

Advances in Welding Science & Technology
ASM, Metals Park, Ohio, USA
1987, pp. 187-191

THERMODYNAMICS OF THE ACICULAR FERRITE TRANSFORMATION IN ALLOY-STEEL WELD DEPOSITS

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ABSTRACT

The mechanism by which acicular ferrite grows is not understood and this hinders the development of phase transformation models for predicting the microstructure of welds. In this work, we report experiments on the thermodynamics of the formation of acicular ferrite from austenite. It is found that on cooling below the B_s temperature, the degree of transformation to acicular ferrite increases, from zero at B_s , to an extent depending on the undercooling below the B_s temperature. However, at any given isothermal transformation temperature, the transformation ceases prematurely when the carbon concentration of the austenite reaches the level where diffusionless decomposition becomes thermodynamically impossible. The transformation thus exhibits the "incomplete reaction phenomenon" typical of the bainite reaction. Acicular ferrite is found to be similar to bainite, its growth being diffusionless, with carbon redistribution occurring after the initial formation of the ferrite plate. Acicular ferrite differs morphologically from bainite because it nucleates intragranularly on inclusions and because of hard impingement between plates nucleated on adjacent inclusions.

ACICULAR FERRITE (α_a) is a phase formed by the transformation of austenite (γ) during cooling of low-alloy steel weld deposits (1-3). It forms in a temperature range where diffusional transformations become relatively sluggish and give way to displacive transformations such as Widmanstätten ferrite, bainite and martensite. Its morphology consists of non-parallel plates of ferrite, which during the early stages of transformation nucleate on inclusions present in the columnar austenite grains typical of weld deposits. Subsequent plates may nucleate sympathetically on these inclusion-nucleated plates (4), so that a one-to-one correspondence between the number of active-inclusions and the number of α_a plates is not expected.

The term "acicular" means "shaped and pointed like a needle" but it is generally recognised that α_a has in three-dimensions the morphology of thin-plates. The true aspect ratio of such plates has never been measured but in two-dimensional sections, the plates typically are about 10 μm long and perhaps 1 μm wide (this implies that the true aspect ratio should be much smaller than 0.1). In wrought steels, the formation of plate-shaped ferrite (whether it be bainite or Widmanstätten ferrite) is always accompanied by a change in the shape of the transformed region (5). This shape change can be described as an invariant-plane strain with a significant shear component and is generally 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 atoms in the substitutional sites are concerned. Interstitial atoms, such as carbon may diffuse during transformation (especially when the ferrite forms at low-undercoolings where the partitioning of carbon is a thermodynamic necessity), without affecting the shape change or the displacive character of the transformation (5,6). The shape change accompanying the formation of α_a has never been determined experimentally.

The classification of microstructures on the basis of morphology is of considerable use in the study of structure-property relationships, but the prediction of microstructure requires a deeper understanding of the transformation mechanism. It is now well established that the presence of acicular ferrite in the primary microstructure of low-alloy steel weld deposits leads to improved toughness, but the mechanism of its formation is not understood. Attempts at modelling the primary microstructure thus rely on deducing the α_a content by difference rather than by direct prediction (3). In this work, our aim was to deduce fundamental information about the growth of acicular ferrite, focussing on whether the growth is diffusionless.

To enable the study of acicular ferrite without substantial interference from other reactions, a high-hardenability manual metal arc weld was deposited 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, the weld consisting of 27 runs with 3 runs per layer, the interpass temperature being 250°C. The composition of the weld was found to be Fe-0.06C-0.27Si-1.84Mn-2.48Ni-0.02Mo wt.% and the deposit contained an oxygen level of 320ppm by wt. together with 0.01Al, 0.02Ti and 0.01Nb wt.%.

Specimens for dilatometric experiments were machined from homogenised (1200°C for 3 days while sealed in a quartz tube containing a partial pressure of pure Ar) weld metal. The specimens were 3mm diameter cylindrical rods with a 1.2mm diameter bore and a 20mm length, electroplated with a 7µm layer of nickel, to provide constraint and avoid surface nucleation or surface degradation effects. The experiments were carried out in a 'Theta' high-speed dilatometer, the specimens being austenitised at 950°C for 10 mins (or 1200°C for 30 mins) before helium gas quenching to the isothermal transformation temperature. The dilatometer was calibrated using pure Ni and Pt rods, so that absolute length changes could be measured.

