MICROSTRUCTURE EFFECTS ON THE TOUGHNESS OF FCAW-FERRITIC WELD METAL.

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As part of a marine technology program weld joints were produced by flux-cored arc welding including hyperbaric underwater welding. Weld deposits with different oxygen, nitrogen and manganese contents were studied in terms of microstructure and toughness. With increasing nitrogen content, the CVN toughness is drastically reduced due to increased amounts of retained martensite/austenite, which causes cleavage initiation. On varying the oxygen content from 200ppm to 850ppm the CVN toughness passes through a maximum at medium levels. Fracture toughness tests (CTOD) show similar tendencies. Complex variations in the microstructure are responsible for this effect. Cleavage fracture is initiated by inclusions preferentially in large ferrite zones. Recommendations are made for improved hyperbaric welding.

INTRODUCTION

One of the essential design requirements for steel welded joints is the attainment of sufficient toughness in order to avoid failure by brittle fracture. Aside from geometrical factors such as the size and shape of the joint, and the thickness of the material, the microstructure of the weld metal and the heat affected zone can strongly influence the fracture toughness. The morphology and size distribution of the ferrite, as well as the microphases present (martensite/austenite, carbides, inclusions) have been shown to be important factors in determining the weld metal mechanical properties /1-12/. In such complex microstructures it is not surprising that there are still open questions as to how elements like oxygen, nitrogen, manganese etc. influence the microstructure and toughness, especially in the ductile/brittle transition.

This study is part of a marine technology program at the GKSS Research Centre and is directed towards the development of hyperbaric welding for underwater repair and construction work with special attention to toughness properties. Several test series were carried out at pressures from 1 bar to 30 bar (= 300m water depth). Most of the tests were performed with a C-Mn wire and some with a C-Mn-1%Ni-wire. Various $(Ar + O_2)$-shielding gases with oxygen contents from 0.1% to 10% were used.[2] A pipeline steel X65TM served as base material. In the first welding series nitrogen served as pressurizing gas. Since ineffective shielding at high pressures led to nitrogen ingress into the weld metal, argon was subsequently used for pressurizing. Based on preliminary tests the study concentrated on investigating mainly the effects of nitrogen, oxygen and manganese/nickel on the weld metal microstructure and toughness. Therefore first with the help of a detailed microstructure and fracture analysis including electron microscopy (TEM + SEM), an attempt is made to show the effect of different microstructural constituents on the fracture mode and thus to explain the influence of the above elements (N, O, Mn) on the toughness. In addition, some recommendations for the selection of welding parameters, shielding gases and consumables for hyperbaric welding are given.

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EXPERIMENTAL METHODS

Welding was performed in an unmanned pressure chamber using automatic MIG/MAG welding equipment. The heat input was maintained between 13KJ/cm and 17KJ/cm.
Weld metal specimens were tested in the as-welded condition. Tensile tests on cylindrical specimens and Charpy-V-notch tests on specimens taken from the middle of the plate with the notch perpendicular to the plate surface were carried out. An elastic-plastic fracture mechanics analysis was performed on CT-specimens with thickness 10mm and width 50mm. J-R- and δ-R-curves were determined using the DC potential drop method for measuring the crack extension. Further details of the welding parameters, specimen preparation and test methods are given in Ref. /13/.

RESULTS

Chemical Analysis

The first test series with nitrogen as pressurizing gas revealed that at higher pressures the nitrogen content of the weld metal reached values of up to $\sim$0.06wt%, compared to 0.01wt% under normal atmospheric conditions (table 1, # 1 and 2); more data can be found in Ref./13/. The next welding series with argon as pressurizing gas avoided nitrogen ingress, and showed that increases in the welding pressure and/or the oxygen content of the argon shielding gas led to increases in the weld metal oxygen content (Table 1,# 1-7). A good correlation was found between the partial pressure of the shielding gas oxygen and the weld metal oxygen, details of which will be published elsewhere. The Mn-content was unaffected by variations in pressure, but decreased when using more active shielding gases. The C-Mn-Ni wire showed rather different behaviour, resulting in 2.1wt% Mn when welded at 1 bar compared to 1.6wt% Mn at 10 bar (table 1, # 8 and 9). The contents of other minor elements eg. Nb, V, Ti, Al were little changed for the different conditions, for values see Ref./13/.

