JOURNAL OF MATERIALS SCIENCE 22 (1987) 3812–3820

Review

Acicular ferrite in carbon–manganese weld metals: an overview

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Acicular ferrite has been identified as the microstructural feature which controls the strength and toughness of carbon manganese weld metals. The present overview seeks to summarize the existing understanding of the nature of this phase and the control through changes in matrix hardenability, cooling rate in the temperature range 800 to 500° C, and the presence of suitable non-metallic nuclei. It also highlights those areas where further research may be required to complete our understanding.

1. Introduction

There has been a considerable amount of work on the nature and influence of this important phase in weld metals over the past 10 to 15 years. Various authors [1–4] have discussed the metallographic and crystallographic nature of this phase, its dependence on composition, cooling rate and the presence of non-metallic inclusions to provide suitable nuclei.

The influence acicular ferrite has on the mechanical properties of weld metals has also been the subject of much research and has recently been the subject of an excellent review by Abson and Pargeter [5]. The purpose of this paper is to summarize the existing understanding of the nature and role of this important weld metal microstructure and highlight those areas where further work may be required.

2. Microscopy and crystallography

The phase, acicular ferrite, forms at intermediate transformation temperatures (650 to 500° C) and appears to consist of very fine-grained interlocking laths. It is formed within the prior austenite grains as a series of laths or plates (Figs 1 and 2) which possess average aspect ratios of between 3:1 and 5:1 [2, 6].

Tuliani [6] and Watson [7] carried out very detailed electron microscopy on thin foils and determined that acicular ferrite consisted of lath-like ferrite separated by high-angle grain boundaries, the habit planes for the phase were associated with the [1 0 0], [1 1 0] and [3 3 1] directions. Electron diffraction indicated that the axial ratio was close to unity (measured value = 1.008), which suggested that very little carbon was held in solution. These studies also revealed that the average grain size for the phase was in the range 0.1 to 3.0 μm and that the average dislocation density within the laths was 10^8 to 10^{10} lines/cm^2.

Both of these authors found evidence of small amounts of twinned martensite between some of the acicular ferrite laths which suggested that the carbon had been concentrated in the last remaining austenite by rejection from the transformed ferrite phases. Recent work by Lathabai and Stout [8] has identified a similar twinned martensitic microphase between the acicular ferrite laths. Ricks *et al.* [2] felt that the presence of this carbon partitioning was evidence of a pro-eutectoid-type reaction. In addition, these authors observed the presence of growth ledges between the acicular ferrite and the parent austenite; these ledges had been seen previously to be associated with Widmanstatten development of ferrite side plates by earlier workers [9].

Later authors [10] found that acicular ferrite grew in competition with the higher temperature transformations, such as polygonal ferrite and ferrite side plates, and concluded that acicular ferrite was a true intermediate phase between the pro-eutectoid ferrite and bainite reactions.

The literature to date suggests that there is reasonable agreement concerning the microstructural form and crystallographic nature of acicular ferrite, but further studies are needed to establish the exact growth characteristics of the phase; and in particular how it relates to the other ferrite morphologies, when competitive reactions are occurring.

It is worth pointing out that in the authors experience, the morphology of acicular ferrite can be very difficult to identify by optical microscopy when its formation temperature is close to the polygonal ferrite reaction (> 650° C).

3. Development of acicular ferrite structure

3.1. Role of composition

Without doubt the most important element in terms of controlling microstructure in steel is carbon. In weld metals the levels must be kept low enough to prevent

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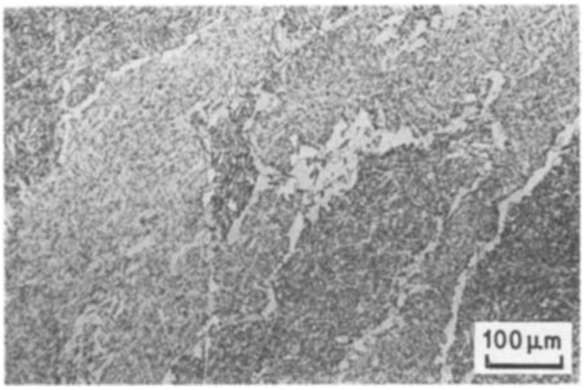

Figure 1 Prior austenite grain boundaries outlined by polygonal ferrite and ferrite side plates. Intragranular areas are occupied by acicular ferrite.

extensive lath martensite formation. At the carbon levels commonly encountered in weld metals, 0.05 to 0.15 wt %, the major effect is to control the amount of carbides formed. This observation has been made by Den Ouden *et al.* [11], who also commented on the refining nature of carbon on the microstructure. This was also the conclusion of the study by Evans [12] who considered that carbon essentially influenced the solidification substructure.

The effect of manganese has been thoroughly researched, the most systematic studies, which included detailed microstructural point counting, being conducted by Evans [13] and Harrison [4].

Evans [13] in MMA weld metals, and Novaes Gomes [14] in SA weld metals, found that increasing the manganese from 0.6 to 1.8 wt % progressively increased the acicular ferrite content whilst reducing the polygonal and side plate ferrites. They also found that the grain size of the acicular ferrite was refined at higher levels of manganese. These studies confirmed the earlier work of Tuliani [6] who found that increasing manganese changed the as-welded microstructure from a "fully bainitic" structure (close inspection of his original micrographs indicate that this was, in fact, a ferrite side plate structure with aligned MAC) to acicular ferrite.

