

Recent Information on Weld Metal*

Introduction

In the wide international field of 'weld metal' research, the most significant to our mind has been the work concerned with its microstructure : how the microstructure is constituted ; how it is affected by chemistry and welding procedures ; what is the relation between weld metal microstructure and toughness. Of no less importance are the investigations relating to non-metallic inclusions : their extent, composition and effect on toughness. These aspects have been highlighted here.

Microstructure

Until 1975 there was no universally accepted standard terminology to describe weld metal microstructures, thereby making it difficult for investigators to compare their work with others. In that year, Widgery¹ categorised the microstructural constituents in as-deposited weld metal as : (1) proeutectoid ferrite ; (2) lamellar component ; (3) acicular ferrite ; and (4) other constituents (carbides, martensite, inclusions). Shortly later, Widgery and Davey² came out with a method of assessing the proportion of these various constituents which involved a systematic point count to determine the constituents found at each of 1000 points arranged in a two-dimensional array. This method and Widgery's terminology were adopted by Dr. Evans of Oerlikon in analysing the effects of various parameters on the microstructure and properties of all-weld metal deposits. The same terminology with minor modifications has been used in international research work on ferritic steel weld metal.

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Approach to toughness testing³

To establish true relation between weld metal microstructure and toughness and present data correctly in published work, one must consider the effect of particular variables such as, for example, weld metal chemistry on the resistance of specific mechanisms of fracture. In weld metal, two main fracture mechanisms are observed : (i) cleavage and (ii) microvoid coalescence. See Fig. 1. Resistance of a given weld deposit to these two mechanism determines the shape of the transition curve (toughness versus temperature). In order to arrive at correct solutions for improving weld metal toughness, factors controlling these mechanisms must be treated separately.

COD testing is considered to be best suited for investigating and distinguishing between the two fracture mechanisms because, firstly, it assesses resistance to initiation of fracture and, secondly, the toughness values can be readily linked to the microstructure at the specimen crack tip by metallographic sectioning and to the mechanism of initiation by fractography. In Charpy-V test, it is impossible to separate clearly the contributions of the two fracture mechanisms except at the extreme ends of the transition curve where 100% crystallinity or 100% fibrosity are developed.

Weld metal used in critical dynamically loaded structures such as offshore platform must have high resistance to cleavage since fracture initiation by this mechanism may lead to catastrophic failure of the structure or vessel. To-day data on factors controlling resistance to cleavage of weld metals is scarce ; on the other hand, information on factors governing microvoid coalescence is more readily available, since upper shelf values from Charpy tests reflect resistance to this mechanism of fracture.

