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THE MECHANISM OF ACICULAR FERRITE FORMATION IN STEEL WELD DEPOSITS

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ABSTRACT

Detailed crystallographic measurements, surface relief experiments, microstructural observations and thermodynamic analysis reveal that acicular ferrite grows by a diffusionless mechanism, in which the parent and product lattices are related by an atomic correspondence. Acicular ferrite is found to be very similar to bainite and differs morphologically because it nucleates intragranularly on inclusions within the weld; the morphology is also modified by hard impingement between plates nucleated at adjacent sites.

ACICULAR FERRITE (α_a) in the primary microstructure of low-alloy steel weld deposits consists of non-parallel plates of ferrite within the austenite grains and its presence seems to correlate directly with improved toughness. The mechanism of acicular ferrite growth is not clearly established, but it apparently nucleates intragranularly at inclusions (1-4) within the large columnar austenite grains so typical of weld deposits. It is not known whether the formation of acicular ferrite leads to a change in the shape of the transformed region. Such a shape change would indicate the transformation mechanism. It is notable that the growth of all other plate-shaped ferrites in steels always leads to an invariant-plane strain-shape change which has a large shear component (5,6). The morphological development of acicular ferrite has also not been investigated and the carbon concentration of the ferrite during growth is unknown. The purpose of this work was to resolve some of these difficulties, focussing on the growth aspects of acicular ferrite.

EXPERIMENTAL

The study of acicular ferrite in welds is made difficult by the high degree of transformation which occurs during cooling. In such welds, most of the parent austenite has transformed by the time ambient temperature is reached so that crystallographic experiments are very difficult to conduct. The high degree of transformation influences the morphology of the ferrite by causing impingement between crystals growing from different sites; it is the morphology that exists during unhindered growth which usually provides clues to the mechanism of transformation. To avoid these difficulties, an unusual experimental weld with a high hardenability and carbon concentration was deposited in order to ensure a low-degree of transformation during cooling to ambient temperature, and to retain considerable quantities of austenite at the same time. The weld was deposited using a manual metal arc technique at a speed of 2mm/s, using a welding current and voltage of 180A and 23V(DC+) respectively. The joint geometry was compatible with ISO2560 and the weld consisted of some 24 runs (interpass temperature 250°C), with three runs per layer. The present work is concerned only with the primary microstructure of the top layer, which is not influenced by dilution or by the heat effect of other runs. The chemical composition of the weld was found to be Fe-0.201C-1.80Mn-0.44Si-0.03Ni-0.04Cr-0.01Mo-<0.01V wt.%, with 212 ppm (by wt.) of O and 110 ppm of N.

Samples from the top run were prepared for transmission electron microscopy (on a Philips 400T, operated at 120kV) by electropolishing in a 5% perchloric acid and 95% acetic acid solution at 50-75V and ambient temperature. Quantitative metallography was carried out on a Quantimet 720 image analysis system. A "Theta" high-speed dilatometer was used to measure the extent of reaction as a function of isothermal transformation temperature; the specimen preparation for this has been described elsewhere.

Some weld metal samples were polished and then sealed in quartz tubes, under a partial pressure of pure argon, and austenitised at

1200°C for 10 mins before continuously cooling to ambient temperature (by dropping the intact quartz tube into iced brine) with the specimen still sealed in the quartz tube. The samples were then examined for surface relief effects accompanying the formation of acicular ferrite, using Nomarski interference contrast microscopy.

RESULTS AND DISCUSSION

MICROSTRUCTURE - The as-deposited microstructure is illustrated in Fig. 1, and shows that the weld consists primarily of acicular ferrite plates with very little allotriomorphic ferrite and Widmanstaten ferrite. When sectioned on a random plane (Fig. 1,2), the acicular ferrite always presented a lenticular section, confirming the view that it has a thin-plate morphology in three dimensions. On such random sections, the thickness/length ratio of the α_a plates was found to be about 0.3, so that the true aspect ratio (after considering stereological effects) should be much smaller. Electron microscopy revealed considerable quantities of retained austenite between the acicular ferrite platelets (Figs. 3a,b). This austenite was used in crystallographic measurements discussed below. Some of the plates contained inclusion particles (Fig. 2) which were presumably responsible for the heterogeneous, intragranular nucleation of α_a (of course, the probability of observing such nucleation sites is rather small, so that inclusions need not be observed in all plates even though they may be present).