Using continuous cooling dilatometry, Harrison [4] and Harrison and Farrar [3] studied the direct influence of manganese on the austenite to ferrite phase transformation. These authors found that manganese depressed the austenite to ferrite transformation temperatures at almost all cooling rates for MMA weld metals, with the polygonal ferrite start temperature being depressed to a greater extent than the acicular ferrite start temperature, thus resulting in progressively more acicular ferrite with increasing manganese additions. They also observed that manganese suppressed the formation of pearlite at typical weld metal cooling rates (typically 3 to 30° C $\sec^{-1}$: 800 to 500° C).

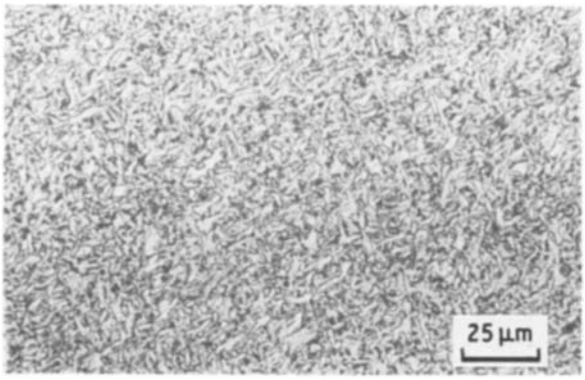

Figure 2 Very fine interlocking laths of acicular ferrite.

The influence of silicon has led to conflicting results in the literature; Dorschu and Stout [15] reported that silicon in the range 0.35 to 0.8 wt % had no effect on microstructure, whereas Tuliani and Farrar [16] observed a definite change from a side plate to acicular structure as the silicon content was increased from 0.2 to 0.4 wt %.

A recent paper by Evans [17] reviews the role of silicon in the range 0.2 to 0.94 wt % and concludes that increased silicon encourages the formation of acicular ferrite, especially when the manganese levels are below 1 wt %. At higher manganese levels the influence of silicon is less marked. Other conclusions from this work were that the aspect ratio of the acicular ferrite was changed and that an increased amount of retained austenite and martensite/carbides was formed between the acicular ferrite laths, this replacing pearlitic carbides seen at lower silicon levels.

Similar conclusions were reached by Court [18] who noticed that increasing the silicon level led to a decreased acicularity and a transformation from lath-like microphases to twinned martensite in the intragranular regions of the microstructure.

In general these results appear to confirm the earlier studies by Widgery [19, 20] on CO_2-shielded weld metals and Abson [21] on submerged arc deposits.

Although silicon can be seen to operate in a similar manner to manganese it must be pointed out that its major influence may be through the oxidation–deoxidation reactions taking place during the arc welding operation and the production of weld metal deoxidation products.

The role of nickel is much clearer. As a weak alloying element, it can be added in large quantities (up to 3.5 wt %) producing a refinement in the structure and a progressive change to acicular ferrite. The work of Taylor and Evans [22] indicated that as the level of nickel was increased the amount of grain-boundary ferrite decreased and the acicular ferrite became progressively refined. At the highest levels (3.5 wt %) they reported a change in the microstructure, with the development of intragranular martensite occurring between the acicular ferrite laths.

This observation is in agreement with the studies of Harrison [4] who reported that the effect of nickel was to depress the transformation temperatures throughout the weld metal cooling range, with the polygonal start temperature being reduced more than the acicular ferrite temperature. As with the case of manganese, this will result in a more favourable growth regime for acicular ferrite. Harrison [4] also reported the presence of martensite at the higher levels of nickel, this amounting to some 10% at a typical MMA cooling rate (30° C $\sec^{-1}$: 800 to 500° C) in a weld metal containing 3.6 wt % Ni.

The role of aluminium seems to be the most complex of all the elements studied. Terashima and Hart [23] who used different flux systems found that the behaviour was complex and depended upon the flux system of the weld metal. They suggest that the changes they observed could be explained by the influence on deoxidation behaviour, soluble aluminium content and the inclusion population. If there is an increase in the soluble aluminium it is believed that this will raise the austenite to ferrite transformation temperature, and simultaneously reduce the extent of titanium oxide formation.

The complex interaction between Ti–Al and oxygen was suggested by the earlier work of Bailey [24]. This work suggested that aluminium could reduce titanium from either the flux or the coating formulation and thus limit the extent of titanium oxide formation.

Recent work by Oldland [25] has also indicated that the changes in the flux type will alter the effect of aluminium. He found that with a fully basic system, increasing the aluminium content reduced the formation of acicular ferrite and favoured the formation of side plates; however, in a semi-basic system the effect was much smaller. He concluded that as the oxygen level was raised through changes in the flux type, that the optimum level of aluminium associated with the formation of acicular ferrite was also increased. This concept of an optimum level is in agreement with the work of Devillers *et al.* [26] who also suggested that the critical aluminium level was determined by the oxygen potential of the consumable system being used. This interaction between the critical levels of oxygen and aluminium on the production of acicular ferrite has recently been rationalized by the work of Grong and Matlock [27]. These authors found that the lowest Charpy 35 J temperature (and hence maximum acicular ferrite content) was obtained when the ratio $[\%\ \mathrm{Al}]\ [\%\ \mathrm{O}]^2$ was ~ 28. Above this level the inclusion size increased rapidly and the number of suitable nucleation sites for acicular ferrite formation was substantially reduced.

