HAZ TOUGHNESS OF Ti-MICROALLOYED OFFSHORE STEEL IN AS-WELDED AND SIMULATED CONDITION

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The CTOD values measured on 34 mm thick SENB specimens taken from multipass 1/2K joint were compared with the values obtained from 8 mm thick SENB specimens with simulated microstructures of CGHAZ (a/W=0.5). Single and double thermal cycles were used to simulate various thermal treatment which HAZ at the weld bond may experience during the welding.The CTOD fracture toughness testing of the simulated specimens can provide toughness values not affected by the mechanical heterogeneity (strength mis-match between weld and base metals) provided one can simulate the microstructure of interest correctly. The examinations of these simulated specimens shows the presence of the local brittle zones (LBZ) in spite of the steel was Ti-microalloyed. An attempt to correlate CTOD and Charpy impact toughness values on simulated microstructures was undertaken.

INTRODUCTION

It is well known that the coarse grained heat affected zone-CGHAZ region of many structural steel welds can be the least tough region of the weld joint. In the literature survey (1), a huge attempt can be noticed for improving the CGHAZ toughness of the modern microalloyed steels. The steel makers succeed to reduce the coarse grain size and the width of CGHAZ and hence improving the toughness properties by using Ti as microalloying element in mainly two mechanisms (1, 2, 3, 4).

Grain boundary pinning by uniformly distributed TiN particles, sized from 0.02 μm to 0.08 μm, but keeping the Nb and V contents low. Presence of further alloying elements serving as nitride formers generally tends to decrease the stability of TiN particles and increases the particle size.

Grain boundary pinning by uniformly distributed stable Ti_2O_3, sized from 0.5 μm to 3 μm but promoting also interferential nucleation of acicular ferrite as main beneficial effect.

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Although TiN is thought to be comparatively stable even at high peak temperatures, complete and/or partial dissolution (depending on the size and composition of the precipitate) can still be expected, since, TiN particles can occur in various sizes ranging from several microns to several hundred angstroms. However, a particle size smaller than 0.1 μm has been found to be effective in grain boundary pinning. Therefore, TiN can only be effective in suppressing grain coarsening in the HAZ if the method of welding, the ratio of Ti/N and the level or presence of other microalloying elements produce Ti precipitates of appropriate size and distribution.Ti-oxides are efficiently used in improving the toughness of steel and also in weld metal deposits due to their dual role, restricting the grain boundary migration and acting as nucleus for acicular ferrite formation (since they are more stable than TiN precipitates at higher temperatures). A sufficient amount of precipitates should also remain undissolved in the HAZ and these should act as pinning and ferrite nucleation sites. Hence, optimum numbered and sized fine TiN or Ti-oxide precipitates must be present if an improvement on HAZ microstructure or toughness of the zone adjacent to weld metal is expected to occur.

In the present paper, the measurement of HAZ fracture toughness of a SAW joint of 40 mm thick TiN treated offshore steel was undertaken. The fracture mechanics CTOD tests were conducted on 34 mm thick specimens taken from multipass 1/2K joint and the values were compared with the values measured on 8 mm thick SENB specimens containing different simulated HAZ microstructures.

The aims of this work were:

-to measure the toughness of different HAZ microstructures by using simulated microstructures by omitting the problem of crack/notch tip location in the specimens taken from actual joints,

-to compare the CTOD results of simulated specimens with CTOD values at stable crack initiation (CTODi) values obtained from full thickness specimens taken from SAW joint which was about 27% overmatching,

-to correlate the CTOD critical values δc and the CTOD values at the onset of stable crack growth δi with Charpy test values

-to discuss the possible effect of weld metal strength overmatching of a real weld joint on fracture behavior and on fracture toughness values.

EXPERIMENTAL DETAILS

Parent Material and Welding Procedure

The C-Mn steel used was in normalized condition and its chemical composition is given in Table 1 which contains the ratios of Ti/N and C/N. The steel has low C, S and is alloyed with Ni. The mechanical properties of the 40 mm thick steel and SA weld metal are given in Table 2. The plates were welded using Tandem SAW multipass procedure with a single bevel butt weld preparation as shown in Fig.1. The welding procedure is given in Table 3.