Toughness

The toughness results are presented in figures 1 and 2 and in Table 1. Comparison of the two weld metals with different nitrogen contents indicates that the room temperature toughness drops severely from 169J at 0.01wt% N to 26J at 0.055wt% N (Table 1, #1 and 2). Increases in the weld metal oxygen content from 430ppm to 720ppm caused by an increased pressure (shielding gas Ar+5% O_2) led to slightly reduced toughness values (figure 1a). The upper shelf energy decreases from about 170J to 145 J, and the transition temperature at 50% of the upper shelf energy increases from -38°C to -27°C. Using a leaner shielding gas in an attempt to reduce the detrimental oxygen effect on toughness at higher welding pressures showed that with increasing pressures (the corresponding oxygen contents are 240ppm and 550ppm), a slight toughness reduction is again observed (figure 1b).
However, the expected improvement of the toughness level due to the lower oxygen content was not observed; the toughness remains basically unchanged. Apparently oxygen variations produced by higher pressure give different results to those caused by different shielding gases.
At high oxygen contents e.g. welding with Ar+10% O_2 at 10 bar giving 850ppm oxygen, the toughness is further reduced significantly (figure

1c). Under these conditions considerable Mn burn-off occurs (Table 1, # 7). At very low weld metal oxygen contents of 200ppm (Table 1, # 6) a drastic shift of the transition temperature by 58°C to higher values occurred by comparison with a medium oxygen level (Table 1, #1)(figure 1c). The upper shelf energy is lowered by $\sim$35J.

Two results for the C-Mn-1%Ni-wire are presented with different Mn-contents of 2.1wt% and 1.6wt%. The oxygen contents are at a medium level. The toughness results in figure 1d show lower CVN-values for the higher Mn-content.
Fracture mechanics tests have only been carried out so far for a limited number of specimens having medium oxygen levels. The results are summarized in figure 2. The values of the crack tip opening displacement are plotted versus test temperature for two oxygen contents (430ppm and 720ppm, Table 1, #1 and 3). For a lower oxygen content (330ppm) δ-values at -20°C are also shown. δ_i is the initiation value for stable crack growth, δ_c the value of δ when unstable fracture occurred before reaching the load maximum, and δ_m is the load maximum δ-value. Both the δ_i and δ_m-values are not significantly affected by the variations in oxygen content studied here. There is however an oxygen effect on the ductile-brittle transition resulting in a shift to higher temperatures with increasing the oxygen content from 430ppm to 720ppm. The transition temperatures for 50% of the δ_m-values for example are -68°C and -50°C. The corresponding values from the Charpy tests are -38°C and -27°C.

Microstructure Analysis

The microstructures were evaluated in terms of the amount of different ferrite morphologies (Acicular ferrite (AF), proeutectoid ferrite (PF), side plate ferrite (SP) and microphases (M/A, inclusions). In addition, the prior austenite grain size was determined, as well as the area fraction of as-deposited and reheated zones along the notch and precrack, respectively, of the specimens. Typical micrographs illustrating the various constituents are shown in figures 3 and 4.
The variations in weld metal oxygen content had the following effects:
1) The oxygen forms oxide/silicate inclusions which based on EDX-analysis contain Al, Si, Mn, Ti, S and Ca. With increasing oxygen content a higher inclusion volume fraction was observed; the average inclusion diameter (0.4μm) did not vary significantly (Table 2).
2) The prior austenite grain size increases when reducing the oxygen level (Table 1).
3) The relative amounts of AF, PF and SP vary only slightly giving less AF and more PF and SP when increasing the oxygen content from 250ppm to 720ppm. At high (850ppm) and low (200ppm) oxygen levels a significant reduction in AF and increases in PF and SP are observed (figure 5). The amount of M/A is between 1% and 2% and only slightly higher for the low oxygen condition.
4) The area fractions of as-deposited and reheated zone remain basically unchanged with increasing oxygen contents resulting from higher pressures.

However, the amount of reheated zones increases with increasing oxygen contents resulting from the use of more active shielding gases.