The influence on the transformation start and finish temperatures has been directly measured by Drury and Bannister [28] using continuous cooling dilatometry, but they were unable to measure any significant change in the transformation start temperature (T_s), as the aluminium level was increased from 0.01 to 0.07 wt % in weld metals containing 1.4 wt % Mn. There was, however, a slight decrease in the amount of acicular ferrite formed at all cooling rates investigated.

Oldland [25] also reported similar findings on the influence of aluminium on T_s temperature, but he did suggest that the formation of acicular ferrite was enhanced when the maximum transformation rate temperature, T_m, was close to T_s. The significance of this observation must be questioned, as the maximum difference in T_m he observed was only 40° C, which is only some two or three times the error associated with these types of measurements. Oldland [25] himself suggests that in high aluminium weld metals, acicular ferrite might be actively suppressed because of the lack of suitable nucleation sites, rather than the encouragement of side plate structures.

It must be said that the role of aluminium is still far from being resolved, because of the difficulty of obtaining systematic compositional changes, when the element plays such a powerful role in the oxidation–deoxidation reactions of the weld metal. This is obviously an area that needs more detailed work.

Other elements which have been found to influence the formation of acicular ferrite are niobium and molybdenum. The influence of niobium has been the subject of much debate in the literature with various conflicting views being advanced. Watson [7] was able to reconcile many of these observations by pointing out the confusion in the microstructural terminology used by the earlier workers. This led him to show that the effect of niobium depended upon the overall hardenability of the weld metal. At low hardenabilities the role of niobium was to produce ferrite side plate structures, whereas at high hardenabilities niobium would produce increased levels of acicular ferrite [29].

These conclusions have been supported in a review by Dolby [30] and by the recent systematic work of Novaes Gomes [14] who studied the interaction between manganese and niobium levels in high heat input submerged arc weld metals.

The major role of molybdenum appears to be to delay the polygonal ferrite reaction and favour the formation of true bainitic structures. Yoshino and Stout [31] and Garland and Kirkwood [32] both claim that additions of molybdenum up to 0.5 wt % increase the amount of acicular ferrite. This was clearly demonstrated by Novaes Gomes [14] who found that at the same levels of manganese (1.2 wt %) the addition of 0.15 wt % Mo increased the acicular ferrite content from 45 to 85%, together with a distinctive refinement in the size of the acicular ferrite. Interestingly this effect was not observed when the manganese level in the weld metal was reduced to 0.8 wt %; in this case the addition of molybdenum did not change the amount of acicular ferrite, but appeared to encourage side plate formation.

In summary, the role of composition on the formation of acicular ferrite seems to depend upon the direct influence the element has on the basic hardenability of the weld metal, the most effective elements being manganese and carbon and possibly silicon.

However, the level of hardenability required to develop acicular ferrite structures is quite critical, and is thus also influenced by other elements such as aluminium, niobium, molybdenum, and nickel, which depending on the initial hardenability level, may either promote or restrict the development of the acicular ferrite phase.

3.2. Role of cooling rate

In many of the early studies confusion over the microstructural terminology of the transformation products meant that the exact role of cooling rate was very confused.

Harrison [4] in a systematic study, using continuous cooling dilatometry of manganese and nickel containing weld metals, was able to separate out the effects, and observed the following direct influences of cooling rate.