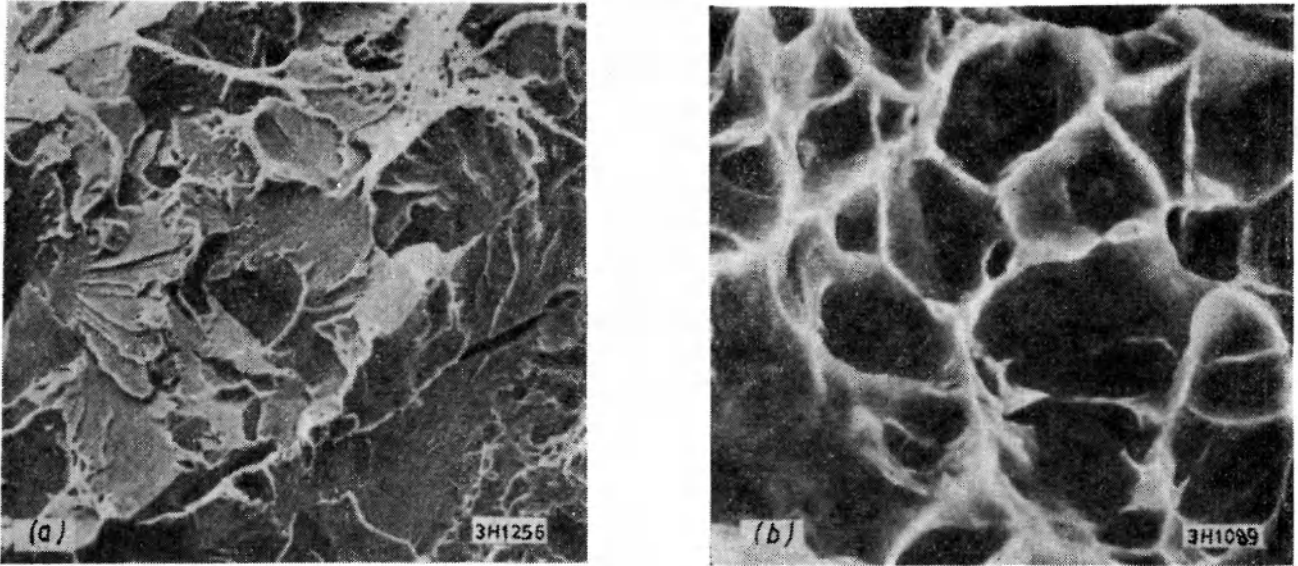


Fig. 1. Fracture mechanisms in weld metal (a) Cleavage ; (b) Microvoid coalescence

Resistance to cleavage

In C-Mn weld deposits, resistance to cleavage is improved and transition temperature lowered as the proportion of interlocking acicular ferrite is raised and that of proeutectoid ferrite and upper bainite is lowered. Also the finer the plate size of the acicular ferrite the better the toughness. In the heat input range of 2-4 kJ/mm, increasing the levels of Mn, Mo and Ni increases the proportion of this constituent and results in improved toughness. Once the level of nearly 90% acicular ferrite is reached, any further increase of Mn or Mo will promote brittle martensite regions and lead to a fall in cleavage resistance.

The effects of alloying elements Mn, Ni and Mo on the development of acicular ferrite microstructures and the resultant toughness are very much dependent on heat input rate which determines the cooling time from 800 to 500°C, i.e. Δt 800-500. Taking the example of Mn, at low heat inputs the increase in Mn necessary to form a given proportion of acicular ferrite would be much smaller. At high heat inputs (i.e. slow cooling rates), a substantial increase in Mn would be necessary to displace the proeutectoid ferrite and upper bainite transformation start curves sufficiently to form acicular ferrite. Thus in submerged-arc welding, changing from S2 wire (1% Mn) to S3 (1.5% Mn) or S4 (2% Mn) is likely to be beneficial at low heat inputs but detrimental at high heat inputs unless additional elements are used to achieve sufficient hardenability to produce acicular ferrite.

The same general arguments apply to Mo and Ni additions which are strong acicular ferrite formers.

Consumables containing 0.5% Mo, 1-3% Ni or combinations of Ni and Mo are often selected for high toughness applications of C-Mn steels. At low heat inputs (Δt 800-500°C \sim 10 sec), Mo and Ni additions would be expected to be beneficial to cleavage resistance by promoting acicular ferrite, but at high heat inputs, the situation could be reversed unless sufficient amounts were added to develop significant proportions of acicular ferrite. Taking the example of Mo, the use of Mo submerged-arc wires will probably be beneficial at heat inputs around 3 kJ/mm, but they will be less beneficial at high heat inputs unless other elements are used to achieve adequate hardenability and to develop acicular ferrite, e.g. Ni and Ti.

Microalloyed weld metal

Considerable research work has been devoted to the effects of elements such as Nb, V, Ti, B and Al on weld metal toughness. These elements may be present in small quantities as deliberate additions to the consumables or may be picked up in dilution with the parent steel.

The role of Nb is a little puzzling, but has been explained by the fact that Nb acts to inhibit proeutectoid formation at austenite boundaries but does not prevent ferrite plates growing into the grains to form upper bainite structures. From CCT considerations, Nb pushes back the polygonal ferrite nose, but not that for upper bainite. Thus for weld structures of low hardenability, i.e. those in which the polygonal ferrite and upper bainite fractions are high and acicular ferrite low, Nb will tend to increase the proportion of upper

bainite and acicular ferrite at the expense of polygonal ferrite at austenite boundaries. The yield strength will also increase and the net effect would be a fall in cleavage resistance. For weld structures already containing a high proportion of acicular ferrite due either to a low heat input or adequate alloy content at a high heat input, Nb will still act to decrease the polygonal ferrite fraction, but the resulting overall increase in acicular ferrite fraction is more important than a small increase in the bainite fraction. However the net effect of Nb additions in this case is zero, since the improvement in acicular ferrite content is offset by the formation of small segregated lath martensite regions. The aforesaid holds good as long as Nb is below 0.04%. Above this level, Nb has been found to markedly lower toughness even if the acicular ferrite content is high.