TABLE 1- Chemical Composition of the Steel StE 355Ti in %

Steel Type							
	C	Si	Mn	P	S	N	Al
	0.09	0.43	1.46	0.013	0.003	0.0071	0.046
StE 355+Ti	Cu	Ni	Ti	Nb	Ca	C/N	Ti/N
	0.12	0.44	0.016	0.022	0.001	12	2.25

P_{cm}=0.190, Ceq=0.370

TABLE 2- Mechanical Properties of the Parent Steel and SAW Weld Metal

Steel Type	σ_y* (MPa)	σ_u* (MPa)	Elong. l_5 (%)	Charpy V impact energy (J)* -10°C	-40°C	Mismatch factor M
StE 355	388	515	32.5	292	300	-
Weld metal	492	578	24.4	166	140	1.27

*Transverse direction

TABLE 3- SAW Welding Procedure

Tandem-SAW welding procedure	
Number of passes	10
Wire/flux	OE-SD3/OP121TT
Heat input	40 kJ/cm, $\Delta t_{8/5}$=40 sec
Preheating temperature	100°C
Interpass temperature	200°C

Thermal Simulation Procedure

The specimens for microstructural simulation of the HAZ region were extracted from the parent plate in the rolling direction with the dimensions of 8x15x70 mm. Various single and double thermal cycles were carried out to simulate different HAZ microstructures. The peak temperatures of the simulations are given in Table 4. Single cooling time ($\Delta t_{8/5}$=40 sec) was used for all thermal cycles which corresponds to the SAW condition.

HAZ Fracture Toughness Testing

The HAZ toughness of the multipass weld and simulated microstructures was measured using Charpy V-notch and CTOD specimens. The single edge notched bend (SENB) CTOD specimens were machined from the middle of the as-welded multipass joints in Bx2B geometry (B=34 mm). The reason was the misalignment of the joints. CTOD specimens were notched and fatigued through-thickness in the

HAZ and tested at -10°C. For the simulated microstructures, the SENB specimens were 8 mm thick

TABLE 4- HAZ Simulation Procedure Data for Single and Double Thermal Cycles

Specimen designation	First cycle Tp1 (°C)	Δt8/5 (s)	Second cycle Tp2 (°C)	Δt8/5 (s)
O	1380	40.2	-	-
A	1370	40.0	705	82+
B	1390	40.5	907	43
C	1380	40.0	960	42
D	1380	40.0	1025	41.5
E	1360	40.0	795	80+

+ instead $\Delta t_{8/5}$, cooling time $\Delta t_{5/3}$ was measured; $\Delta t_{8/5} \sim 1/2 \Delta t_{5/3}$

and were approx. Bx2B type. During the CTOD tests the DC potential drop technique was used for monitoring the stable crack growth (5).

Load line displacement (V_{LL}) was also measured with reference bar to minimize the effects of possible indentations of the rollers. Fatigue precracking was carried out with "step wise high R-ratio" (SHR) procedure for all specimens (6, 7). This technique is successfully used at GKSS Research Center for as welded specimens to obtain a uniformly shaped fatigue precrack. The SHR technique uses two stress ratios, R=0.1 for crack initiation and growth of about 1 mm then, stress ratio of R=0.7 with the allowable maximum load, until the required a/W ratio is obtained. The CTOD values were calculated in accordance with BS 5762 (δ_{BS}) (8) and in the case of real weld specimens also directly measured with GKSS developed δ_5 clip gauge on the specimen's side surface at the fatigue crack tip over gage length of 5 mm (9).

RESULTS

CTOD Results

The CTOD results (8) obtained from SENB specimens extracted from multipass welds and simulated microstructures are given in Fig. 2. The critical values of CTOD and CTOD data for the initiation of the ductile tearing (δ_i) are shown in Fig. 3 (see also the explanation of symbols in the tables 4 and 6). δi value is defined as the value of CTOD for the crack growth of 0.2 mm in accordance with the ESIS procedure (10). In these figures, the CTOD values of the HAZ multipass weld joint, (F), can be compared with the results of the simulated microstructures (O-E). After CTOD testing, the post-test sectioning and microstructural examinations were conducted for all specimens to identify the fatigue crack tip microstructure and location of the initiation.