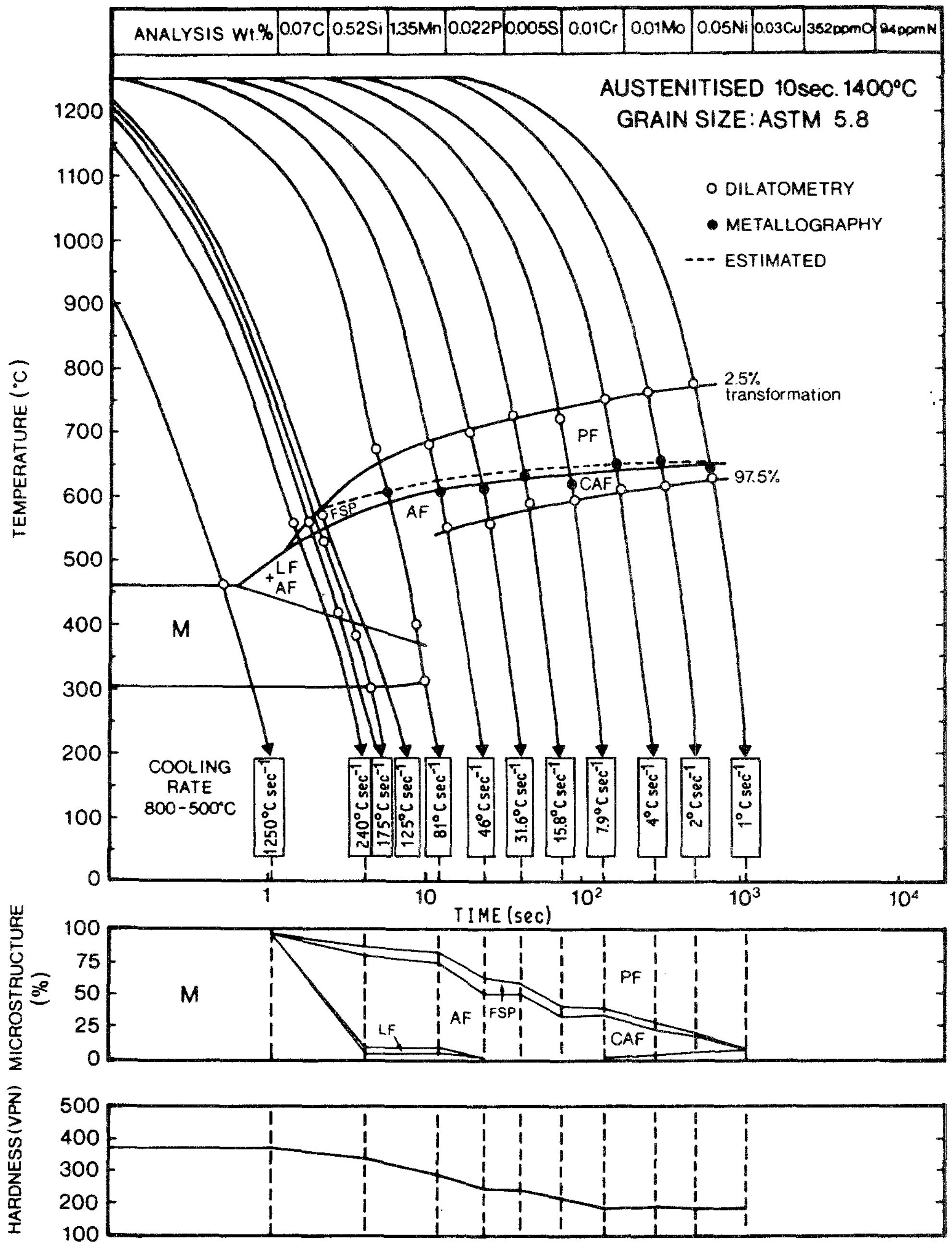

Figure 3 Continuous cooling diagram for a commercial 7016 weld metal. (PF = polygonal ferrite: FSP = ferrite side plates: CAF and AF = coarse and fine acicular ferrite: LF = lath ferrite and M = lath martensite).

(i) In general, the overall effect of increased cooling rate is to lower the transformation temperatures.

(ii) At very low rates ($< 1°\mathrm{C\,sec^{-1}}$: 800 to 500° C) the dominant transformation products were polygonal ferrite and pearlite. As the cooling rate was increased the polygonal ferrite became refined and was increasingly confined to the prior austenite grain boundaries as observed in the earlier work of Kinsman and Aaronson [33]. Also at these rates the materials appeared to show some tendency to develop Widmanstatten side plates with these side plates developing directly from the interface of the polygonal phase.

(iii) When cooled at intermediate rates (e.g. $15°\mathrm{C\,sec^{-1}}$: 800 to 500° C) almost all of the welds studied showed some evidence of intragranular products. These transformation products consisted of acicular and coarse acicular ferrite.

(iv) At the highest rates studied ($> 200°\mathrm{C\,sec^{-1}}$: 800 to 500° C) a lath ferrite structure developed. This consisted of parallel laths of ferrite separated by retained austenite, martensite-austenite (M.A), or carbides; when these rates were coupled with a high basic hardenability this lath structure was replaced by a true martensitic transformation.

The importance of these results may be seen if compared with measured cooling rates in practical situations. If one considers low heat input welding conditions of $\sim 1\,\mathrm{kJ\,mm^{-1}}$ on a 20 mm thick plate, this would produce a cooling rate of $\sim 75°\mathrm{C\,sec^{-1}}$. This cooling rate applied to a commercial 7016 electrode would produce some 35% polygonal and side plate structures and some 60% acicular ferrite, as seen in Fig. 3.

If one considers the same consumable used at a higher heat input of $\sim 4\,\mathrm{kJ\,mm^{-1}}$ on a 20 mm thick plate this would result in a cooling rate of $\sim 20°\mathrm{C\,sec^{-1}}$. At this cooling rate some 60% polygonal and side plate structures would develop with only 40% transforming to acicular ferrite. A similar continuous cooling transformation (CCT) diagram approach has been used by Kenny *et al.* [34] who were able to demonstrate the direct interaction between cooling rate and composition by employing low oxygen potential TIG welding. Their results confirm the transformation trends seen by Harrison, and highlight the direct involvement of inclusions in nucleating acicular ferrite in low oxygen weld metals (typically 20 to 100 p.p.m. in their work). When the oxygen level was increased to 300 p.p.m., by the use of flux cored wires, there was significant intragranular acicular ferrite development.

The major influence of the cooling rate is, therefore, to determine the degree of supercooling prior to the transformation and this must directly affect the thermodynamic driving force for the austenite to ferrite reaction.

The combined influence of basic hardenability and the degree of supercooling has been discussed in detail by Harrison and Farrar [3] who postulated a microstructural model which combined these factors together with the presence of suitable nucleation sites for the different transformations. This model is further discussed in Section 4.

3.3. Role of inclusions and nucleation sites

Following the original work by Ito and Nakanishi [36], who observed that the microstructure and properties could be influenced by the reduction of the oxygen levels to below 200 p.p.m., two recent papers, Lathabai and Stout [8] and Liu and Olson [37] and the work of Harrison [4], have sought to understand and summarize the effects of oxygen (and hence inclusions).