The plates have an apparent length of a few microns and the α_a/γ boundary gently curves to generate the lenticular outline of the plates (Fig. 2). The tips of the plates were also found to be smoothly curved. These morphological observations are consistent with a displacive transformation mechanism in which the lenticular plate shape of α_a arises through the need to minimise the strain energy associated with the displacive transformation. The observations are inconsistent with growth involving a ledge mechanism since the α_a/γ interface is certainly not found to be facetted in any way: more often than not, it exhibits clear curvature. The acicular ferrite plates were on a few occasions found to consist of "sub-units" (Fig. 4); such sub-units are found during the formation of bainite, which grows by the martensitic propagation of discrete platelets (sub-units) which together constitute a bainite "sheaf".

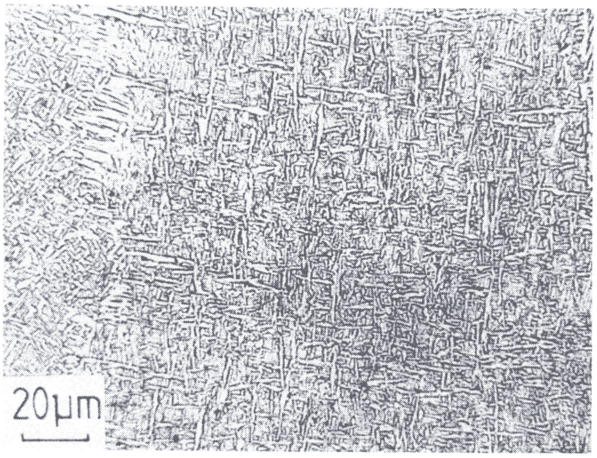

Fig. 1 - Optical micrograph showing the primary microstructure of the weld deposit.

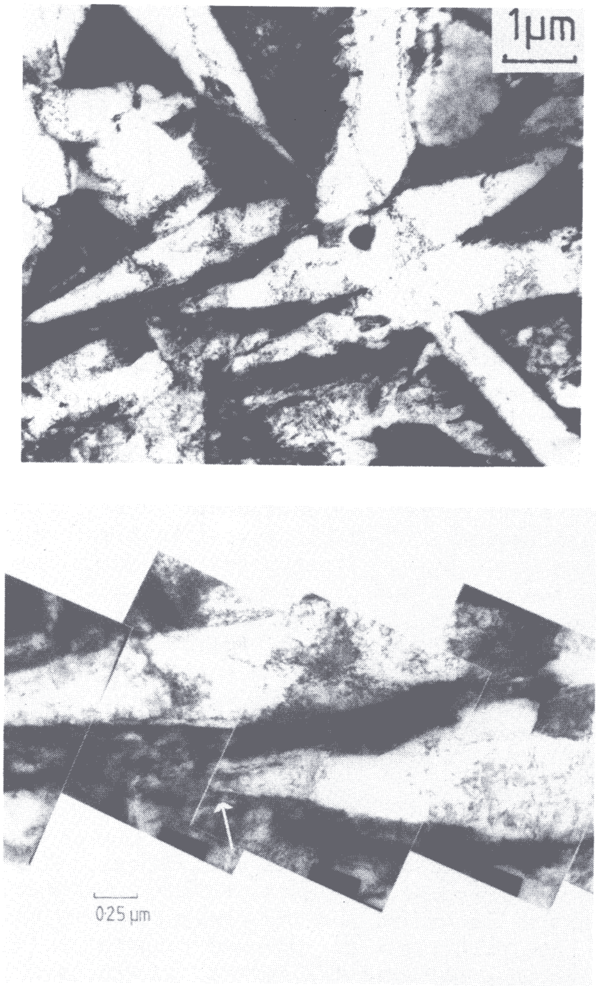

Fig. 2 - (a) Electron micrograph illustrating the lenticular morphology of acicular ferrite. (b) Montage illustrating the detailed tip morphology of acicular ferrite.