While Nb inhibits the formation of proeutectoid ferrite, vanadium has been found to inhibit upper bainite formation. The effect of vanadium primarily depends on heat input and hardenability of weld metal. In cases where high heat input is combined with low weld hardenability, the increase in V is not sufficient to achieve the hardenability necessary to form substantial acicular ferrite; on the other hand the yield strength increases with V additions and the net result is a fall in cleavage resistance. If the initial weld alloy content is sufficient to develop a reasonable proportion of acicular ferrite for the given welding procedure, the V additions increase the proportion of acicular ferrite by decreasing the upper bainite fraction, and cause weld metal toughness to improve.

Vanadium additions to Nb weld deposits have been found to promote proeutectoid ferrite and lower yield strength. This effect might be explained by the role of V in inhibiting upper bainite formation as mentioned earlier. The result is higher proportions of proeutectoid ferrite and acicular ferrite and improved cleavage resistance.

Titanium additions up to 0.04% max. have been found to improve Charpy-V toughness of mild steel welds deposited by the CO₂ process by increasingly refining the microstructure and promoting more acicular ferrite. Above this level, the microstructure is further refined but the yield strength increases markedly and the toughness gets reduced. In alloyed weld metals deposited by the MMA, Submerged-arc or MIG processes, similar good effects have been observed but the optimum level of Ti has been found to be 0.015% or less. In all the above investigations the welding conditions and alloy contents were such as to develop acicular ferrite

even in the absence of Ti. Titanium additions increased hardenability and in particular inhibited upper bainite formation and promoted acicular ferrite. Evidence has been gathered that Ti and V additions are more effective in combination than when added alone in the same way as Nb and V are.

As a microalloying element in weld metal, boron has stolen considerable limelight in the last five years. Its useful property is that it can eliminate proeutectoid ferrite and promote fine grained structures without the disadvantage of significant solid solution hardening which one sees with Mn, Mo, etc.

The demand for high heat input, high deposition rate submerged-arc welding procedures together with good low temperature weld metal toughness properties in C-Mn structural steels has led to the development of new submerged-arc solid or fluxcored wire containing controlled amounts of B, Ti and Mo⁴ (a patented solid wire called Ti-Bor-22 contains 0.22 Mo, 0.0038 B and 0.03 Ti besides 1.50 Mn). B and Mo additions give hardenability and hence a low austenite transformation temperature producing a strong and tough acicular ferrite, while Ti prevents contamination of the boron by N₂ or O₂. The total effect is to increase the proportion of acicular ferrite in the microstructure and refine the acicular ferrite grain size, thereby leading to significant increases in sub-zero Charpy-V impact values. Using Ti-Bor-22 wire with highly basic flux to weld 50 mm t BS 4360 50 D plate by multipass technique at heat input of 7.9 kJ/mm, Charpy-V values of over 100 J at -60°C have been reported⁵ for the weld metal which had 84% acicular ferrite and no ferrite side plates.

This concept of Ti-B additions was recently extended to manual electrodes. The conventional 0.5-3.5% Ni type electrodes which give good Charpy-V impact values down to -50°C in flat position have been found to possess inadequate COD property even at -10°C for welds deposited in vertical position which normally involves high heat input rates. On the other hand, a typical Ti-B electrode without Ni and having the composition given below has been reported⁶ to give, in addition to excellent Charpy-V values, COD values of 0.25 mm or more at -50°C even for the vertically deposited welds at heat input of 3.5 kJ/mm. This exceptional resistance to fracture by cleavage has been ascribed to the microstructure which consists of fine structures of proeutectoid ferrite and acicular ferrite, free of side plate ferrite.

C	Si	Mn	Ti	B	YS N/mm ²	UTS N/mm ²	Eln. %
0.07	0.20	1.65	0.04	0.004	510	570	30

This development has special relevance to site welding of LPG storage tanks where considerable amount of vertical welding is involved and adequate toughness at -50°C has to be ensured.