The relative length change ($\Delta l/l$) as measured on a dilatometer can be converted into volume fraction of transformation; the method is described in detail elsewhere (7). The calculations, however, require the linear expansion coefficients and lattice parameters of austenite and of ferrite. The ferrite linear expansion coefficient e_α was determined by first tempering specimens at 600°C for 10 mins to decompose any retained austenite and then recording the length change during cooling at a rate of 0.063°C/s. The measurements do not therefore account for the presence of a very small amount of carbide. The γ expansion coefficient e_γ was measured while the specimen was in the γ single-phase field. The ferrite lattice parameter (a_α) at ambient temperature was measured using a Debye-Scherrer technique (CuK_α radiation) and Newton-Ralphson extrapolation using the 011, 002, 112, 022 and 013 lines. Having determined the volume fraction v of ferrite obtained by transformation, the carbon concentration of γ , x_γ is obtained from the relation:

$$x_\gamma = \bar{x} + [v(\bar{x} - s)/(1 - v)] \quad (1)$$

where s is the amount of carbon in solution in the ferrite or tied up as carbides. \bar{x} is the average carbon content of the alloy. Since carbide precipitation does not occur rapidly over the time scale of the experiments, s is taken to be 0.02 wt.%.

PRIMARY WELD MICROSTRUCTURE AND TTT DIAGRAM - The primary weld microstructure, consisting mainly of acicular ferrite and microphases is illustrated in Fig. 1. Fig. 2 shows the calculated TTT curve for the weld (8); in such diagrams, the upper C curve represents the time taken for the initiation of diffusional transformations such as allotriomorphic ferrite and pearlite, whereas the lower C curve represents the time taken for the initiation of displacive transformations such as bainite and Widmanstätten ferrite (8). It is notable, however, that in this alloy, the Widmanstätten ferrite start temperature (W_s) equals the bainite start temperature (B_s). This means (9) that Widmanstätten ferrite should not be observed at all and immediately implies that the acicular ferrite observed in the weld deposit is likely to be bainite if it is not a completely new transformation. Fig. 2 also shows that for any reasonable weld cooling rate, the cooling curve cannot intersect the diffusional C curve, so that the alloy only transform by a displacive mechanism. The alloy is therefore ideally suited to avoid interference from other transformation products during the study of acicular ferrite. In addition, transformation during the quench to the isothermal transformation temperature can easily be avoided. Of course, a weld deposit is necessarily chemically inhomogeneous (10), and some formation of allotriomorphic ferrite and Widmanstätten ferrite cannot thus be avoided in the solute depleted regions, as seen in Fig. 1.

GRAIN SIZE EFFECT - The problem of resolving the nature of α_a is made worse by the fact that it does not readily occur in wrought steels, due to the absence of suitable intragranular heterogeneous nucleation sites and because the γ grain size of such steels is deliberately kept small; transformations initiated at γ grain boundaries thus swamp the interior regions as well. However, if after deposition, a weld is reheated into the