The main result of the increased weld metal nitrogen content is a greater amount of M/A phase in the as-deposited material and of carbide aggregates (pearlite and grain boundary carbide films) in the reheated zones. An increase in the nitrogen content from 0.01wt% to 0.055wt% raises the M/A area fraction from 1.5% to $\sim$12%. The M/A phase was identified by TEM

and its volume fraction was determined by optical microscopy using a special etch. Two M/A-morphologies were found, one within the AF mainly at grain boundary triple points (figure 4a), the other between the SP (figure 4b).
The main effect of the different Mn-contents in the C-Mn-Ni wire was a marked increase of M/A phase from 1.5% to ~10% when comparing 2.1wt% Mn and 1.6wt% Mn.

Fractography

Fracture surfaces were studied by scanning electron microscopy; in addition some specimens were split perpendicular to the fracture surface so that crack initiation could be studied.
In the upper shelf region a dimple type fracture occurred. Cleavage initiation took place preferentially within the coarser ferrite regions e.g. grain boundary ferrite and large grains in the reheated zones (figs. 6a and 6b). A number of sources of initiation was observed. Firstly, cleavage can start from inclusions (figure 6c). These inclusions are either of the type described in the previous section, or they are in a few cases, Al-rich (probably Al_2O_3) or Mn- and S-rich(probably MnS). The minimum inclusion size for initiating cleavage was between 1µm and 2µm; too few measurements have been made, however, to give a reliable value. Secondly, the M/A phase can also initiate cleavage as shown in figure 6d and confirmed by EDX. The size of the M/A phase is between 0.5µm and about 10µm. A third initiation site was observed preferentially in the reheated zones, where decohesion of grain boundaries occurred and started a cleavage crack (figures 6e and 6f). SEM contrast and TEM observations suggest that the grain boundary carbide films are responsible for this effect. Generally the reheated zones show smaller cleavage facets than the as-deposited material.

DISCUSSION

The discussion will first focus on the effects of oxygen, nitrogen and to a limited extent of Mn/Ni on the microstructure, the fracture mechanism and the corresponding toughness. In addition, some conclusions for hyperbaric FCAW-welding will be drawn.

The toughness in the upper shelf is only moderately affected by the various microstructures. They all show a dimple type fracture controlled mainly by the inclusions and possibly by the M/A phase and carbides. These constituents can be treated as hard particles, and composition changes of the various elements which lead to increases in their volume fractions should reduce the toughness. The only exception is the low oxygen condition (200ppm) which produced a lower CVN-upper shelf value than the medium oxygen condition. The reason for this is not yet clear; probably to a certain extent also the matrix material with varying yield stress and work hardening behavior can affect the void growth process.
Much larger microstructure effects are observed in the ductile/brittle transition. An attempt is made to discuss the effect of the various elements on the toughness based on the observed cleavage mechanisms. The role of oxygen on the microstructure development has been studied frequently /6-10/; there are, however, still many uncertainties concerning the effects on the fracture process. Oxygen, present in the form of oxide/silicate inclusions, has two important effects: Firstly, high densities of inclusions lead to small austenite grain sizes (Tables 1 and 2)

probably through pinning of the grain boundaries. This subsequently favors the formation of grain boundary ferrite (PF) and sideplates (SP), (figure 5 and Table 2). Secondly, inclusions act as nucleation sites for acicular ferrite (AF) promoting its formation at higher oxygen levels. Both processes are competitive, and at high oxygen contents there is more PF and SP because of the smaller grain size; at very small oxygen contents a similar effect is observed, in this case because of a lack of AF-nucleation sites. Cleavage fracture occurs as a combined process of initiation, preferentially at inclusions, and growth through large ferrite zones (PF, SP) (figure 6c). An increase in the number and size of such areas which occurs at very high and very low oxygen levels should lower the toughness. When comparing the size of the initiating inclusions (>~1µm) with the average inclusion diameter (~0.4µm, Table 2), one can see that only a small fraction of the inclusions determines crack nucleation. The lower critical diameter observed corresponds well to a value (~1.5µm) derived by Tweed and Knott /14/. Thus not only the inclusion content but also its size distribution could be an important factor for the fracture process. On the other hand it appears as important to consider the influence of the crack length caused by cleavage at an inclusion, and this is determined by the surrounding ferrite dimensions. It is therefore not completely clear whether the oxygen effects on toughness are mainly through the higher inclusion density or the amount of large ferrite zones.