SURFACE RELIEF - Pre-polished specimens were transformed to examine (using Nomarski interference microscopy) the surface relief accompanying the transformation of austenite to acicular ferrite. The results are presented

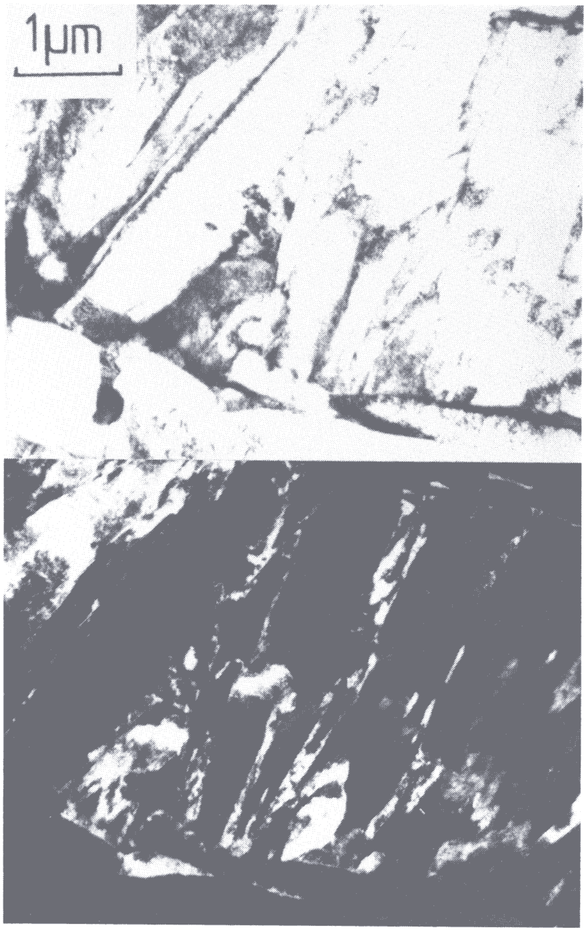

Fig. 3 - Bright and Dark field images of retained austenite in the primary weld microstructure.

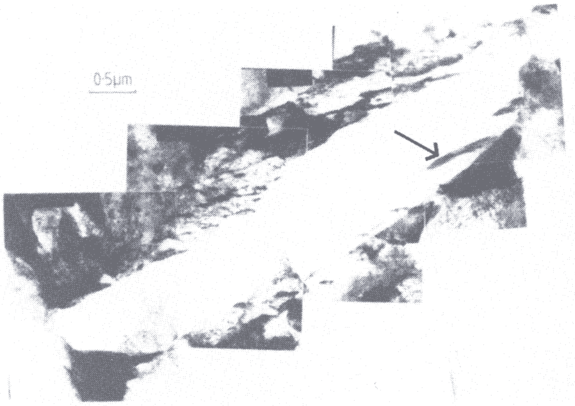

Fig. 4 - Electron micrograph illustrating the sub-units which make up what macroscopically appears to be a single plate of acicular ferrite.