In summary, these authors conclude that the role of inclusions (in particular the role of oxygen) in influencing the phase transformations can be considered as follows:

(a) Direct alteration of the prior austenite grain size may occur if the austenite grain boundaries are pinned by suitable inclusions. The small prior austenite grains formed may then favour the grain-boundary reactions, and therefore polygonal ferrite will dominate over the intragranular reactions, which produce acicular ferrite. In their studies, Ferrante and Farrar [10] concluded that this pinning of austenite boundaries followed the classical Zener theories (as modified by Hellman and Hillert [38]), and that a reasonable correlation with observed grain sizes could be obtained if the weld metal oxygen level was below 1000 p.p.m.

It has also been suggested that inclusions present at grain boundaries can act as direct nucleation sites for the production of ferrite side plate structures [32]. This would allow a rapid transformation of the austenite due to the very high growth rates associated with Widmanstatten ferrite side plates [39].

(b) Inclusions can develop large thermally induced stresses due to the differences in the elastic modulii compared with the matrix. The presence of these stress/strain fields can then provide favourable sites for the nucleation of acicular ferrite [2, 4, 26]. Another similar effect is the presence of an increased dislocation density around the inclusion which again provides suitable nucleation sites. The lattice distortion in the vicinity of dislocations could also assist nucleation by either reducing the strain energy of the growing embryo or by altering the highly localized composition by diffusion.

Recent work by Johnson and Voorhees [40] on TiC particles in an Sn–Sb alloy has demonstrated that localized variations in composition in the presence of elastically interacting second-phase particles can be as much as 20 to 50%. Their studies have shown that these effects can exist for particles of $\sim 0.2\,\mu$m diameter and with relative elastic misfit strains of $\sim 1\%$. Similar effects could apply to weld metals in which there are very small deoxidation particles (such as seen in aluminium- and titanium-containing welds). If this mechanism was operative, then it would have a significant effect on the localized hardenability and hence transformation behaviour.

Localized variations in composition, in particular manganese, were suggested by Farrar and Watson [35], but these ideas were rejected by Barritte *et al.* [41], Pargeter [42] and Bhatti *et al.* [43], who could find no evidence of such localized concentrations of elements such as manganese in the close proximity of the inclusions studied.

However, as the effects suggested by Johnson and Vorhees [40] occur only over distances of some 0.1 μm, it might be possible that other researchers have missed this effect. In this connection it is interesting to notice that several authors [2, 18, 44] have published photomicrographs which clearly show acicular ferrite laths developing from pairs of inclusions (with diameters of ~ 0.5 to 0.8 μm), although it must be stated that there are other examples given of single inclusions nucleating acicular ferrite.

This microcompositional change is certainly worthy of further detailed study, as it may well provide a significant contribution to the nucleation mechanism.

(c) Direct nucleation of acicular ferrite on suitable inclusions. This can be viewed as the operation of lattice disregistry. If this is low then it is easy to produce an epitaxial growth of the phase on the nucleating particle as was observed in the excellent micrographs produced by Ricks *et al.* [2]. This raises questions as to exact chemical nature of the surface of the inclusion. Opinions in the literature as to the most effective particles in nucleating acicular ferrite have ranged from titanium-rich [45], aluminium-rich [42, 43], simple Mn-silicates [10], Al–Mn silicates which are not associated with sulphides [26] and complex copper sulphides on the surface of the inclusion [18].

Ricks *et al.* [2], however, have postulated that the nucleation of acicular ferrite is primarily promoted by the ability of the inclusion to reduce the nucleation barrier by acting as an inert substrate*. These various observations, which are often directly contradictory, seem to indicate that the exact chemistry and crystallographic type of the inclusion may not be the controlling features for acicular ferrite formation.

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*The work of Dowling *et al.* [*Met. Trans. A.* **17A** (1986) 1611] supports the inert high energy surface hypothesis.