Simulated and Real Multipass HAZ Microstructures Impact Toughness Testing

The Charpy impact toughness for the simulated microstructures are presented in Table 5a. Fracture appearance transition temperature-FATT and the maximum hardness values obtained for each microstructure are included in Table 5b. The same data obtained from multipass weld HAZ are presented in Table 6.

TABLE- 5a Charpy Toughness of Simulated Microstructures

Specimen designation	Energy (J)* -60°C	-40°C	-20°C	0°C	20°C	40°C	60°C
O	-	13	20	68	176	-	-
D	17	38	128	-	210	253	-
C	18	23	85	-	186	220	-
B	22	33	119	-	197	-	-
E	-	11	32	-	126	152	-
A	-	13	13	-	55	-	135

*average of three specimens

TABLE- 5b FATT, Hardness, Shift Temperature, CTOD Transition Temperature and Calculated Critical CTOD Value

Specimen designation	FATT (°C)	Hardness HV10	ΔT (°C)	FATT-ΔT (°C)	δc at FATT-ΔT (mm)
O	+11	213	31	-20	0.13
D	- 8	210	38	-46	0.15
C	0	204	41	-41	0.14
B	- 6	208	39	-45	0.15
E	+29	225	25	+ 4	0.14
A	+43	230	24	+19	0.10

TABLE- 6 SA Weld Joint HAZ Charpy Impact Toughness and Hardness Values

Specimen designation	Location	Energy (J) * -10°C	Energy (J) * -40°C	Hardness HV10
F	Cap layers	197	149	189
	Middle layers	191	156	-
	Root layers	179	103	179

* average of three specimens

DISCUSSION

Charpy-V Test Results

It is clear that high Charpy-V impact toughness values of the real HAZ at the weld bond (Table 6) are the average toughness of several microstructural regions due to relatively large machined notch tip radius where more ductile areas of HAZ also contribute to the deformation and fracture. This implies that the Charpy-V test produces unreliable results if quantitative CGHAZ/LBZ toughness of various multipass welds is going to be assessed. This can be proved by Charpy impact toughness values obtained from simulated specimens with uniform microstructures (Table 5a). Different HAZ regions represented by various thermal simulation procedures exhibit different toughness and hence varying sensitivity for LBZs appearance at the testing temperature. It has to be pointed out that the cause for difference of real multipass and simulated HAZ impact toughness can not be the deviation of cooling time. The analyzed dependance of impact toughness and cooling time shows only slight change in the range of 30 - 50 sec.

CTOD Test Results

The standard CTOD fracture toughness results (Fig. 2) show much higher toughness values (F) for the specimens extracted from multipass weld joint if they are compared with the values of simulated specimens. It was expected to obtain similar or even some better toughness values by conducting measurement on the six different types of simulated microstructures. But the measurement on the thick SENB specimens of multipass weld joints did not show any low CTOD values. On the other hand, comparison of the "intrinsic" fracture toughness values (crack initiation at Δa=0.2 mm) obtained on both specimen types is better (Fig. 3 - B, C and D), due to its size independent nature. But even in this case, the fracture toughness of real SA weld joint is slightly higher than in all simulated cases. The reason is full sampling of the CGHAZ microstructural constituents in simulated specimens (100%) compared to full thickness CTOD specimens extracted from real weld joints (approx. 20-30%), despite of higher constraint and overmatching effect in the latter, which should lower the CGHAZ fracture toughness. The lowest fracture toughness values, have appeared in the simulated unaffected coarse grained (UCG) HAZ - single cycle microstructure designated by O. The second thermal cycle applied between AC1 and AC3, (A) and below AC1, (E), did not improve the toughness of the UCGHAZ. The second thermal cycle above AC3, (B, C and D), as expected, have improved the fracture toughness due to refinement of the UCGHAZ, but the fracture toughness level of the real weld HAZ was still not reached.

