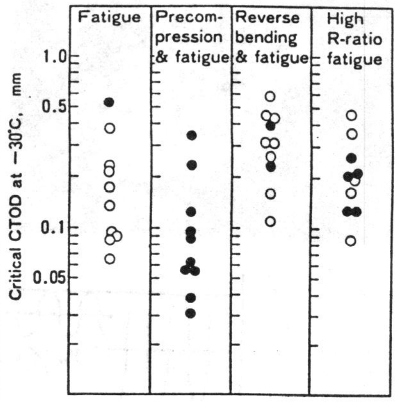
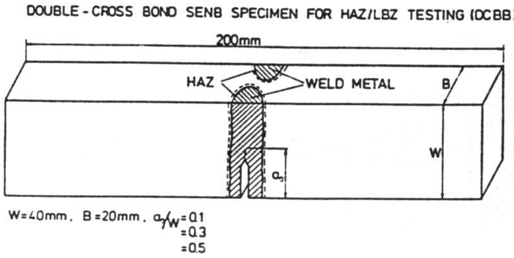
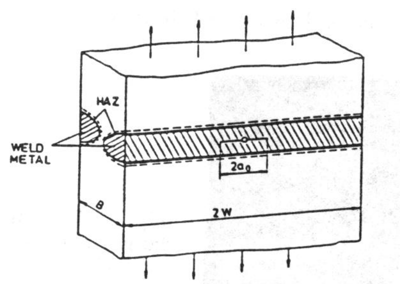


Fig. 15 Effect of fatigue precracking methods on CTOD values (23).



a)



b)

Fig. 16 The double cross bond specimens for testing of HAZ/LBZs.
a) SENB specimen, b) CCT specimen

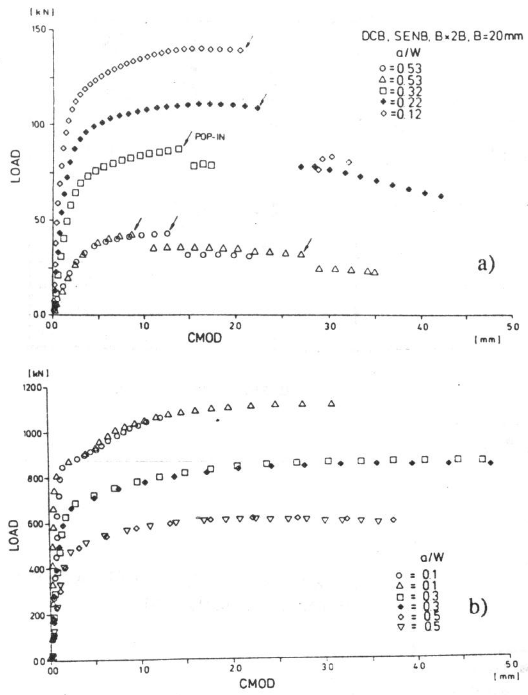


Fig. 17 The load/CMOD curves for CTOD and CCT specimens having identical LBZs at the crack tips, showing the effect of constraint on fracture behaviour of LBZs (24, 25).

a) For CTOD specimens (note to **brittle** pop-ins initiated from LBZs),

b) For CCT specimens (note to **ductile** fracture mode despite the presence of LBZs at the crack tips).

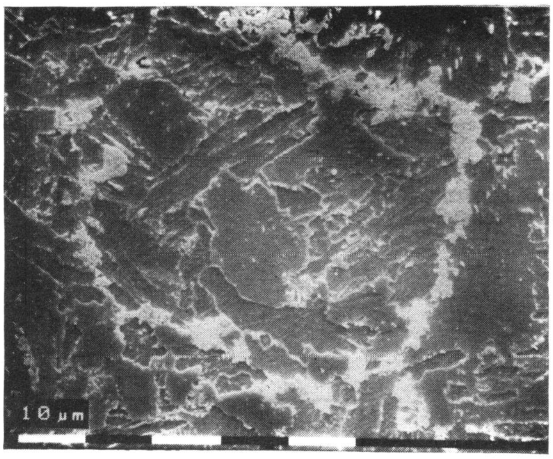


Fig. 18 ICCGHAZ microstructure shows the M-A-C formation on the grain boundary (8).