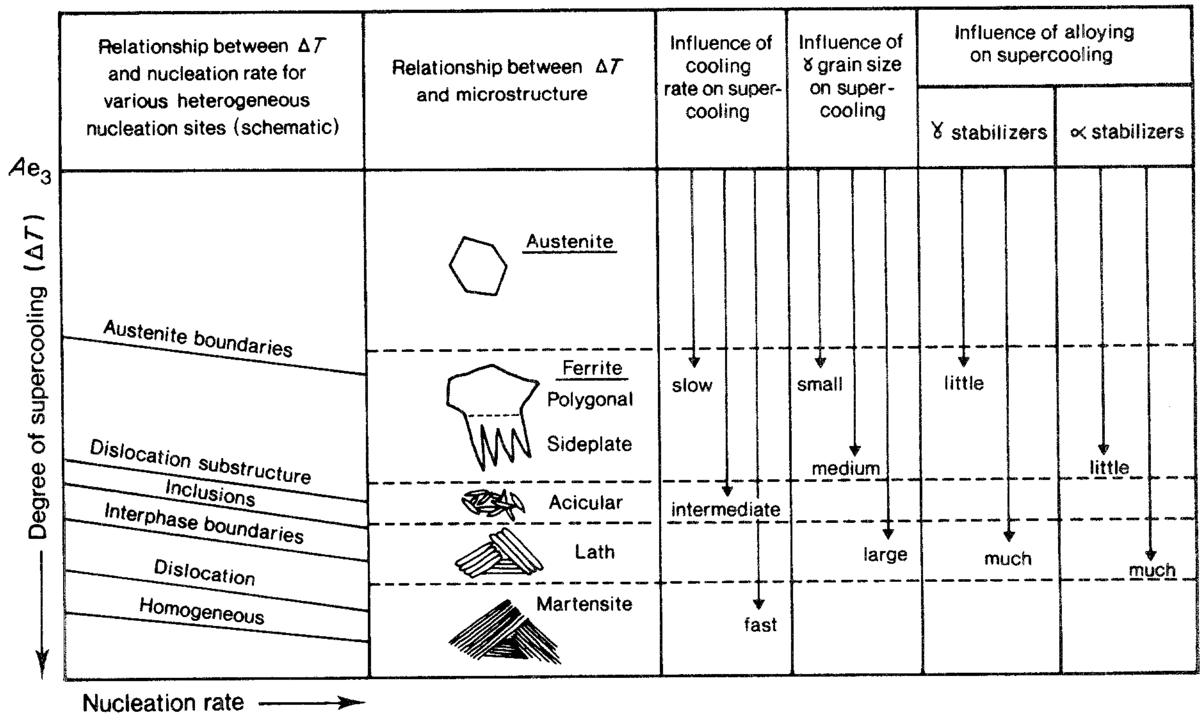

Figure 4 Schematic model showing the influence of cooling rate, γ grain size, and alloying on supercooling, and the proposed relationships between supercooling, nucleation rates, and microstructural development during continuous cooling.

This is an area which still needs critical study, as the surface chemistry of the inclusion may well be the actual transformation energy control rather than the bulk analysis or bulk crystallography. The very detailed work of Court [18] seems to provide clear evidence for this surface effect. He concluded that the initial nucleation of acicular ferrite occurred at inclusions followed by a subsequent sympathetic nucleation mechanism. The manganese copper sulphide globules attached to the normal oxysilicate inclusions were the most potent sites for nucleation. Court, however, pointed out that the surface chemistry of the inclusion was not the controlling feature due to the very high (30%) disregistry measured but that the localized strain fields developed around the inclusion were the major control.

(d) The size distribution of the inclusions may determine the availability of nucleation sites. Depending upon their size these inclusions could produce different effects. Liu and Olson [37] have suggested that small inclusions may effectively pin the austenite grain boundaries and hence not only restrict grain size but also provide good nucleation sites for polygonal ferrite.

This, however, could lead to a site saturation at the boundaries and hence limit growth, thus allowing the remaining austenite to transform intragranularly on other suitable inclusions. The bulk of the transformation to acicular ferrite could then proceed by the sympathetic mechanism as suggested by Ricks *et al.* [2]. It has also been suggested by Liu and Olson [37] that this size-distribution effect can explain the refinement observed with additions of boron. Boron is known to segregate to the austenite boundaries where it can poison the grain boundary and thus reduce polygonal ferrite formation. This will allow the bulk of the austenite to transform intragranularly on suitable inclusions to acicular ferrite.

These authors felt that the dominant factors governing the formation of acicular ferrite was the presence of coarse austenite grains and a mean inclusion diameter of $> 0.2 \mu m$.

From this brief review of the literature it is obvious that the role of inclusions in the formation of acicular ferrite is still far from being understood. There is still need for further critical study to resolve the issue, but it must be stressed that it may prove very difficult to separate out the exact roles of the different factors, such as matrix hardenability from the direct inclusion effects; both of which contribute to the overall transformation process.

In general it may be concluded that the role of inclusions in nucleating acicular ferrite is probably secondary to the effects of hardenability and/or the cooling rate of the weld metal. These factors must be favourable for the formation of the intragranular products before inclusions can act as suitable nucleation sites.

4. Microstructural model for acicular ferrite formation

In an attempt to reconcile these different factors Harrison and Farrar [3] used the earlier work of Harrison [4] to construct a schematic model for the formation of acicular ferrite. This is illustrated in Fig. 4, which shows the influence of cooling rate, austenite grain size and composition on the degree of supercooling, ΔT, between the Ae_3 and the observed transformation start temperature, T_s. This figure also shows the relationship between supercooling, nucleation rate and microstructure for the various nucleation sites.

Clearly certain assumptions must be made concerning the major nucleation sites: (a) the polygonal ferrite is nucleated at the austenite boundaries; (b) the acicular ferrite is nucleated on suitable inclusions or dislocation sites within the austenite; (c) the lath ferrite is sympathetically nucleated from other laths. This reaction requires that there has been prior nucleation by another mechanism; (d) the lath martensite is dislocation nucleated. These assumptions do not preclude the possibility that some micro-structural types may nucleate on other sites, but it is assumed that the order given will apply to the majority situation.

The influence of the different factors is shown by means of vectors in Fig. 4. For example a combination of fast cooling rate, large austenite grain size and high hardenability will produce a high degree of supercooling and hence a martensitic structure. On the other hand, slow cooling, small austenite grain size and low hardenability will result in little supercooling and hence grain boundary polygonal ferrite. It is therefore the intermediate conditions that are likely to produce intragranularly nucleated acicular ferrite.

It should be remembered that the vectors represent the maximum degree of supercooling reached and therefore show the lowest transformation products formed in each case, in reality there may well be some higher temperature products present.