Resistance to microvoid coalescence

In C-Mn deposits, fracture by void nucleation, growth and coalescence at non-metallic inclusions is considered less serious with respect to catastrophic failures than fracture by cleavage. Yet if one is concerned with Charpy-V upper shelf values, he must understand the metallurgical factors controlling this mechanism.

For good Charpy-V toughness, the weld metal must have low non-metallic inclusion population, that is, low volume fractions and a high mean free spacing of the inclusions. These can be realistically determined only by using replica techniques and the electron microscope. To keep inclusion volume fractions low, oxygen and sulphur levels in the deposit must be kept minimum. Since the solubility of oxygen in weld metal is extremely low, its main effects are related to the volume fraction and type of oxide reaction products trapped in the solid weld by the rapid freezing over of the weld pool surface. The main oxide inclusions are (Mn, Fe) SiO_3 , SiO_2 , Al_2O_3 , TiO_2 and (Fe Mn)O. Sulphides such as (Mn Fe)S, nitrides, carbides and carbo-nitrides are also present⁷.

The total oxygen level in a weld reflects the volume fraction V_o of oxide inclusions through the approx. formula $V_o = 5.5 [\% \text{ O}]$, based on oxide inclusion densities being $3700\text{-}4000 \text{ kg/m}^3$. Oxygen is determined by vacuum fusion method.

Oxygen levels vary markedly between welding processes. In MMA welding, rutile and acid type electrodes give deposit oxygen content of 0.04 to 0.10%, while basic types give 0.02-0.04%. Fig. 2 shows a clear inverse correlation between oxygen at 300-1000 ppm (parts per million) and Charpy-V impact energy at 0°C for MMA deposits. In submerged-arc welding⁸ the oxygen content \pm range is 0.02-0.17% depending

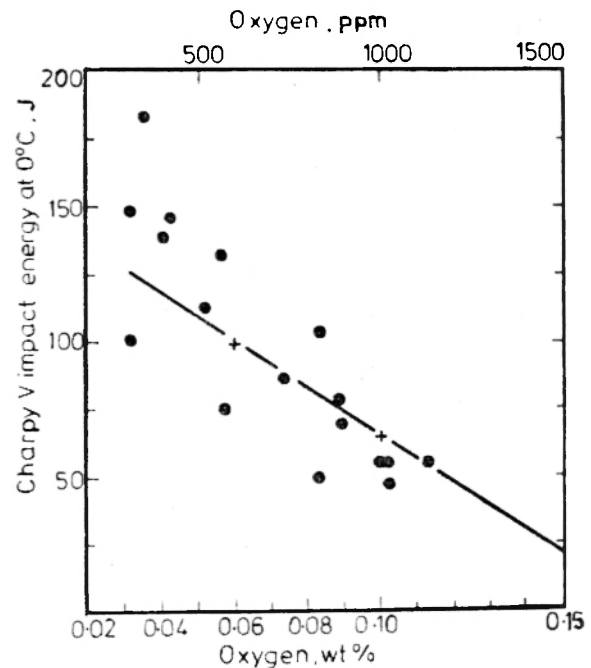


Fig. 2. The effect of oxygen on Charpy-V energy in MMA weld deposits at 0°C .

on the type of flux and weld metal dilution (see Fig. 3), for MIG welding 0.01-0.02% and for CO_2 -welding 0.04-0.06%. One may now understand why for obtaining highest toughness in gas-shielded welding, one is advised to use an argon-rich gas shield and vacuum deoxidised wires and why wires having high percentages of Mn and Si are used when welding under the highly oxidising CO_2 shield.

Sulphur contributes directly to the inclusion volume fraction by forming sulphide inclusions. For highest toughness, one must ensure that this impurity element is kept as low as possible in wires, fluxes and base plates.

Nitrogen is another impurity element one must guard against. Moragaki et al⁹ have studied the effect of N_2 in weld metal on Charpy-V impact property at -30°C and their results are shown in Fig. 4. It is observed that the impact value falls steeply when weld N_2 content exceeds 100 ppm (0.010%). Ingress of N_2 into weld metal is the result of its ionisation in the arc. The degree of ionisation increases with arc temperature.