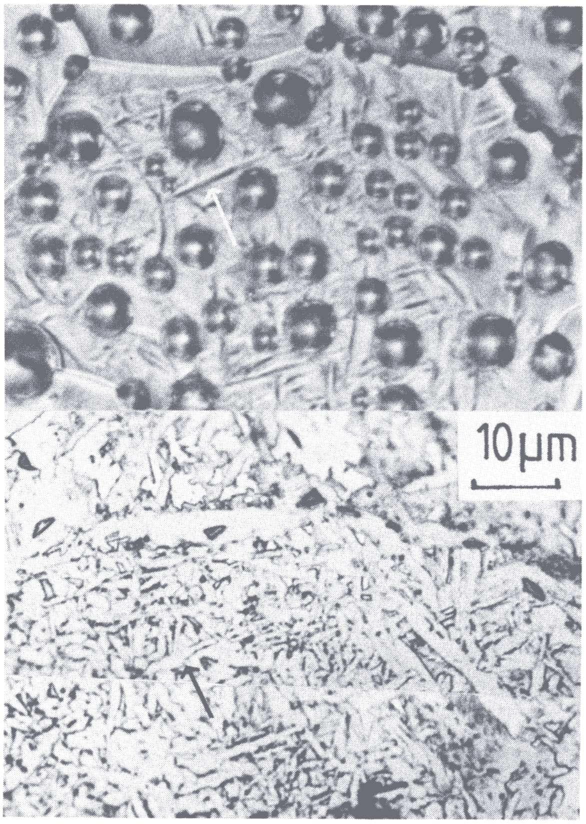

Fig. 5 - (a) Nomarski interference micrograph illustrating the surface relief accompanying the formation of acicular ferrite. (b) Corresponding optical micrograph.

in Figs. 5a,b and indicate that the formation of plate-shaped acicular ferrite is accompanied by a change in the shape of the transformed region. The uniformity of the contrast indicates that the shape change can be described as an invariant-plane strain with a significant shear component. This shape change can be taken to imply a displacive transformation mechanism and the existence of an atomic correspondence between the parent and product phases, at least as far as the atoms in the substitutional sites are concerned.

CRYSTALLOGRAPHY - For displacive transformations where the parent lattice changes into that of the product by some kind of a deformation of the former, the two lattices must <u>always</u> be intimately related (6). For steels, this means that the γ/α orientation relation must fall within the "Bain Region" (7-9). The homogeneous deformation which accomplishes the γ-α transformation is the Bain strain, which does not rotate any plane or direction by more than about 11°, so that any set of corresponding planes and directions can be made parallel after this strain by a rotation of not more than 11°. This 11° region about the Bain orientation is called the Bain region; for example, the Kurdjumov-Sachs and Nishiyama-Wasserman orientation relations fall within this region. For displacive transformations the γ/α orientation should always fall within this region. Diffusional growth on the other hand, occurs by the uncoordinated transfer of atoms across the interface so that the transformation is not limited by the presence of parent phase grain boundaries. It follows that the orientation relationship between α and the γ grain in which it grows, will not necessarily fall within the Bain region if the nucleation event occurs in one grain but growth in another adjacent grain.

The large amount of retained austenite found was used to advantage to determine the crystallography of acicular ferrite. All of 32 measurements of the α_a/γ orientation measurements were found to fall within the Bain region, the majority of which were within 6° of the Kurdjumov and Sachs relationship (Table 1).

Table 1 - Orientation relationship between γ and α_a. θ is the angle (degrees) between the closest packed planes of the two lattices, $\{111\}_\gamma$ and $\{011\}_{\alpha_a}$. ϕ is the angle between corresponding close-packed directions $<01\bar{1}>_\gamma$ and $<11\bar{1}>_{\alpha_a}$.

θ	ϕ	θ	ϕ	θ	ϕ
6	6	6	6	3	3
10	10	8	8	7	7
0	8	0	0	6	6
6	6	7	7	5	5
5	5	0	2	0	0
3	7	6	6	6	6
2	2	7	7	0	0
4	4	0	0	4	4
0	0	10	10	0	0
10	10	9	9	7	7
9	9				

Single-surface trace analysis to determine approximately the habit plane indices of acicular ferrite, with respect to the parent austenite was also carried out using electron diffraction. The results of some sixty experiments, some of which are illustrated in Fig. 6, indicate that the habit plane is near $\{0.117\ 0.675\ 0.729\}_\gamma$, which is very close to the $\{3\ 10\ 15\}_\gamma$ habit plane of many conventional martensites (10).

