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Correlation of modified crack tip opening distance with heat input to the heat affected zone of high-strength low-alloy steels

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Abstract

Investigated is the dependency of the crack tip opening displacement (CTOD) on the local microstructure of the heat affected zone in high-strength low-alloy (HSLA) microalloyed steel. Since the initiation of the crack tip location could not be controlled in fatigue, any possible correlation between heat input in welding and fracture toughness could be smeared. Modified CTOD data are defined; they show that the fracture resistance of the weld joint decreased as the heat input increased.

1. Introduction

Materials used to construct offshore structures [1] need to be qualified by CTOD tests for weld joints. For plates over 50 mm thick, high thermal gradients could prevail in the HAZ and develop strain states that are conducive for crack growth. The vulnerability of such situations deserves attention.

Conducted in this work is a series of CTOD tests of the HAZ of a high-strength low-alloy (HSLA) microalloyed steel whereby the heat input in welding is varied. One of the objectives is to determine any possible correlation between the heat input and measured value of the CTOD, δ. While approval of the weld joint is based only on having $\delta > 0.20$ mm, it would be more informative to determine whether the fracture toughness of the HAZ would be affected by variations in heat input.

Such a correlation, however, may be hidden by the microstructure variations depending on the heat input within the HAZ which is only a few millimeters wide. That is, any dependency between δ and heat input can be easily obscured by the uncertainty of the crack tip location in the HAZ which changes from specimen to specimen.

A modified CTOD, say δ^*, is introduced where the change in the local microstructure is reflected by the extent of subcritical crack growth prior to the onset of rapid fracture, the threshold of which can be identified by δ_c. The fracture toughness is, therefore, referred to a particular microstructure. A correlation between δ^* and heat input thus appear to prevail when the data are plotted graphically.

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2. Experimental procedure

2.1. Material

The base material is a HSLA microalloy steel in accordance with BS 7191 450 EMZ type 2 standard [2]. The chemical composition is given in Table 1. The steel is full-killed, fine grained, quenched and tempered; it has a yield strength of $\sigma_y = 434$ N/mm^2 and a tensile strength of $\sigma_u = 545$ N/mm^2.

2.2. Weldment

A K-shaped butt joint is welded by a submerged arc process. The dimensions are 980 mm (weld length) × 750 mm and 75 mm thick as shown in Fig. 1. The temperature was kept around 115°C. Heat inputs are from 1.348 to 2.326 MJ/m. Three specimens were used for the CTOD test and one was kept as reserve.

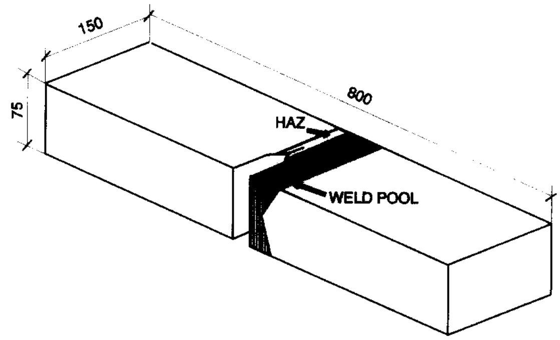

Fig. 1. Schematic of K-shaped butt weld joint.

2.3. Metallography before fracture

Two sections at a distance of 50 mm from the weld ends were removed and prepared for metallographic observation; they were polished and etched in 2% nital. The grain size in the HAZ adjacent to the weld was determined over the central two-thirds portion of the thickness. This is done according to ASTM E112 [3]. The object is to align the notch in the HAZ.

2.4. CTOD test

A single edge notched (SEN) three points bend specimen was used for the CTOD test as recommended by BS 5762 [4] with the cross-section as $B \times 2B$, where B is the thickness.

All specimens were compressed locally by a total of 1% of thickness in the ligament area so as to maintain the shape of the fatigue flaw growing as evenly as possible throughout the thickness. The effect of residual stresses arising from the weld thermal cycle is thus minimized.

Displacement controlled tests were carried out at a temperature of −10°C. An expression for δ is obtained in terms of the yield strength of the HAZ; it is taken as the mean value of the base material and the yield strength measured for the weld metal from a tensile test. Values were correlated for −10°C test temperature [5].