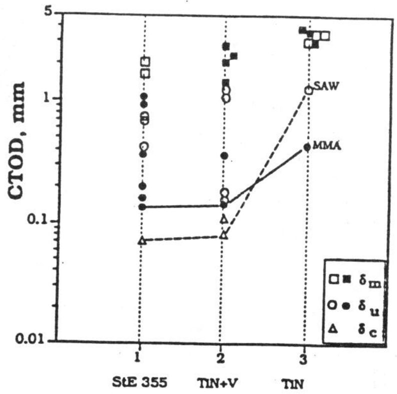


Fig. 19 Improvement of the CTOD toughness in Steel 3 (28).

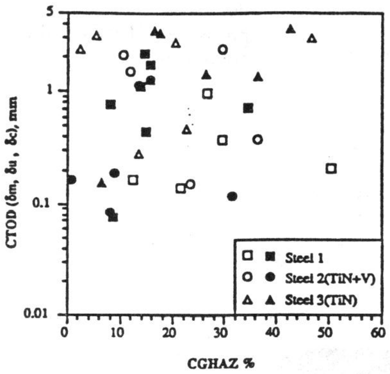


Fig. 20 The CTOD vs. CGHAZ (%) data for three steels (28).

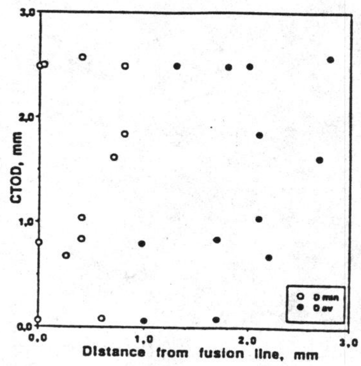


Fig. 21 The CTOD vs. distance between fusion line and fracture plane correlation (28).

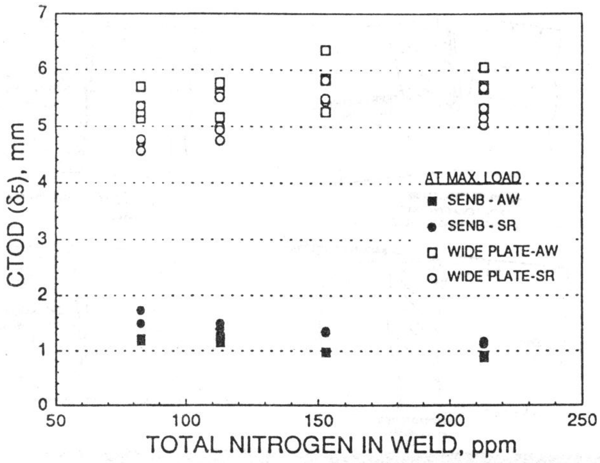


Fig. 22 Effects of specimen geometry and nitrogen amount on CTOD (δ_5) values obtained from SENB and CCT weld metal specimens (36).

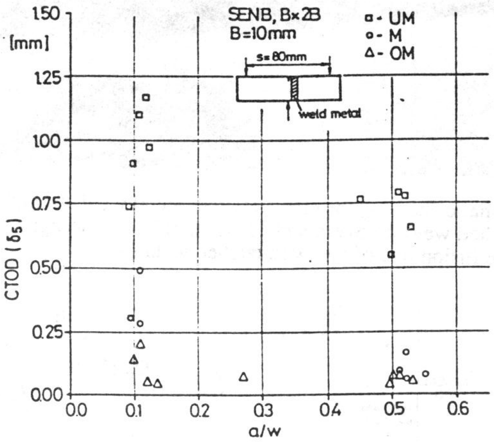


Fig. 23 The effects of mismatching and a/W ratio on CGHAZ apparent toughness (41).

