

THE GROWTH RATE OF ALLOTRIOMORPHIC FERRITE IN  
LOW-ALLOY STEEL WELD DEPOSITS

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Abstract

Low-alloy steel weld deposits (manual metal-arc) containing different concentrations of carbon have been studied both theoretically and experimentally. The amount of allotriomorphic ferrite obtained at the austenite grain boundaries is found to decrease with increasing carbon concentration. By treating the growth of this ferrite in terms of the carbon diffusion-controlled migration of a planar austenite-ferrite interface, it seems possible to predict some of the observed microstructural variations as a function of chemical composition. The analysis is based on ferrite growth occurring under paraequilibrium conditions, without any redistribution of substitutional alloying elements during transformation; the concentration dependence of the diffusivity of C in austenite is fully taken into account. The results indirectly explain the variation of acicular ferrite content as a function of carbon; the variations seem to be independent of the oxygen and sulphur contents.

Key Words: Weld Microstructure, Oxygen Effect.

1 INTRODUCTION

It is generally accepted that the presence of substantial quantities of allotriomorphic ferrite ( $\alpha$ , often referred to as "grain boundary proeutectoid  $\alpha$ ") can lead to poor toughness in low-alloy steel weld deposits; the relatively large grains of  $\alpha$  offer little resistance to cleavage crack propagation. Allotriomorphic ferrite is the first phase to form on cooling below the  $Ae_3$  temperature and nucleates heterogeneously at the boundaries of the columnar austenite grains which are so typical of weld deposits. These boundaries rapidly become decorated with virtually continuous layers of ferrite, so that subsequent transformation simply involves the diffusional thickening of these layers, a process which can kinetically be treated in terms of the normal migration of planar austenite-ferrite interfaces.

Due to the relatively high cooling rates involved in the welding of low-alloy steels, and because of the low alloying element content of the resulting weld deposits, the growth of allotriomorphic ferrite is assumed throughout to occur under paraequilibrium<sup>1</sup> conditions. If interface attachment processes are not rate controlling, paraequilibrium transformation refers to the formation of  $\alpha$  from alloy austenite without any redistribution of substitutional alloying elements, even on the finest conceivable scale. Carbon does partition during transformation, such that its partial molar free energy is equal in both the phases at the interface. Growth therefore occurs at a rate controlled by the diffusion of carbon in austenite ( $\gamma$ ), with the ratio of substitutional/iron atoms being constant throughout the material, including at any interface regions.

The present work is study on microstructural evolution in a series of low-alloy steel weld deposits, in which the carbon alone was systematically varied; quantitative microstructural measurements are related to a model of ferrite growth based on paraequilibrium transformation theory.

## 2 EXPERIMENTAL DETAILS

The weld metals were deposited using the manual metal arc welding technique, the joint geometry being compatible with ISO 2560. The welding current and voltage used were 190A and 23V(DC+) respectively, the weld consisting of some 27 runs with 3 runs per layer; the interpass temperature was 250°C.

Optical microscopy was carried out on cross-sections of the welded specimens. The polished specimens were initially etched in a 1% aqueous Picric acid solution, followed by an etch in 1% nitric acid dissolved in amyl alcohol; the final etch was in Klemm's reagent. The microstructure was quantified according to the standard procedure recommended by the Welding Institute<sup>2</sup>, using a Swift point counter with a sample of at least 500 points per specimen. The results refer only to the top layers which represent the as-deposited microstructure.

## 3 RESULTS AND DISCUSSION

By varying the composition of the electrodes used (but keeping the welding conditions constant throughout) weld deposits with significantly different carbon contents were obtained; the chemical analysis results are presented in Table 1 and it is evident that the experiments were successful in keeping every other alloying element concentration nearly constant.

Table 1: Chemical Analysis results. All concentrations are in weight percent, except O and N which are in parts per million by weight.

	C	Si	Mn	P	S	Al	N	O
1	0.029	0.55	1.08	0.008	0.011	0.002	72	342
2	0.061	0.53	1.01	0.010	0.009	<0.001	79	307
3	0.080	0.58	1.15	0.010	0.010	0.001	52	310
4	0.100	0.55	1.02	0.010	0.010	0.001	62	267

The experimentally observed microstructural differences between the four samples are listed in Table 2.

Table 2: Volume fractions of various microstructural constituents.

	Allotriomorphic Ferrite	Widmanstatten Ferrite	Acicular Ferrite
1	0.59	0.28	0.13
2	0.40	0.31	0.29
3	0.29	0.21	0.50
4	0.24	0.14	0.62

The above classification ignores the presence of very small amounts of "microphases" (i.e., retained austenite, martensite, carbides etc.) and although such assessments can be subjective, allotriomorphic ferrite is easy to morphologically distinguish and is the most reliable determination.