2.5. Metallography after fracture

Fractured specimens were sectioned perpendicular to the fractured surface in front of the fatigue flaw. Grain sizes sampled along the crack tip were assessed. Different grain sizes, in percentage, were measured over the central two-thirds portion of the thickness.

The location of crack initiation was traced by optical and scanning electron microscopy. Slow stable crack extension (excluding stretch zone) prior to rapid fracture was also measured and averaged over the thickness.

Table 1
Chemical composition of HSLA microalloy steel

C	Mn	P	S	Si	Cu	Ni	Cr	Mo
0.097	1.360	0.009	0.002	0.403	0.153	0.512	0.123	0.008
Al	Nb	V	Ti	N	Cu + Ni + Cr + Mo	Nb + V + Ti	Al/N	Ceq.
0.032	0.016	0.001	0.003	0.007	0.797	0.020	5.0	0.394

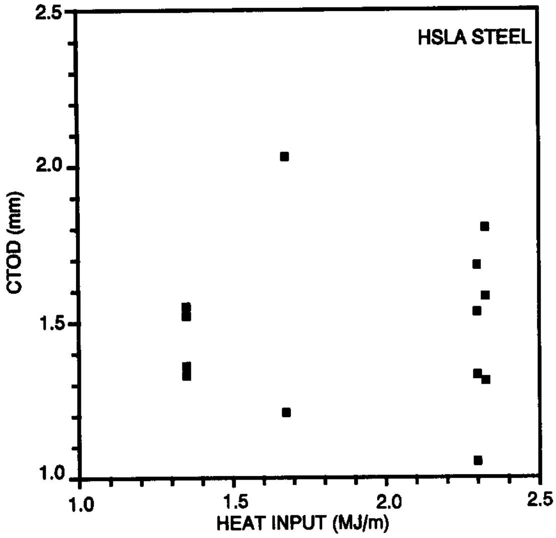

Fig. 2. Crack tip opening displacement versus heat input of weldment.

3. Experimental results

From the load versus clip gauge displacement records, values of δ have been calculated from the equation

$$\delta = \frac{(1-\nu^2)K^2}{2\sigma_y E} + \frac{0.4(2B-a)v_p}{0.8B+0.6a+z} \quad (1)$$

where E is Young's modulus, ν Poisson's ratio and σ_y the yield strength corrected for the test temperature. The crack length is a while z is the distance from the clip gage to the notch edge. In Eq. (1), v_p is the plastic displacement at the notch edge.

Plotted in Fig. 2 are the CTOD δ versus the heat input. The scattered data reveal no definite correlation. Fig. 3 gives the increase in CTOD with subcritical crack growth a prior to the onset of rapid fracture. A metallographic study was made to establish a relation between crack growth a^* and microstructure entity which corresponds to a local region of the HAZ heated from 723 to 870°C. This region is referred to as the intercritical HAZ, and in Fig. 4 the percentage of the fatigue crack front sampling this particular zone is shown.

4. Discussion of results

4.1. Microstructure effect

It is difficult to control the growth of the fatigue crack in the HAZ [6–8], which presents a problem in the consistent assessment of the CTOD. Standard practices [4] require the fatigue

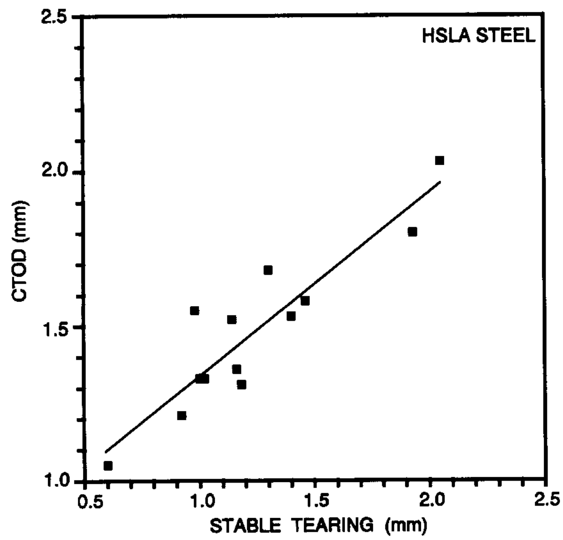

Fig. 3. Linear regression of CTOD versus subcritical crack growth.