The discrepancy in the toughness results, when comparing figures 1 and 2 can be explained by the observation that in the case of the leaner shielding gas less reheated area was produced, while when varying the oxygen content by pressure (instead of shielding gas) the area fractions of as-deposited and grain refined material were roughly unaffected. Assuming that the reheated zones are intrinsically tougher (grain size effect) the lower toughness level for the leaner shielding gas can be understood. In addition, the variations in Mn-content have to be considered, since they can vary microstructure and toughness considerably /11/. According to Ref./11/ for example the difference in Mn-content between weld metal No. 1 and 5, which is caused by different Mn burn-off due to the different shielding gases, can cause a reduction in toughness for the higher Mn-content. In addition, for the condition with 850ppm oxygen the lower Mn-content compared to the other conditions has to be considered, since according to Ref./11/ a Mn-reduction from 1.4 wt% to 1.0wt% can in itself cause a reduction of AF and an increase in PF, which would presumably lower the toughness. These latter observations show that caution is necessary when interpreting oxygen effects, since direct effects may be masked by more indirect effects, such as the area distribution of as-deposited and reheated zones or variations of other elements.
A comparison of the Charpy toughness and the fracture toughness shows that variations in oxygen between 330ppm and 720ppm have no significant effect on the δ_m values (figure 2). This is not necessarily in contradiction to the CVN-results, since small effects are very difficult to detect in the fracture mechanics test; a scatter band of ±15% at least has to be considered. For the ductile/brittle transition similar shifts in CVN-values and δ-values are obtained. The transition temperatures themselves, however, are about 25°C lower for the δ -values. Further tests are necessary to show if a systematic correlation between fracture toughness and CVN toughness can be established.

The effects of nitrogen on the toughness can be explained by its effect as an austenite stabilizer promoting the formation of M/A phase. For our welding conditions no indications of nitride formation were found. The

retained austenite transforms to hard, twinned martensite upon plastic deformation as has also been shown by X-ray diffraction, and this hard martensite can initiate cleavage (figure 6d). The elongated morphology between the side plates could be particularly detrimental.
The high Mn-content in the C-Mn-Ni wire has been shown to lower the toughness, a result that was also found in a recent study by Taylor et al./12/. According to our observations this is due to an increase in the M/A phase with similar effects on the fracture process as for the nitrogen. The question arises as to why 10% M/A in a higher nitrogen content weld has a more embrittling effect than 10% M/A caused by a high Mn-content, although chemical composition and the rest of the microstructure are similar. It appears that the M/A in the high nitrogen weld metal is coarser and has a tendency to form bands of particles. This suggests that the morphology, size and distribution of the M/A phase must also be considered when explaining its influence on toughness.

The role of the carbides (grain boundary carbides and pearlite) has not been studied extensively so far; the carbides are present mainly in the reheated zones, with their distribution and morphology depending chiefly on the M/A distribution from which they form. Detrimental effects on toughness are here also to be expected (figures 6e and f).

Finally it should be concluded that studying the fracture process is a complex problem in this type of microstructure. Several factors must be taken into account e.g. the mixture of as-deposited and reheated zones, and also the volume fractions of the ferrite types and of the microphase, and their size and distribution. A quantitative interpretation of toughness results, resulting from a combination of all these factors, is therefore very difficult.
With respect to hyperbaric welding the results have shown that good toughness values can be achieved using flux-cored arc welding up to 30 bar. However, several sources of embrittlement have to be avoided. Firstly, effective shielding has to be ensured by a pressure-dependent flow setting to avoid nitrogen ingress. Secondly, the oxygen content of the weld metal has to be kept at an optimum level (~350ppm) by employing leaner shielding gases at higher pressures. Furthermore, with varying shielding gases the Mn-burn off varies, and in order to maintain the Mn-levels at about 1.3-1.5wt%, modified wires will be required.