It can be concluded that the HAZ fracture toughness measured is highly effected by crack tip microstructure. The lowest fracture toughness can be obtained by positioning the crack mainly into the CGHAZ microstructure. If this case is combined with the highest constraint condition the cleavage crack initiation can occur from the CGHAZ of the real weld joints by dislocation piling up mechanism suggested by RKR local fracture criterion (11, 12). It is evident that the estimation of "intrinsic" fracture toughness on simulated specimens for different HAZ regions can be rather informative to control the LBZ susceptibility of the steel even if the CTOD results do not indicate any embrittlement in the multipass weldments.

The fracture behavior of the LBZs can be influenced by the strength mismatching of the weld joint. High strength and tough weld metal can provide a possibility for the HAZ notched CTOD specimens to fracture in unstable fashion. If the fatigue crack tip is located in the vicinity of the CGHAZ (in the overmatched weld metal having good toughness), the brittle crack can still initiate at the CGHAZ under the influence of the strength mismatching, as shown in Fig. 4. Higher strength in weld metal side of the specimen will not allow the plastic zone at the crack tip to develop, because of softer base material and the consequence is constraint raising at the CGHAZ. Consequently, the CGHAZ reaches the critical condition at lower nominal stress/load and therefore fracture may predominantly initiate and remain at the brittle CGHAZ as shown in Fig. 4. Fracture behavior of the CGHAZ should also be examined in terms of mechanical heterogeneity of the weld joint since, identical CGHAZ microstructure can give substantially different toughness values (i.e. apparent) if one varies the weld metal strength mismatch.

CTOD Fracture Toughness and Charpy Impact Toughness Correlation

In general, for medium strength steels and medium thicknesses, the Charpy transition curve moves relatively to the higher temperature compared to that of the CTOD.The Charpy transition temperature is defined by FATT. In the case of CTOD, the transition temperature is assumed to be the temperature at which the critical CTOD value (δc) becomes equal to the critical value for the onset of stable crack growth (δi). At this temperature the fracture nature is changed. The difference of two transition temperatures is sensitive to the strain rates and notch acuity of the impact Charpy and CTOD tests and yield stress of the material and thickness of the specimen respectively (13).

$$\Delta T = 133 - 0.125\sigma_y - 6B^{1/2} \quad (1)$$

ΔT is the transition temperature shift, σy the yield stress of the simulated microstructure and B the thickness of the CTOD specimens with simulated microstructure. The critical value of the CTOD at the temperature T equal FATT-ΔT is the maximum possible δc since at higher temperatures fracture nature is no more brittle, therefore stable crack growth has to commenced. So, the conservative value of δc at the temperature FATT-ΔT is connected with the Charpy value vE at FATT.

$$\delta c_{(T=FATT-\Delta T)} = 0.001 vE_{(FATT)}. \quad (2)$$

δc values calculated by relation (2) using Charpy data from Table 5b are shown in the same table. The calculated values are compared in Fig 5 with those measured on the CTOD specimens for all six simulated microstructures. It is obvious that the dash line of equal CTOD values separates quite well the data for measured δi and data for measured δc and δu. It is assumed that δi at the temperature above transition temperature has to be at most equal than δc at the transition temperature. Generally, it is higher. Due to only one testing temperature of CTOD tests (-10 C°), measured data of δc have to be equal or lower that calculated values. Measured values of δc on CTOD specimens at -10°C are lower or approx. equal than δc at transition temperature, values of δi are higher. Therefore, it can be concluded that the correlation is quite satisfied.

Metallographical Results

The TEM examination of the steel plate microstructure revealed that Ti-rich precipitates are too small (below 22 nm) to be fully effective in grain boundary pinning process (14). Therefore, single cycle thermal simulation produced rather coarse microstructure which consists mainly ferrite with second phase and some primary ferrite. Transcristal cracking adjacent to fracture surface of the single cycled microstructure was the consequence of it. The SEM examination of the specimen A and E clearly show transcristal brittleness in the double thermal cycle microstructures which contain elongated ferrite side plates and some undecomposed M/A constituents as a second aligned phase. The presence of the side plates with aligned second phase instead of beneficial acicular ferrite in both real weld and simulated HAZs can be the consequence of the presence of Nb, which increases the CGHAZ hardenability and formation of TiNb(N) with lower solubility temperature than pure TiN. In order to enhance the formation of acicular ferrite in CGHAZ, the Nb and V content should be kept low. On the other hand, the ratio of C/N should be also kept sufficiently low to avoid formation of carbon rich precipitates with lower solubility temperature.