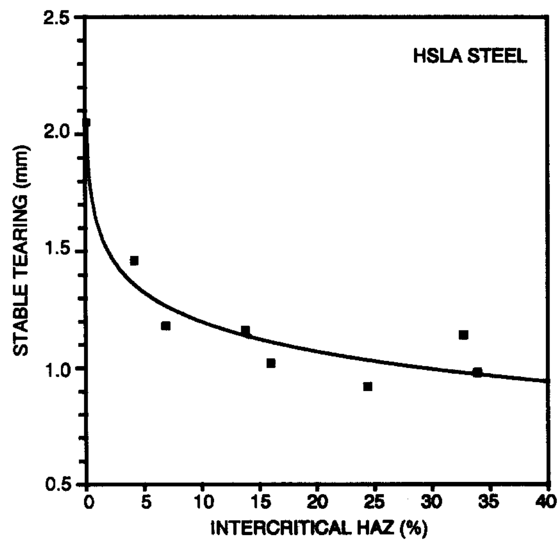

Fig. 4. Subcritical crack growth versus percentage of intercritical HAZ.

flaw to be within 0.5 mm of the fusion boundary through the central two-thirds portion of the specimen thickness. The grain coarsening zone of the HAZ can be sampled. This is potentially the region vulnerable to crack propagation. The actual location of the crack is not known after fracturing of the specimen and found by optical microscopy. Only 20% of specimens could satisfy the aforementioned requirement [9].

Research on HSLA steels [10,11] has identified regions that are more vulnerable to those with a coarse-grained band; they correspond to those with no increase of the average grain size and exhibit brittle behavior. The region is known as the intercritical HAZ; it experienced heating during the weld thermal cycle up to the α/γ phase of the steel. Pearlite colonies appear at the grain corners. The high magnification photo in Fig. 5 reveals a degenerate structure of ferrite

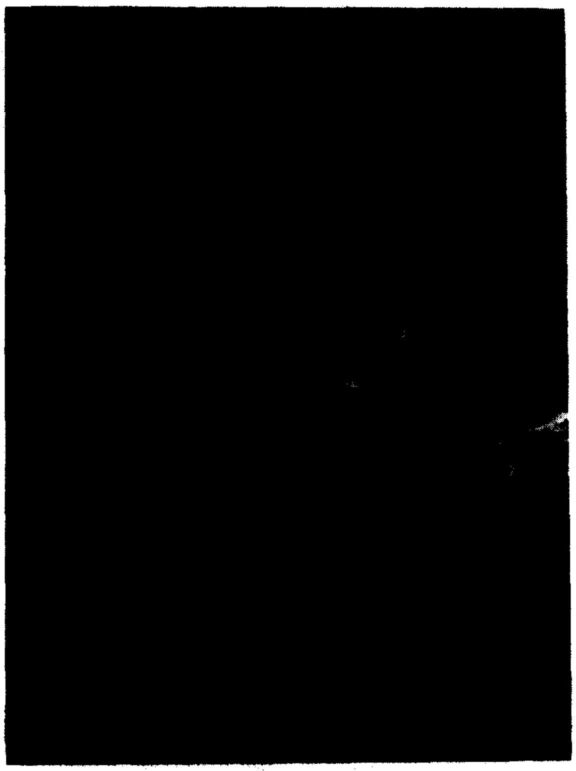

Fig. 5. Morphology of pearlite colonies in base material.