The role of inclusions may now be considered. Complete removal of inclusions from a weld metal removes both the direct (inclusion substrates) and indirect (dislocation substructure) sites for acicular ferrite nucleation. Fig. 4 may be modified by removing these nucleation sites and the corresponding acicular ferrite supercooling regime in the diagram. The modified model now predicts that any combination of cooling rate, austenite grain size and hardenability which results in an intermediate degree of supercooling will now encourage the formation of side plate or lath ferrite structures as observed in the vacuum remelted studies of Harrison and Farrar [46]. This had shown that the removal of inclusions led to the formation of bainitic lath ferrites in high hardenability (1.5 wt % Mn) and ferrite side plates in lower hardenability (0.9 wt % Mn) weld metals.

Recent studies by Dallam *et al.* [47] using different flux formulations in the CaF_2–CaO–SiO_2 system allowed them to develop a series of submerged arc weld metals with different oxygen levels at the same basic hardenability level (0.1 wt % C, 1.0 wt % Mn and 0.4 wt % Si). As with the work of Harrison and Farrar [46], these authors found that as the oxygen content was reduced from ~250 p.p.m. to ~100 p.p.m. there was a change in the microstructure from ~90% acicular ferrite to a bainitic lath ferrite structure with aligned carbides between the laths.

In summary, to produce the best microstructure which contains at least 75% acicular ferrite, the weld metal must have an adequate prior austenite grain size and an optimum number of suitable inclusions. If there are too few inclusions the austenite grains will be large, but there will be insufficient nuclei to promote intragranular acicular ferrite and a bainitic structure will result. A large population of small inclusions, however, will produce pinned small austenite grains thus allowing grain boundary polygonal ferrite to dominate the transformation, before supercooling allows the austenite to reach the acicular ferrite regime.

This competitive transformation model has recently been developed by Bhadeshia *et al.* [48] who used phase-transformation theory to predict the microstructure of the fusion zone. The essential features of their model require a knowledge of cooling rate, chemical composition and prior austenite grain size in a similar manner to that predicted by Harrison [4]. This model confirms that polygonal ferrite is the primary phase formed and that the amount is controlled by the austenite grain size and the natural growth rate of the allotriomorph. The formation of acicular ferrite is then controlled by the amount of austenite remaining and the presence of suitable intragranular nucleation sites such as inclusions (these authors quote an upper limit of inclusions as defined by 260 p.p.m. oxygen).

5. Influence of acicular ferrite content on strength and toughness

Classical studies on the strength and toughness of different microstructural phases has shown that the best combination of properties can be achieved with a very fine matrix grain size which does not possess any continuous embrittling second phases.

This is exactly the situation that can be developed in weld metals that possess high levels of acicular ferrite. The average grain size is $\sim 1.0\,\mu m$ and there are high-angle boundaries between the laths.

Also, by correct control of composition, there can be very low levels of embrittling carbides (or martensite) at either the prior austenite or acicular ferrite boundaries.

Under these conditions low-energy cleavage failure as found in bainitic or pearlitic structures is difficult and the main mechanism must be via microvoid generation and coalescence.

A recent review by Abson and Pargeter [5] has produced a very detailed summary of the work over the past 10 years concerning the factors that govern the as-deposited strength, toughness and microstructure of manual metal arc weld metals. The major conclusions of their review are that the strength and toughness of as-deposited weld metals are determined by having a fine grain size coupled with a high proportion of acicular ferrite. In order to achieve this situation they state that there must be: (i) a correct state of deoxidation with an optimum level of oxygen ~250 to 450 p.p.m., so that there are adequate nuclei for acicular ferrite formation; (ii) an appropriate level of matrix alloying. While there is an optimum level for each of the elements, it must be remembered that there are definite interactions between certain elements such as Mn + C, Mn + Ni and Ti + Al + Si, which will alter the individual optimum levels; (iii) the cooling rate of the weld metal must be matched to the composition and the volume fraction of effective inclusions so as to allow adequate time for the intermediate reactions to take place. These conclusions

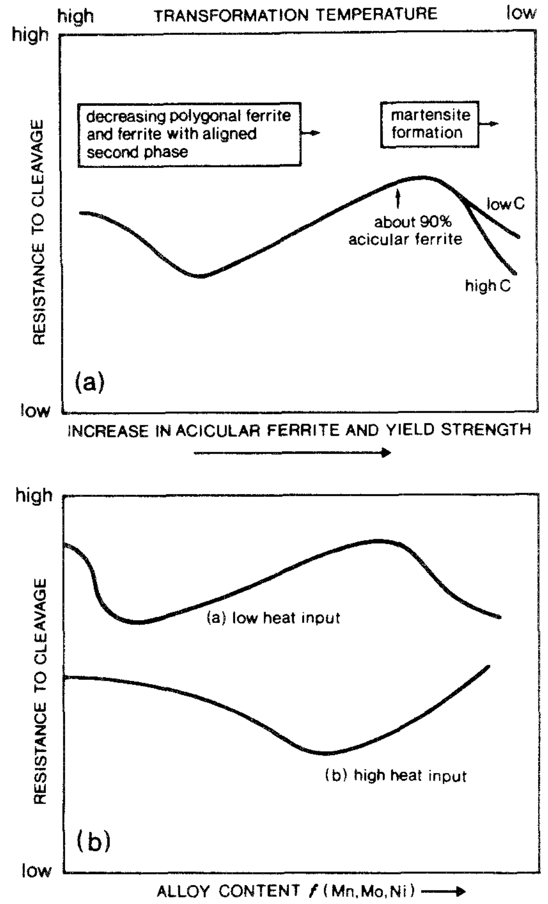

Figure 5 (a) Relationship between microstructure and toughness, and (b) effect of alloying content and heat input on toughness. After Dolby [49].

support the original suggestions of Dolby [49], that the strength and toughness of weld metals could be related to the acicular ferrite content as shown in Fig. 5.