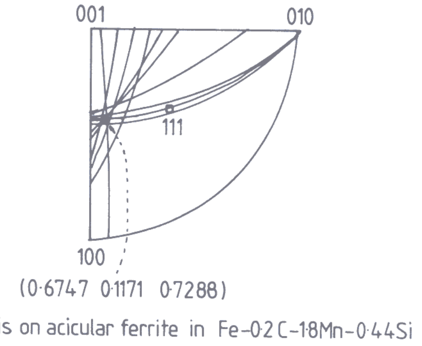

Fig. 6 - Single-surface habit plane trace analysis.

THERMODYNAMIC ANALYSIS - After reaustenitisation at 1200°C for 1 hour, the reheated weld metal was isothermally transformed for various time periods and at several temperatures prior to quenching to ambient temperature. The amount of acicular ferrite formed was measured on the Quantimet 720 image analysis system (the measurements refer to the α_a plates alone and do not include microphases as is conventional practice). The results are presented in Fig. 7a and show that the degree of transformation increases with undercooling below the B_s temperature, and thermodynamic analysis (the detailed interpretation of which is presented in refs. 11-13) of these data (Fig. 7b) showed that the formation of acicular ferrite ceased when the carbon concentration of the residual austenite reached the T_o' phase boundary. These results strongly suggest that acicular ferrite is really bainite which nucleates

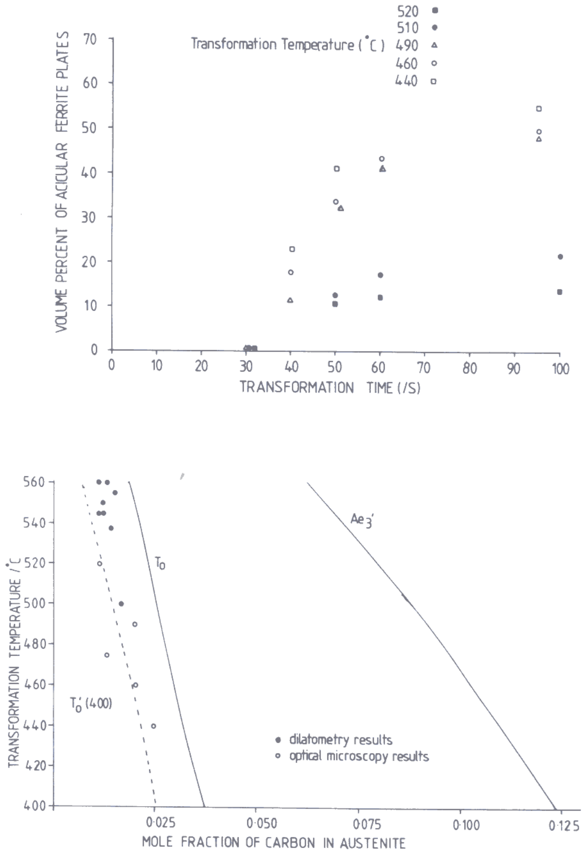

Fig. 7 - (a) Results of quantitative metallography on specimens partially isothermally transformed to acicular ferrite. (b) Thermodynamic analysis of results presented in (a), and further results obtained by dilatometry.

intragranularly; the growth of acicular ferrite is thus diffusionless and the carbon is redistributed between α and γ after the transformation event.

CONCLUSIONS

Our results suggest that the growth of acicular ferrite is diffusionless and occurs by a displacive transformation mechanism. The formation of α_a is accompanied by an invariant-plane strain shape change which implies the existence of an atomic correspondence between the parent and product phases. The crystallographic results are also consistent with this in that the orientation relationship between α_a and γ is always found to lie within the expected Bain region. The detailed shape of acicular ferrite also suggests a displacive transformation mechanism which does not involve interface advance by a step mechanism.

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