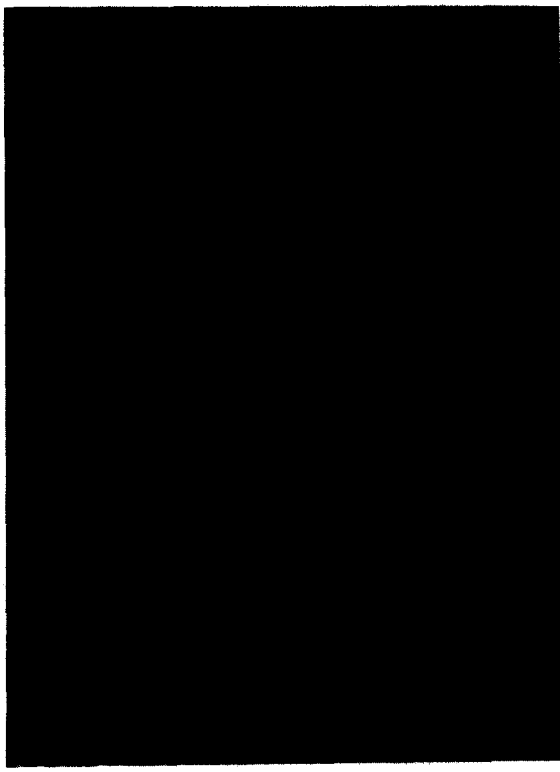

Fig. 6. Martensitic islands in intercritical HAZ.

and spheroidized cementite. In multi-run welds, one of these colonies is reheated by a second pass and transformation to austenite takes place. Subsequent cooling produces an "island" of martensite and/or bainite, as can be seen in Fig. 6. Both are very brittle constituents; their presence could have an appreciable influence on the outcome of the CTOD measurement.

The top limit of the heat input for these steels (3–4 MJ/m; typical value around 2 MJ/m) could be increased up to values of 8 MJ/m [12] with an increase in the deposition rate of 150%. Such levels of elevated heat input necessitate an understanding of the fracture behavior of the welded joint. The scattered data in Fig. 1, however, provide no definite trend between δ and heat input. The same results were arrived at [13] for the same test and similar material. To circumvent the problem of data scatter, a modified CTOD approach will be used.

4.2. Modified CTOD

Assume that the subcritical crack growth behavior is indicative of the material microstructure. Pop-ins have been identified with the crack tip location in the coarse-grained zone [14]. The event can also occur following subcritical crack growth. The farther the crack tip is away from the fusion line, the greater the amount of subcritical crack growth. In the same vein, increase in the area of brittle martensite islands tends to decrease the stage subcritical crack growth prior to rapid fracture, Fig. 3.

More specifically, consider a subcritical crack growth of $a_r^* = 1.00$ mm associated with a given reference microstructure. Fig. 2 gives a corresponding value of $\delta = 1.28$ mm. For a different test, the crack would encounter a different microstructure and hence $a^* = 2.00$ mm. A linear regression corresponding to a variation of 100% in a^* gives an increase of 72% in δ. The amount of 0.0072 mm must be added (or subtracted) from the measured δ in each test. That is, for each 0.01 mm that a^* differs (positive or negative) from the reference value of a_r^*. As an example, suppose that a test gives $\delta = 1.80$ mm and $a^* = 1.93$ mm, a modified CTOD δ^* would be

$$\delta^* = 1.80 - (93 \times 0.0073) = 1.13 \text{ mm} \quad (2)$$

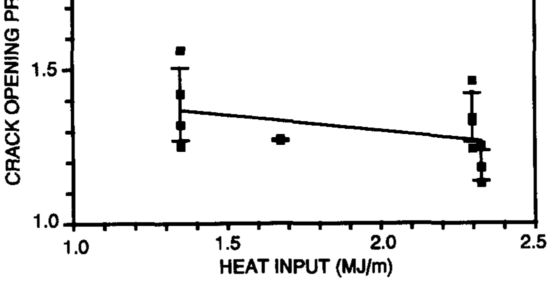

Fig. 7. Modified crack tip opening displacement versus heat input of weldment.

The same applies to the remaining tests. Fig. 7 shows a decrease of modified CTOD δ^* with heat input. This implies a loss of the weld joint resistance to fracture as the heat input is increased.

5. Conclusions

CTOD measurements are greatly affected by the location of the fatigue crack in the HAZ depending on the microstructure. Scatter of the test data between the CTOD δ and heat input is overcome by defining a modified CTOD δ^* that accounts for the effect of subcritical crack growth. The value of δ^* decreased with increasing heat input. This implies a reduction in the fracture resistance of the weld joint as the weldment heat input is raised.

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