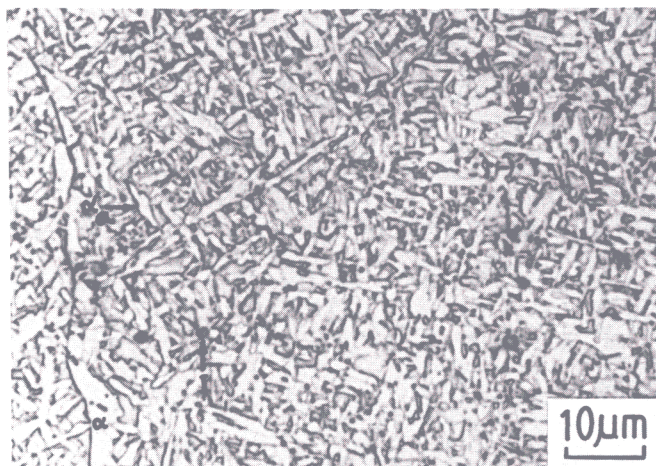


Fig. 1 - Primary microstructure of the weld deposit

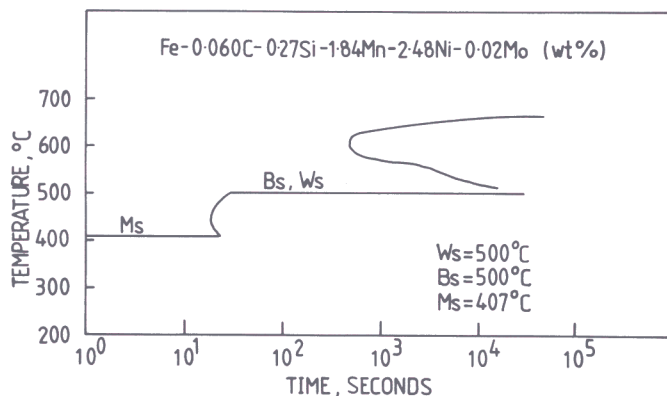


Fig. 2 - Calculated TTT curve for the Fe-Mn-Si-Ni-Mo-C weld

austenite phase field to generate large γ grains then these new grains still contain the inclusions which were present in the original deposit. Isothermal transformation of this structure in a temperature range where α_a usually forms in welds should then lead to the intragranular nucleation and growth of plates whose morphology is identical to that of the α_a observed in weld deposits. Fig. 3 illustrates the results of isothermal transformation experiments on reheated weld metal; in one case the weld was reaustenitised at a temperature of 950°C for 10 mins. (Fig. 3a) and another sample was austenitised at 1200°C for 30 mins prior to isothermal transformation at 460°C for 30 mins before quenching to ambient temperature. The higher austenitisation temperature has clearly led to a larger austenite grain size, and subsequent isothermal transformation has led to the intragranular nucleation of plates of ferrite which can be identified morphologically with acicular ferrite found in welds. On the other hand, in the specimen reaustenitised at 950°C, the small grain size has prevented intragranular nucleation, the ferrite plate nucleating at the austenite grain boundaries and growing by a sub-unit mechanism to give the morphology of sheaves of upper bainite. These results imply that acicular ferrite is not a new transformation product, but is simply intragranularly nucleated bainite. Its morphology differs from conventional bainite because it nucleates intragranularly, whereas in wrought steels, bainite nucleates initially at γ/γ or γ/α boundaries. α_a does not normally grow in sheaves because the development of sheaves is stifled by hard impingement between plates nucleated independently at adjacent sites.

BAINITE OR WIDMANSTÄTTEN FERRITE? - A further experiment is needed to clearly establish that acicular ferrite is similar to bainite and not to intragranularly nucleated Widmanstätten ferrite. Both these latter products grow by a displacive transformation mechanism, the details of which are discussed

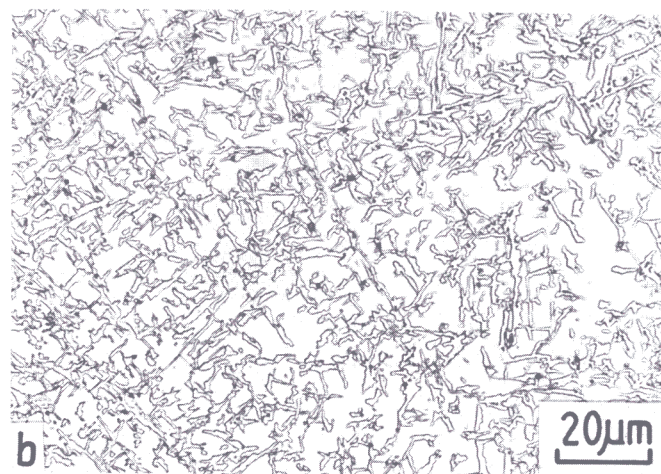
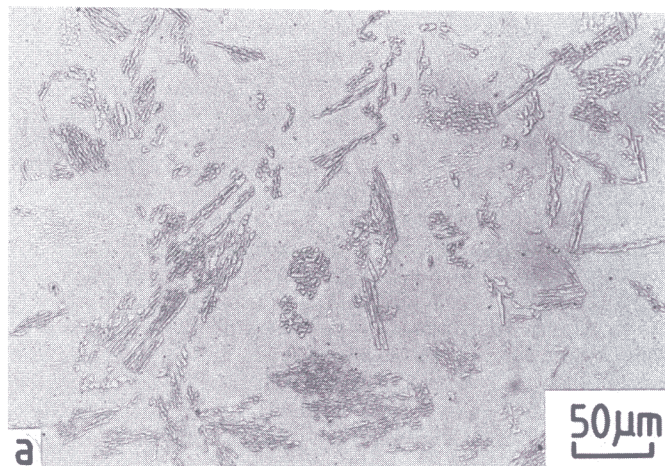