Cellulosic and basic electrodes have the lowest nitrogen levels of MMA electrodes of around 0.004-0.014% by virtue of the heavy shielding effect of the gases from the decomposition of the cellulose and basic carbonates respectively. By comparison, acid and rutile coated electrode welds have higher typical

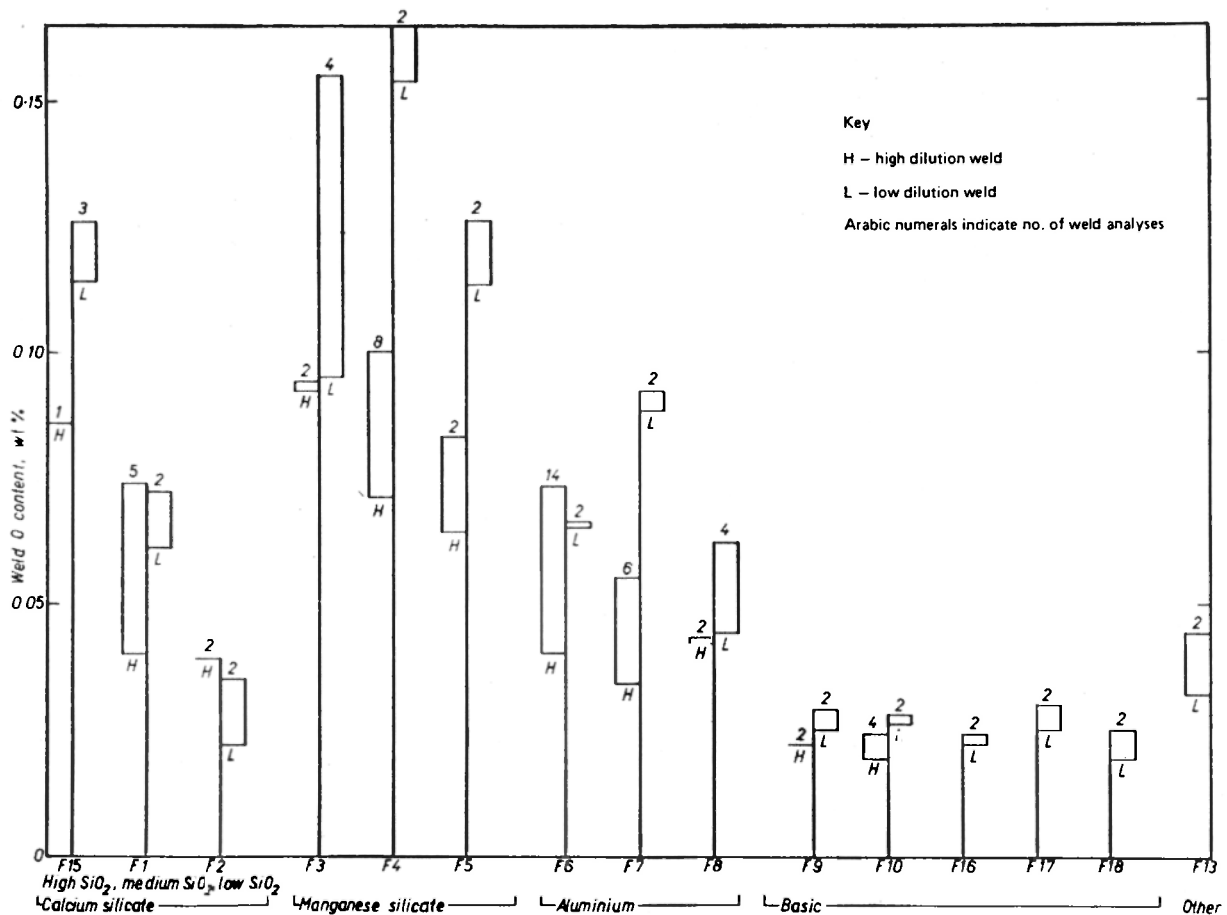


Fig. 3. Ranges of O content from welds made with different fluxes

mean figures of 0.022 and 0.027% N₂ which is attributable to the longer arc lengths and higher oxygen levels. Over-baking the acid or rutile types increases nitrogen levels because of reduced H₂/H₂O shielding; basic types are unaffected up to 450-500°C because of the carbonate stability¹⁰. Submerged-arc welds have nitrogen range of 0.005-0.011%, most of it derived from wire and plate. The CO₂ process is prone to nitrogen entrainment in the arc. It is usually caused by some defect in the equipment or faulty welding technique or by windy conditions. Porosity occurs in CO₂ welds when the N₂ content exceeds 0.020% and this happens when nitrogen in the shielding gas is above 3% for short circuiting arc and above 15% for spray-type arc¹¹. The so-called 'open arc' weld deposits given by self-shielding flux-cored wires can have N₂ levels as high as 0.070-0.20% unless the wires have correct amounts of Al and Ti to play the role of final deoxidation and nitride fixing. TIG and MIG welds have also been known to contain nitrogen in spite of argon shielding, more so in draughty conditions.

WI Scheme