CONCLUSIONS

A series of flux-cored arc welding tests with two wires (C-Mn,C-Mn-Ni) was carried out, including hyperbaric welding up to a pressure of 30 bar. Butt welds were produced with a constant heat input, but different ($Ar+O_2$)shielding gases. For the weld metal the following results were found:
1) Due to ineffective shielding weld metals with different nitrogen contents from 0.01wt% to 0.055wt% were observed. The oxygen content increased at higher pressures, and also when using oxygen richer shielding gas.
2) Increasing nitrogen contents lead to greater amounts of the martensite/retained austenite phase (M/A), and to a severe loss in CVN-toughness. The retained austenite transforms to hard, twinned martensite upon plastic deformation and initiates cleavage fracture in the ductile/brittle transition. Carbides from decomposed M/A can also initiate cleavage fracture. High Mn-contents (> 1.6wt%) in the Ni-containing weld metal like nitrogen produce increased amounts of M/A with a corresponding toughness reduction.
3) Variations in the oxygen content have a complex effect on the microstructure changing for example the inclusion density and by this the γ-grain size, the area fractions of acicular ferrite, proeutectoid ferrite and side plates. At high and low oxygen levels (200ppm and 850ppm e.g.) a CVN-toughness reduction is observed compared to a medium oxygen level (~400ppm). Cleavage fracture occurs preferentially in large ferrite zones (PF, SP) initiating from inclusions above a certain size. Fracture mechanics tests for medium oxygen levels show a higher transition temperature of CTOD at an increased oxygen content similar to the CVN-values, but with lower transition temperatures. The results are discussed on the basis of the observed microstructures.
4) For hyperbaric welding it is suggested that a) proper flow conditions for the shielding gas be ensured in order to avoid ingress of pressurizing gas b) leaner ($Ar+O_2$)-shielding gases be used at higher pressures in order to achieve optimum weld metal oxygen contents and c) suitable wires be developed to take account of the varying burn-off of Mn and Si under hyperbaric welding conditions.

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Table 1: Welding conditions and results (tensile and CVN-test chemical composition, microstructure)

No	Welding Shielding Gas	Press [bar]	Tensile Test Re_L [MPa]	Rm [MPa]	RA [%]	CVN-Test Upper Shelf [J]	Trans. Temp.(50%) [^oC]
1	Ar+5% O_2	1	538	610	74	169	-38
2	Ar+5% O_2	20	610	738	56	26J at 20^{o}C	--
3	Ar+5% O_2	29	517	601	73	145	-27
4	Ar+0.3% O_2	1	554	642	70	157	-31
5	Ar+0.3% O_2	29	578	665	71	150	-22
6	Ar+0.1% O_2	1	574	683	--	133	+20
7	Ar+10% O_2	10	416	499	74	130	-15
8	Ar+5% O_2	1	628	734	65	140	- 2
9	Ar+5% O_2	10	546	627	75	140	-16

No	Chemical Composition [wt%] C	Mn	Si	Ni	O	N	Microstructure AF [%]	PF [%]	SP [%]	M/A [%]	γ-grain-size [µm]
1	0.08	1.50	0.38	-	0.043	0.010	72.5	24	2.2	1.3	69
2	0.08	1.57	0.53	-	0.060	0.055	59	26	2.7	12.3	64
3	0.08	1.47	0.37	-	0.072	0.009	69	25	4.5	1.5	48
4	0.08	1.61	0.39	-	0.024	0.011	74	22.9	2.1	1.0	61
5	0.08	1.66	0.47	-	0.055	0.011	71	25.9	1.8	1.3	56
6	0.09	1.70	0.47	-	0.020	0.012	25	52	20.6	2.4	110
7	0.07	1.01	0.21	-	0.085	0.010	30	60	8.5	1.5	42
8	0.07	2.11	0.45	1.14	0.039	0.015	73	14	3.4	9.6	--
9	0.05	1.61	0.35	0.96	0.056	0.015	72	23.7	2.4	1.9	--

Table 2: Inclusion volume fraction and mean diameter for different weld metal oxygen contents