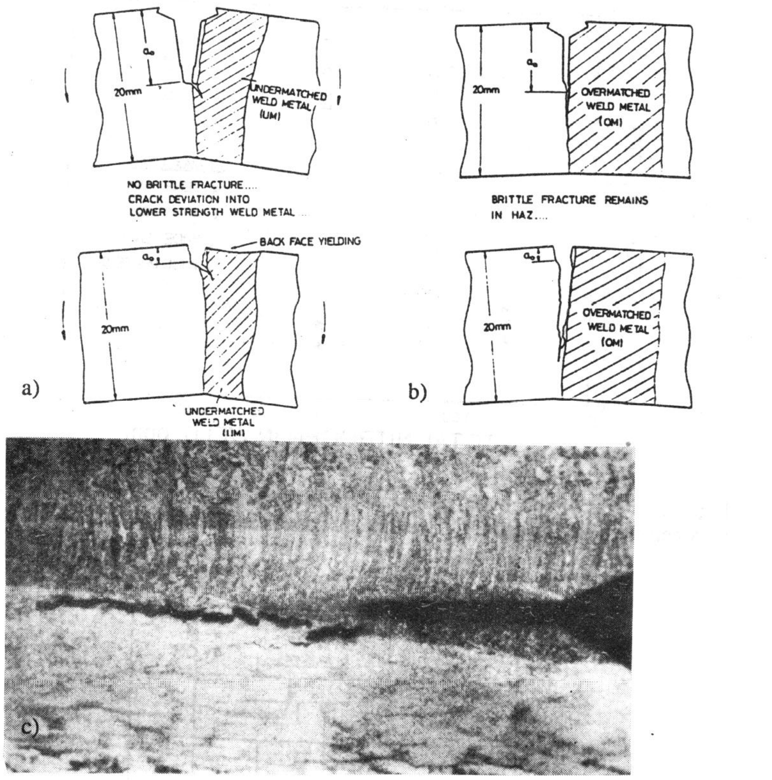


Fig. 24 Schematic showing the effect of weld metal yield strength on crack path (41). a) undermatched weld, b) overmatched weld, c) Photomicrograph showing the brittle fracture at the fusion line of the overmatched weld (8).

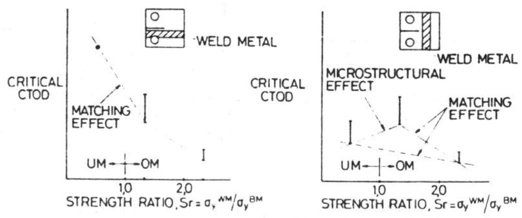


Fig. 25 Schematic showing the effects of specimen geometry and weld metal mismatching on apparent CTOD toughness (43).

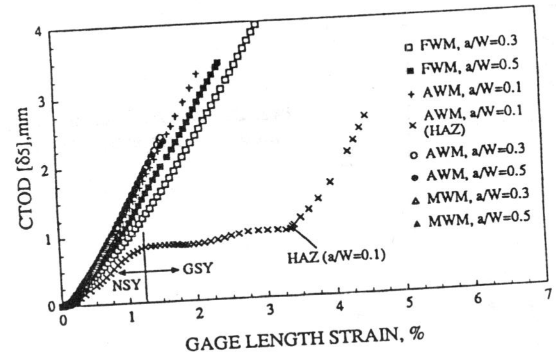


Fig. 26 The relationship between CTOD and gage length strain for CCT panels having transverse undermatched austenitic weld metal (AWM), overmatched ferritic (FWM) and martensitic (MWM) weld metals (44).

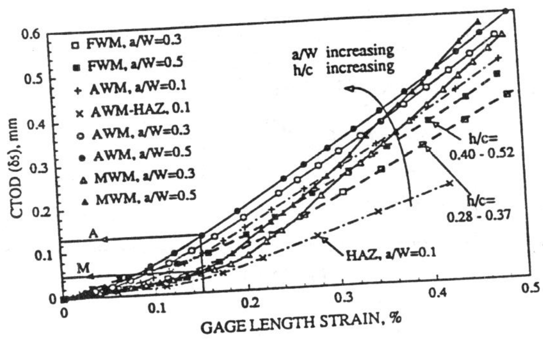


Fig. 27 Effect of mismatching, a/W and h/c ratios on the CTOD vs. strain relationship for same specimens as in Fig. 26. (44)

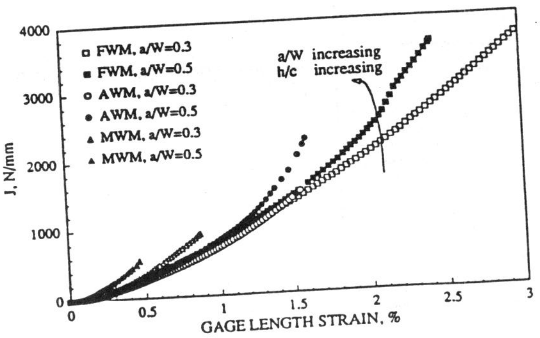


Fig. 28 Effect of mismatching, a/W and h/c ratios on the J vs. strain relationship for same specimens as in Fig. 26. (44).