Having realised the inadequacy of Widgery's terminology on weld metal microstructure referred to earlier, the Welding Institute put forth in 1980 a revised scheme¹² whose objective is to quantify two main features of any weld microstructure, first the proportion of the various constituents present and second their effective dimensions.

Initially, the prior austenite grain size is determined. Its shape is usually revealed by the presence of ferrite which formed at the austenite boundaries during transformation on weld cooling. A linear intercept technique is usually satisfactory for determining the grain dimensions. Sometimes it is not possible to determine this because the position of the grain boundaries cannot be recognised easily.

In the next step, the six categories of constituents as shown in table 1 are looked for and quantified. For this purpose, magnification of X500 is adequate.

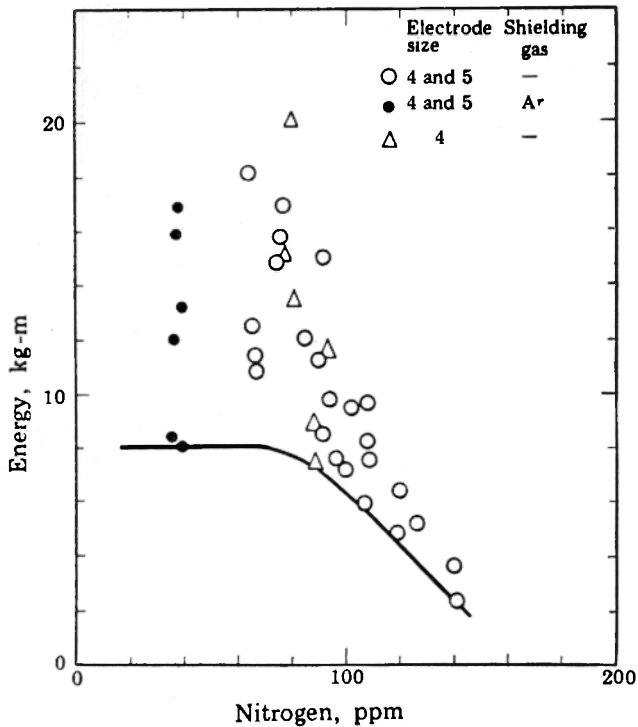


Fig. 4. Relation between Charpy-V notch energy absorption at -30°C and nitrogen content of weld metals.

Fig. 5 gives examples of the categories. Table 1 also lists equivalent descriptions used elsewhere. Each of the categories has been given a code to help in point counting and in annotating micrographs as is done in Fig. 5.

This new technology differs from Widgery's in the following respects :

- (i) The term 'proeutectoid ferrite' is replaced by more descriptive headings which recognise that ferrite can occur either in association with austenite grain boundaries (GF) or can be apparently intragranular (PF) ;
- (ii) replacing the term 'lamellar component' by 'ferrite with aligned M-A-C' which means ferrite containing martensite-austenite (M-A) or carbide (C) phases ;
- (iii) identifying martensite as a separate constituent from ferrite-carbide aggregates such as pearlite.