Fracture surfaces of the simulated specimens O, A and E show the absence of crack tip blunting and ductile tearing and hence transcristal cleavage fracture with some intercristal portion appeared. The fracture facets were coarse (about 130 μm), while the fracture facets of simulated specimens B, C and D showed size about 20-30 μm and exhibited large amount of stable crack growth prior brittle fracture.

CONCLUSIONS

An experimental programme to compare the fracture toughness of the CGHAZ of the multipass SAW joint and various simulated HAZ microstructures at the weld bond on Ti-microalloyed StE 355 steel has been carried out. The results of this study can be summarized as:

In reviewing the respective literature, various studies show that an increase of alloying elements generally causes a deterioration of weldability and CGHAZ toughness properties if the interrelationship between the elements is not finely balanced. The effect of the Nb and V in Ti-microalloyed steels still requires further attention.The size and distribution of the Ti-precipitates play an important role in the grain growth control and hence the toughness.

The CTOD fracture toughness of the CGHAZ of the multipass weld and simulated CGHAZ can be rather different. This is due to the amount of CGHAZ sampled and mismatching of the real joint. The initiation of the ductile tearing values (size independent intrinsic toughness values) obtained from both specimen types are more close to each other.

Thermal simulation procedure provides simple screening test with respect to the embrittlement of the steel in HAZ at the bond, but provides very conservative CTOD toughness values. However, the results on simulated specimens show that even in the case of Ti microalloyed steel, the HAZ LBZs can develop if the particles size and distribution in the steel is not optimum (between 0.02 - 0.08 μm).

If the Charpy transition curve and mechanical properties of the CGHAZ are available one can quite good asses the fracture toughness value and CTOD transition temperature for further practical use.

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Figure 1 SA weld cross section

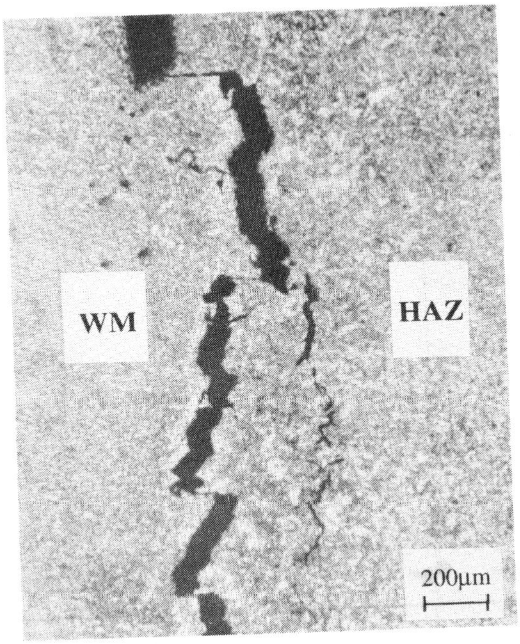

Figure 4 Fatigue crack tip in the overmatched weld metal but brittle fracture initiates and reamains at the CGHAZ during the CTOD test