OXYGEN [WT %]	INCLUSIONS VOL. FRACT. [%]	MEAN DIA-METER [µm]
0,024	0,20	0,36
0,043	0,44	0,40
0,072	0,96	0,42

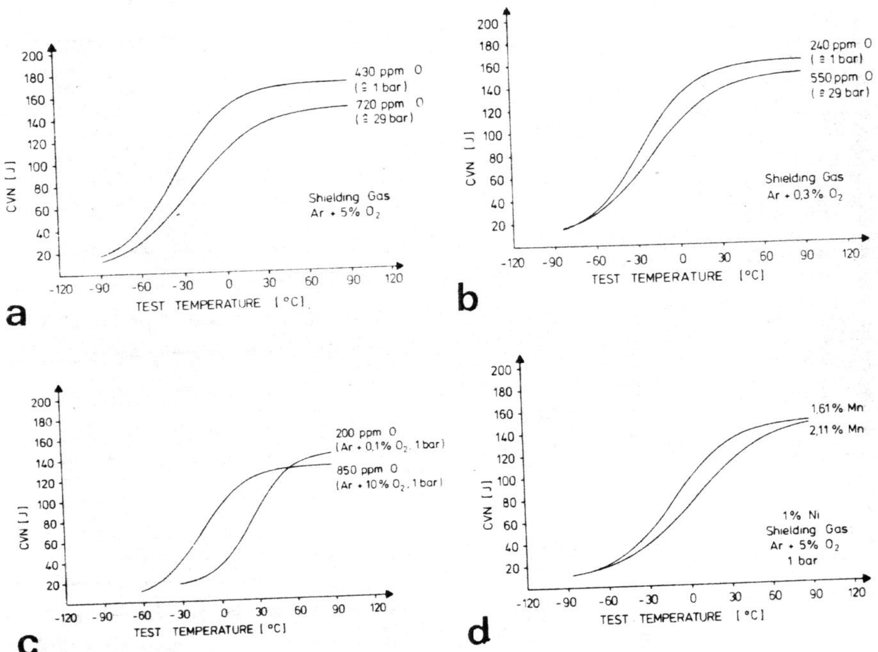

Figure 1: CVN-Toughness as a function of temperature
a) - c) C-Mn weld metal
d) C-Mn-Ni weld metal

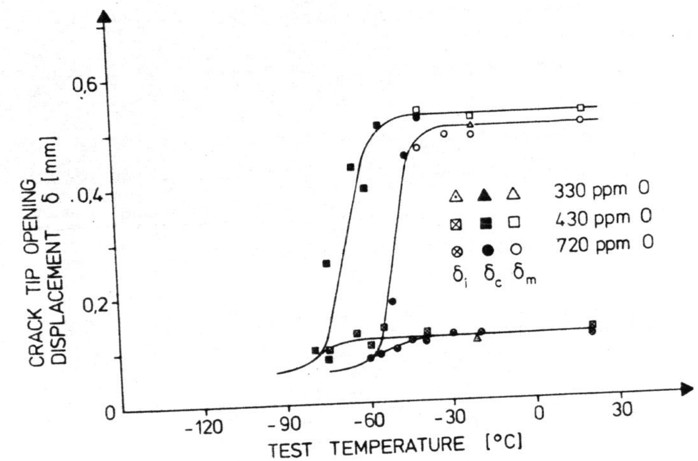

Figure 2: Crack tip opening displacement δ as a function of temperature for different oxygen contents, C-Mn weld metal

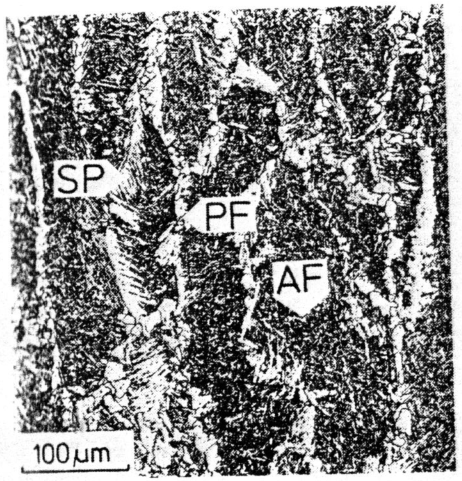

Figure 3: Optical micrograph illustrating acicular ferrite (AF), proeutectoid ferrite (PF) and sideplates (SP).