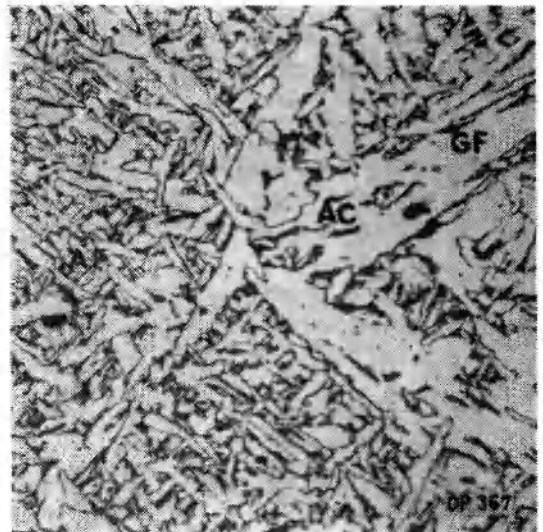
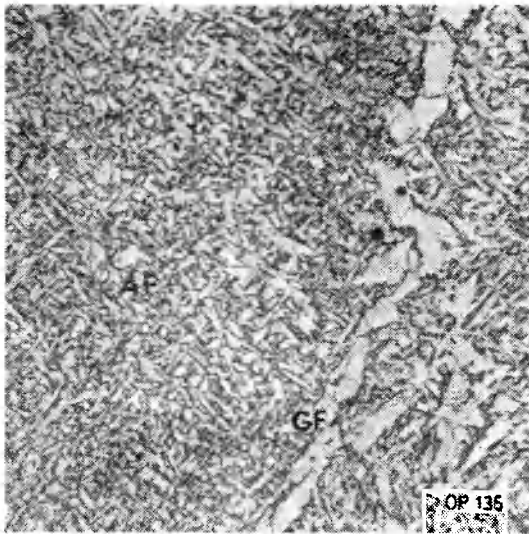
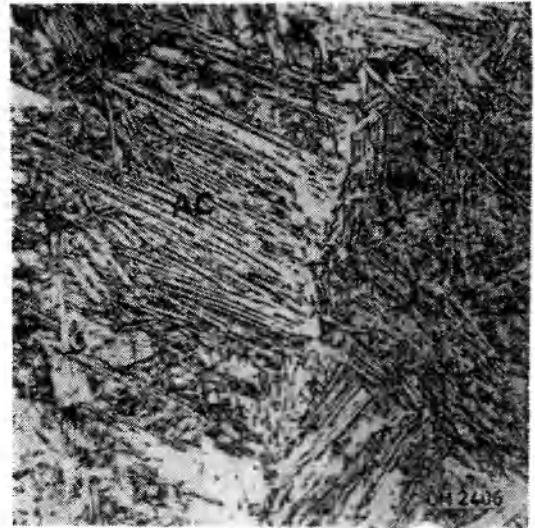
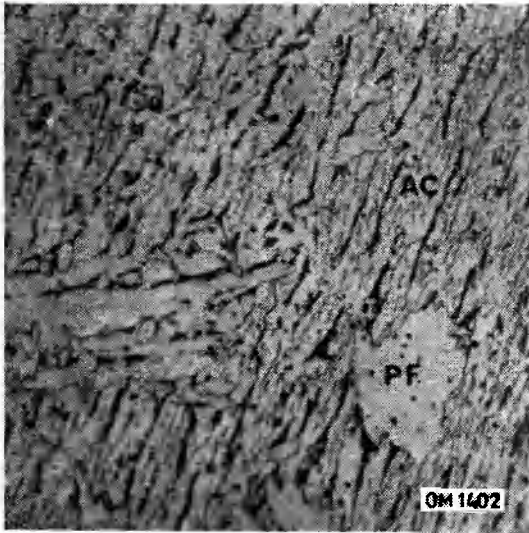
The term 'bainite' is not included because confusion surrounds its use and very special details are required to identify the constituent.

In the third step, specifically those constituents revealed by a picral etch are observed at high magnification. They are M-A phases and/or carbides whose proportion, type and distribution may be correlated with mechanical properties. Their proportions which can often be up to 10% of the microstructure can be usefully quantified by point counts. Further subdivision into twinned or lath martensite is possible through transmission electron microscopy.

In point counting, errors may occur due to (a) selection of inappropriate region or section, (b) careless specimen preparation, (c) difficulty in classifying constituents, or (d) statistical method used.