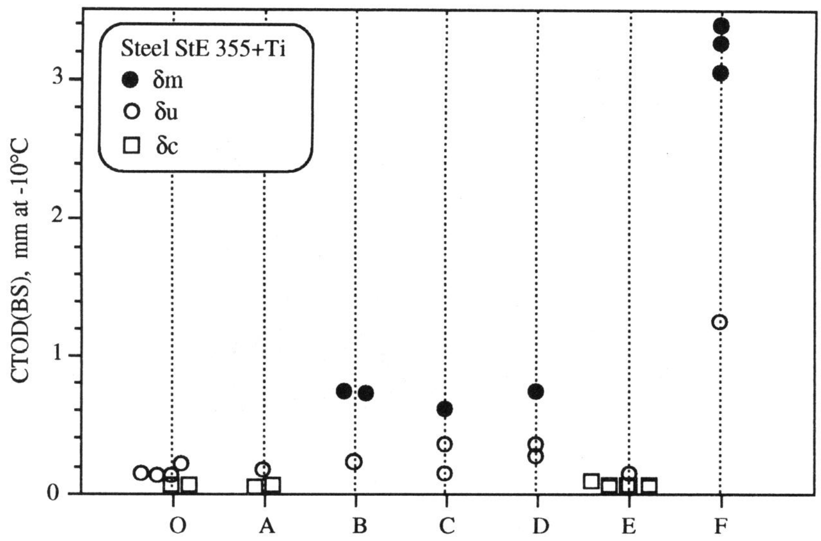

Figure 2 HAZ "apparent" fracture toughness of simulated and SWA microstructures

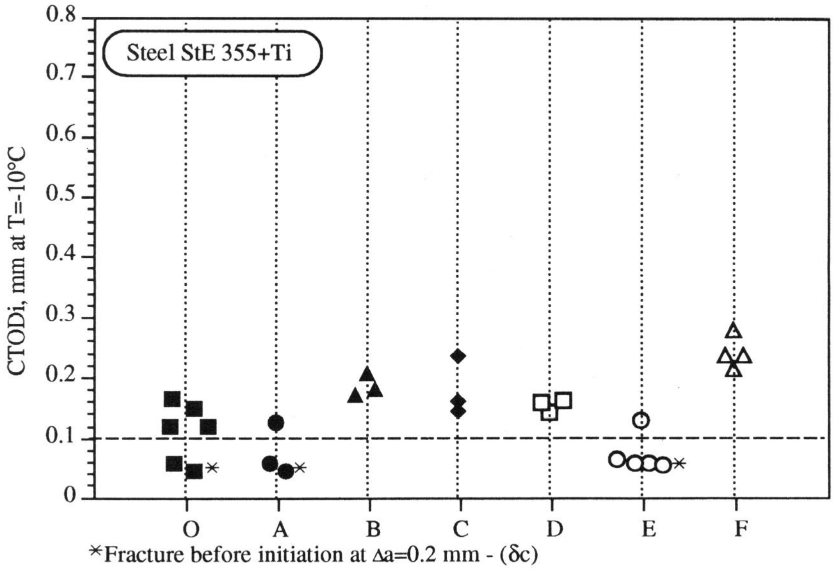

Figure 3 HAZ "intrinsic" fracture toughness of simulated and SAW microstructures

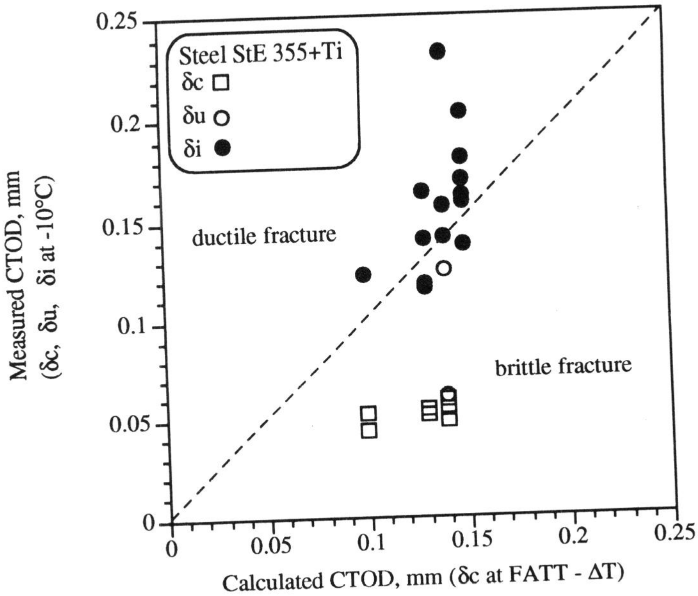

Figure 5 Comparison between measured and calculated CTOD values