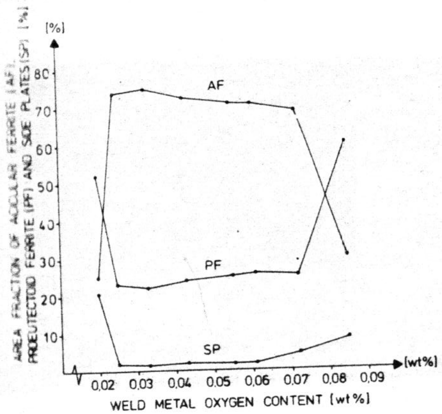

Figure 5: Area fractions of acicular ferrite (AF), proeutectoid ferrite (PF) and side plates (SP) as a function of oxygen content.

Figure 4: TEM micrographs illustrating the microphases
a) Martensite/austenite in acicular ferrite, inclusions
b) Martensite/austenite in side plates
c) Carbides

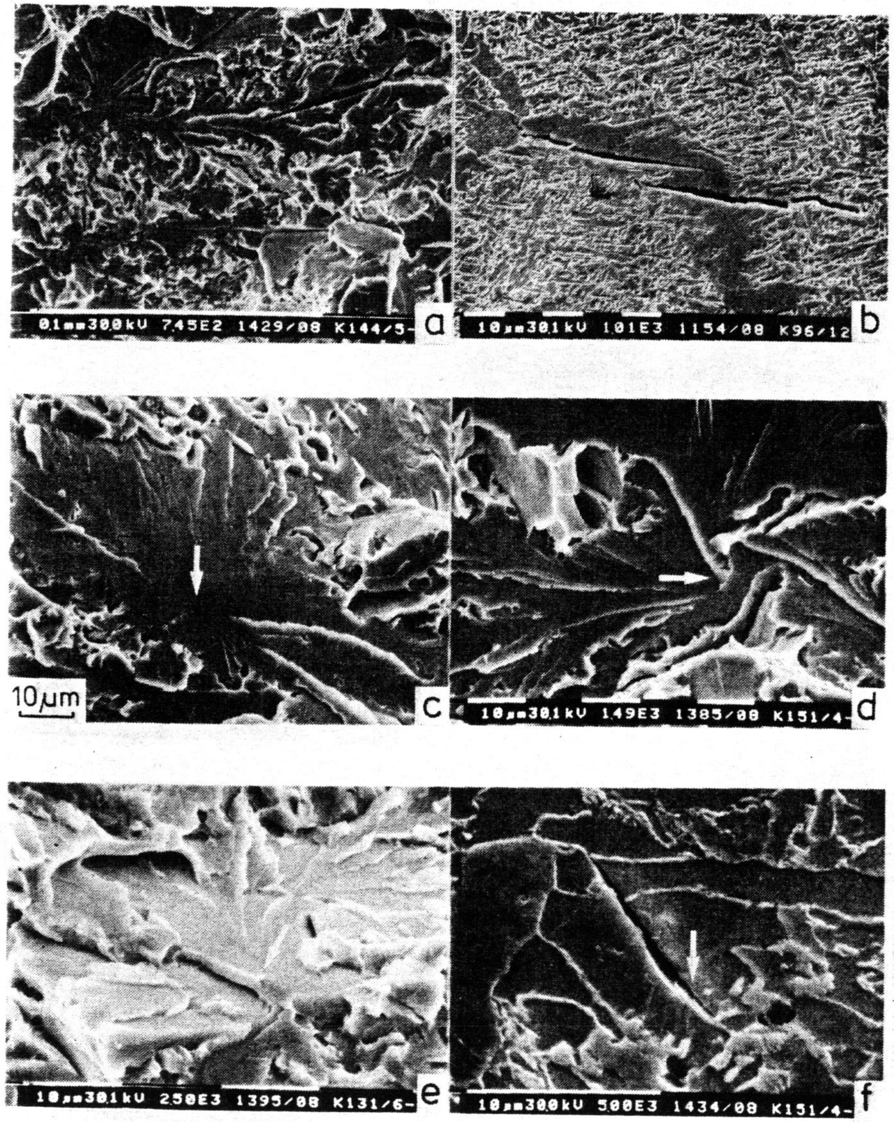

Figure 6: SEM-micrographs illustrating cleavage fracture
a) typical fracture surface
b) formation of cleavage cracks preferentially in grain boundary ferrite
c) crack initiation at inclusion
d) crack initiation at M/A phase
e) + f) crack initiation at carbides