Fig. 3 - (a) Bainite obtained after austenitisation of weld metal at 950°C followed by isothermal transformation at 460°C. (b) Acicular ferrite obtained after austenitisation of weld metal at 1200°C followed by isothermal transformation at 460°C. Details of heat-treatments given in text.

elsewhere (6-9,11). The formation of Widmanstätten ferrite involves the redistribution of carbon between the parent and product phases, but bainite initially forms by a diffusionless mechanism and the carbon later partitions into the remaining austenite (6-9,11). This difference can be exploited to deduce whether acicular ferrite grows like conventional bainite or like Widmanstätten ferrite.

Considering first the diffusionless formation of bainite, the reaction should go to completion since there is no diffusion necessary. In practice, the whole of the austenite grain does not transform instantaneously because of kinetic restrictions (e.g. heterogeneous nucleation); even if the first plate forms without diffusion, it has an opportunity to reject its excess carbon into the residual austenite. Any further increment of transformation is therefore associated with a lower free-energy change, due to the higher carbon content of the austenite from which it has to form. Eventually, a stage is reached where the

transformation becomes thermodynamically impossible since the free energies of the residual (Fig. 4) austenite and bainitic ferrite of the same composition become identical (6-9, 11). The locus of such positions, as a function of isothermal transformation temperature defines the T'_0 curve, where γ and α_b (with a certain amount of stored energy associated with transformation strains) of the same composition have equal free energies. The corresponding curve for stress-free γ and α of identical composition is conventionally called the T_0 curve. The bainite reaction should therefore stop when the carbon concentration of the austenite (x_γ) reaches the level given by the T'_0 for the isothermal transformation temperature concerned (6-9,11).

The $Ae3'$ curve may similarly be defined for the growth of α_w , which involves the para-equilibrium transformation with the partitioning of carbon. If the small amount of strain energy associated with the back-to-back formation of mutually accommodating plates of Widmanstatten ferrite is ignored (9), the α_w reaction should stop when x_γ reaches the level given by the $Ae3'$ curve for the isothermal transformation temperature concerned. For any given T , the carbon concentration given by the $Ae3'$ curve is much larger than that given by the T_0 curve, so that a measurement of x_γ at reaction termination can clearly distinguish between α_w and α_b . Hence, our aim was to isothermally transform a weld metal in a temperature range where it is known to transform to just acicular ferrite, and by studying the maximum extent of transformation in terms of

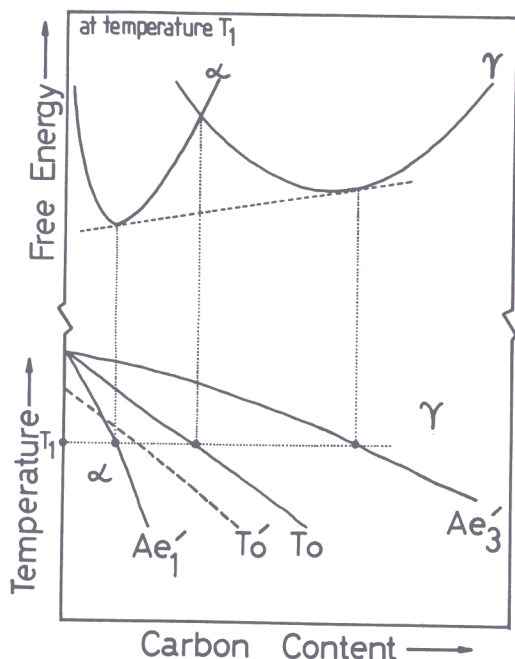


Fig. 4 - Schematic diagram showing the construction of phase boundaries from free energy data. The T'_0 line includes the effect of 400J/mol of strain energy due to transformation.

the above criteria, to decide whether α_a resembles Widmanstatten ferrite or bainite with respect to its growth mechanism.