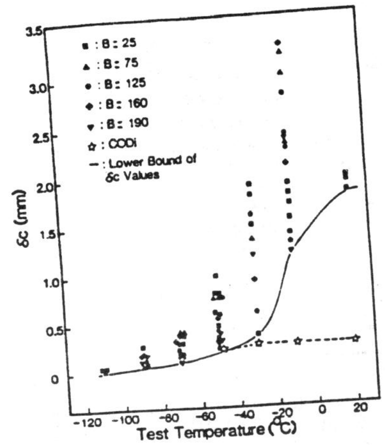


Fig. 29 CTOD fracture toughness as a function of temperature (48)

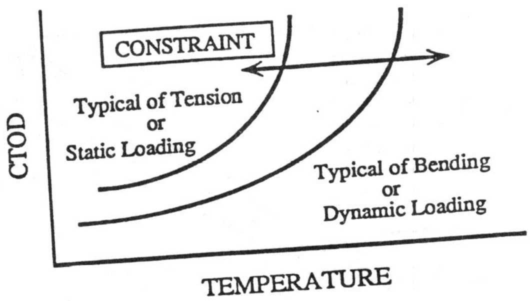


Fig. 30 Effect of constraint on the ductile-to-brittle transition.

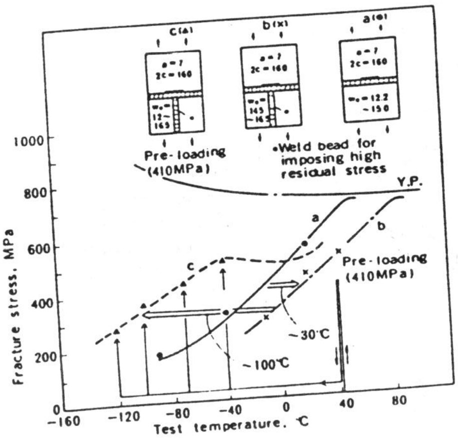


Fig. 31 Effect of welding residual stress and pre-loading (51).

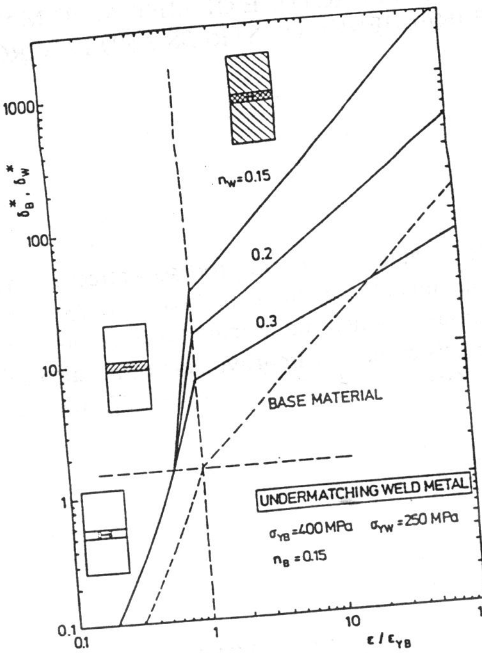


Fig. 32 Normalized CTOD as a function of the applied strain normalized by the base material yield strain for a specific case of undermatching (58).

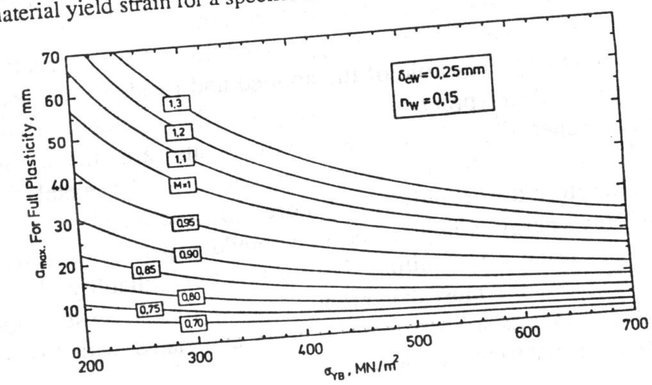


Fig. 33 Maximum crack length at attainment of full base material plasticity (58).