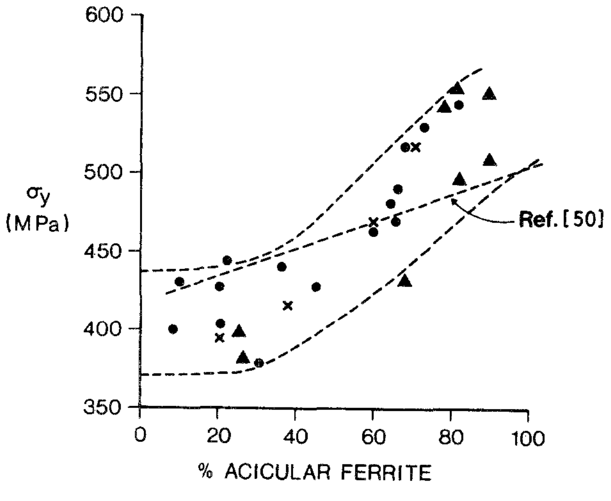

Figure 6 Variation of yield strength with the proportion of acicular ferrite for C–Mn–Nb weld metals. (▲) [7], (●) [14], (x) [13].

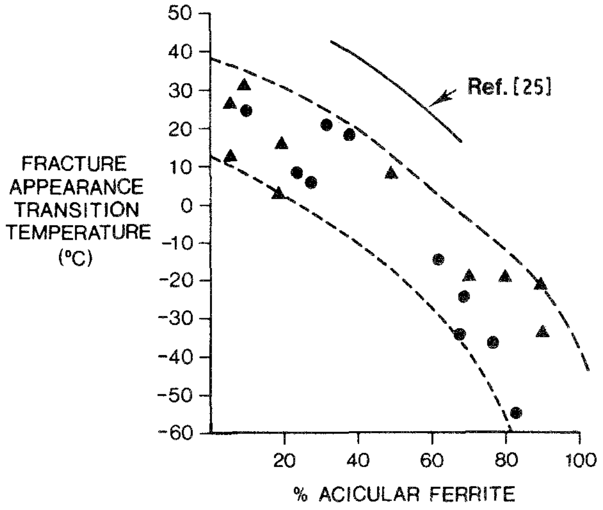

Figure 7 Variation of fracture appearance transition temperature with the proportion of acicular ferrite for C–Mn–Nb weld metals. (▲) [7], (●) [14].

The work of several authors [7, 13, 14, 25, 50] has been plotted in Figs 6 and 7. These clearly show the improvements produced both in Charpy "V" toughness and yield strength due to increasing acicular ferrite content. It is interesting to note that the original predictions of Dolby [49] are generally still applicable, except there has been no direct evidence for a fall in toughness at the highest levels of acicular ferrite, although it must be said that it is very difficult to produce weld metals with acicular ferrite contents > 85% without introducing too high a degree of solid solution hardening into the matrix. Recent studies by Terashima and Hart [23] and Fleck *et al.* [51] appear to confirm these overall trends for the role of acicular ferrite. In particular, the work of Fleck *et al.* [51] has confirmed the work of Abson and Dolby [52] that there is a significant effect of the acicular ferrite lath size on the weld metal toughness. In their studies they have shown that as the lath size was reduced from 2.4 to 1.4 μm there was a shift of some 40 J at −40° C, compared with a similar shift in Charpy "V" energy as the acicular ferrite content was increased from 30 to 80 vol %.

6. Conclusions

1. Acicular ferrite is formed at intermediate transformation temperatures (500 to 600° C). It consists of very fine-grained (typical sizes 0.1 to 3.0 μm) interlocking laths. These laths are separated by high-angle grain boundaries and can contain complex second phases at the boundaries, such as retained austenite, martensite and carbide aggregates.

2. The proportion of acicular ferrite developed within a particular weld metal appears to depend primarily upon the correct cooling rate in the range 800 to 500° C.

3. The most effective elements determining hardenability appear to be manganese, carbon and possibly silicon. However, the exact level of hardenability to develop acicular ferrite is critical and is influenced by other elements such as aluminium, niobium, molybdenum and nickel, which depending upon the initial hardenability level may either promote or restrict the development of the acicular ferrite phase.

4. The major influence of the cooling rate is to determine the degree of supercooling prior to the transformation and needs therefore to be balanced to the basic hardenability of the austenite matrix.

5. The nucleation of acicular ferrite is strongly influenced by the presence of suitable intragranular deoxidation inclusion sites. Opinions as to the most

effective composition of these inclusions is strongly divided, and it is probable that localized stress/strain fields and/or compositional changes at the inclusion/austenite matrix are the controlling factors.

6. The prior austenite grain size also appears to be a controlling feature as it will determine the balance of the competitive reactions which occur at the grain boundaries or those occurring intragranularly.

7. The formation of acicular ferrite seems therefore to depend upon a competitive phase transformation model which must take into account the matrix hardenability, degree of supercooling, prior austenite grain size and the presence of suitable intragranular nucleation sites.

8. The proportion of acicular ferrite formed within a weld metal appears to control both the strength and toughness. The best combination of properties is obtained when there is more than 65% of the phase and the average lath size is ~ 1 μm.

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*Received 22 January
and accepted 15 April 1987*