DILATOMETRY AND THERMODYNAMIC ANALYSIS -
The dilatometric specimens were austenitised at 950 for 10 mins or at 1200°C for 30 mins before quenching to the isothermal transformation temperature. The isothermal transformation curves are presented in Fig. 5 and are seen to have a classical sigmoidal shape. It is immediately apparent that the maximum degree of transformation at any T increases as T decreases below the B_s temperature, showing that the weld metal exhibits the classical 'incomplete-reaction phenomenon' which is an identifying feature of bainitic transformation (6-9,11). From the data of Fig. 5, it is possible to calculate x_γ corresponding to the maximum degree of transformation, and these calculations are presented in Fig. 6, along with the relevant phase boundaries (calculated as in 6,9,11). The calculations use the following measurements: $e_\gamma = 1.15 \times 10^{-5} \text{ K}^{-1}$, $e_\alpha = 1.769 \times 10^{-5} \text{ K}^{-1}$, and the lattice parameter of ferrite in the weld, at ambient temperature = 2.8723 angstroms. It is clear that acicular ferrite formation ceases at the T'_0 boundary, providing strong evidence that the formation of acicular ferrite is diffusionless, the redistribution of carbon into the residual austenite occurring at a stage after growth. The results also imply that the strain energy accompanying acicular ferrite growth is of the order of 400J/mol (9).

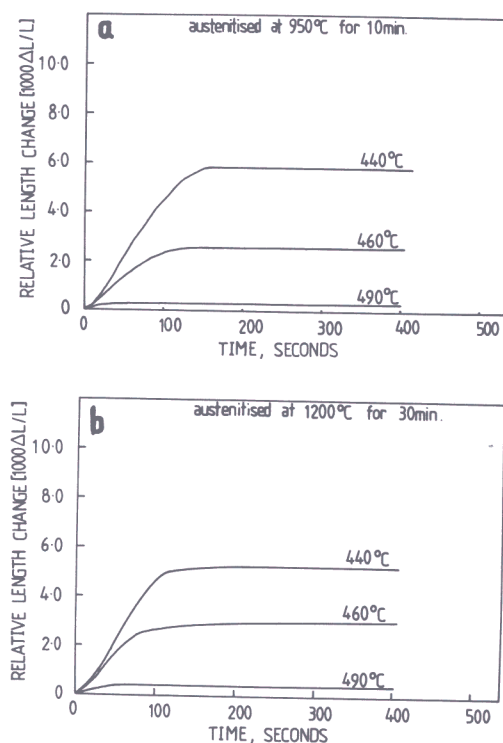


Fig. 5 - (a) Isothermal transformation curves for weld reaustenitised at 950°C. (b) Isothermal transformation curves for weld reaustenitised at 1200°C.

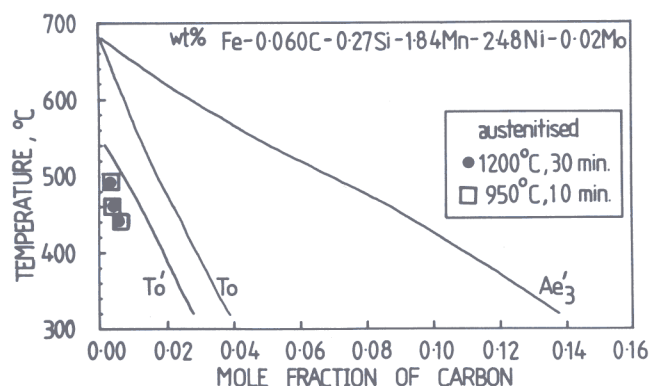


Fig. 6 - Phase diagram, with experimental data on the carbon concentration of austenite at the termination of isothermal transformation. The reactions stop well before the Ae_3' curves are reached.

CONCLUSIONS

It is found that the formation of acicular ferrite at any particular transformation temperature ceases as the carbon concentration of the residual austenite reaches the T_0' boundary, and that the transformation exhibits an "incomplete reaction phenomenon". These results provide strong evidence suggesting that the growth of acicular ferrite is diffusionless, with carbon partitioning into austenite after the transformation event. In this respect, acicular ferrite is identical to bainite. Evidence has also been provided to show that the morphology of acicular ferrite differs from that of bainite simply because the former nucleates intragranularly at inclusions within large austenite grains. It is concluded that acicular ferrite is in fact bainite which nucleates intragranularly in the large austenite grains typical of weld deposits.

ACKNOWLEDGEMENTS

The authors are grateful to Professor D. Hull for the provision of laboratory facilities at the University of Cambridge, to ESAB for the provision of weld samples and to the Chinese Ministry of Education for financial support.

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