References

1. Widgery D. J. *Welding Journal*, 1976 Mar p. 57-s.
2. "A technique for the characterisation of weld metal microstructures" by Widgery and Davey, IIW Doc.11-A-389-76.
4. "Towards improved submerged-arc weld metal" by Garland & Kirkwood, *Met. Con.* May/June 1975.
5. "Effect of Ti and B additions to multipass SA welds in 50 D plate" by Still & Rogerson, *Met. Con.* July 78.
6. "Development of high-COD Ti-B bearing covered electrode", by Koshio et al., 11W Doc. 11-955-81.
7. "Oxygen in steel weld metal" by Blake, *Met. Con.* Mar 79.
8. "Influence of flux on element transfer during SA welding" by Davis & Bailey, *WI Res. Rep.* 66/1978 of June 1978.
9. "Development of an electrode for steel structures in low temperature service" by Moragaki et al., 11W Doc. 11-746-75.
10. "Nitrogen in steel weld metal" by Blake, *Met. Con.* Apr 79.
11. "CO₂ welding" by Smith, WI publication 1970.
12. "A Scheme for the quantitative description of ferritic weld metal microstructures" by Abson & Dolby, *WI Res.* Apr 1980.



20 μm

Key

- GF = Grain boundary ferrite
- PF = Polygonal ferrite
- AF = Acicular ferrite
- AC = Ferrite with aligned M-A-C
- FC = Ferrite - carbide aggregate (includes pearlite)
- M = Martensite

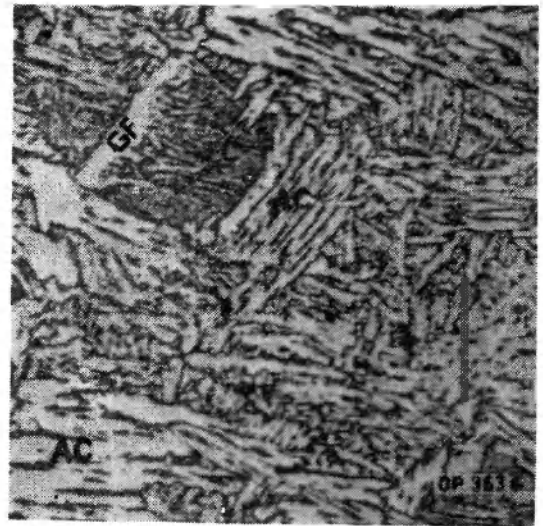


Fig. 5. As-deposited weld microstructures showing the various microstructural constituents.

TABLE 1 Proposed scheme for the quantitative description of weld metal microstructural constituents using the light microscope.

<i>Orientation of section (Please indicate)</i>		<i>Sketch (if required)</i>		
Transverse				
Longitudinal				
Plan				
Other				
<i>I Determination of prior austenite grain size—nital etch</i>				
Mean linear intercept= μm				
<i>II Examination of constituent phases—nital etch</i>				
Preferred magnification=X500 Magnification used=X				
Microstructural constituent	Code	Previous descriptions	Characteristic dimension of constituent or colony,* μm	Proportion, + %
Grain boundary ferrite,	GF	Proeutectoid ferrite, ferrite veins, blocky ferrite, polygonal ferrite		
Polygonal ferrite	PF	Ferrite islands		
Acicular ferrite	AF	Acicular ferrite		
Ferrite with aligned M-A-C	AC	Ferrite side plates, upper bainite, feathery bainite, lamellar product		
Ferrite-carbide aggregate (includes pearlite)	FC	Pearlite, ferrite + interphase carbide		
Martensite	M	Martensite		
<i>III Examination of microphases—picral etch</i>				
Preferred magnification range=X 1000 to X 2000 Magnification used=X				
Microstructural constituent	Code	Characteristic dimension of constituent, μm	Proportion, + %	
Ferrite	F			
Ferrite—carbide aggregate	FC			
M-A constituent (includes retained austenite, lath martensite and twinned martensite)	M-A			

*e.g. Ferrite band width, acicular ferrite lath width, colony size of ferrite with aligned M-A-C. (see text).
+Quote to nearest whole number. See text for discussion of accuracy.