### 3.1 Isothermal Transformation Diagrams

Time-Temperature-Transformation (TTT) curves, for the four alloys of Table 1, are presented in Fig. 1a; the curves are calculated according to Ref.3. The upper C curves, or diffusional C curves, refer to the time taken for the initiation of the allotriomorphic ferrite or of the pearlite transformation (the latter at sufficiently low temperatures). The lower or shear C curves refer to the initiation of displacive transformations, such as the Widmanstatten, acicular ferrite and bainite reactions<sup>3,4</sup>. It should be noted that the precise conditions for the occurrence of a particular transformation product depend on factors specified in Ref.4. The thermodynamically calculated<sup>5,6</sup> martensite start temperatures are also presented in Fig. 1, as horizontal lines.

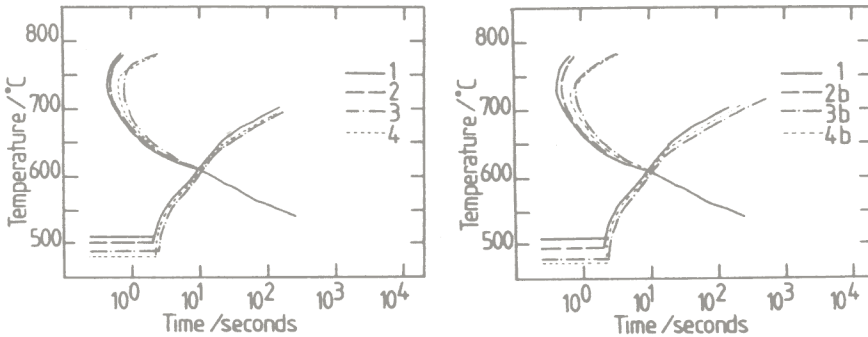


Fig. 1 (a) TTT curves for the steel compositions given in Table 1  
(b) TTT curves modified to account of 0.2 vol. frac. of  $\alpha$

The striking feature of Fig. 1a is that the four steels (Table 1) show little difference, as far as the kinetics of reaction initiation are concerned. It is clearly doubtful that the microstructural differences observed experimentally can be accounted for in terms of variations in TTT curves. To investigate the possibility that the TTT curves might change significantly after some initial degree of transformation to  $\alpha$ , further calculations (Fig. 1b) were carried out, based on the original steel compositions, but with increased carbon contents to allow for the partitioning of carbon into the residual austenite. Hence, Fig. 1b refers to steels which have already transformed to 0.20 volume fraction (v) of  $\alpha$ ,

giving  $\bar{x}$  an austenite carbon content ( $x_y$ )

$$x_y = \bar{x} + \{v(\bar{x} - s)\}/(1 - v) \quad \dots(1)$$

where  $\bar{x}$  is the average carbon content of the steel and  $s$  is the carbon content of the ferrite, taken to be 0.02 wt.pct. Again, Fig. 1b reveals that the TTT curves do not differ sufficiently to account for the data of Table 2.

### 3.2 Growth rate of Allotriomorphic Ferrite

Under the usual weld deposition conditions, most of transformation from continuously cooled austenite occurs over the temperature range 800–550°C. It is evident (Fig. 1) that the present steels do not have sufficient hardenability to avoid the formation of allotriomorphic ferrite on cooling, since the time required to initiate this reaction can be less than one second for all the steels concerned. It is possible that the initial formation of allotriomorphic ferrite controls the further development of the microstructure during subsequent cooling. A high volume fraction of allotriomorphic ferrite would necessarily limit the amount of austenite left for further different decomposition reactions to follow. The microstructural differences depicted in Table 2 might simply reflect variations in the growth rate of allotriomorphic ferrite in the four steels.

The carbon diffusion-controlled thickening of allotriomorphic ferrite by the movement of a planar  $\gamma$ - $\alpha$  interface, can be described by the equation<sup>8</sup>

$$q = \alpha_1 t^{1/2} \quad \dots(2)$$

where  $q$  is the half thickness of the allotriomorph,  $t$  is the time in seconds and  $\alpha_1$  is called the parabolic rate constant, obtained by solving

$$\{2(x_y^{\gamma\alpha} - \bar{x})/(x_y^{\gamma\alpha} - x_{\alpha}^{\alpha\gamma})\} \{(\bar{D}/\pi)^{1/2}\} = \alpha_1 \{ \exp[\alpha_1^2/(4\bar{D})] \} \{1 - \text{erf}[\alpha_1/(2\bar{D}^{1/2})]\} \quad \dots(3)$$

where  $x_y^{\gamma\alpha}$  is the paraequilibrium carbon content in the  $\gamma$ , given by the Ae3' lines of Fig. 2. Details of the calculations and of the significance of the  $T_0$  and  $T'_0$  curves can be obtained from Refs.5–7. The Ae3' line is the  $\gamma/(\gamma+\alpha)$  paraequilibrium phase boundary, given that the austenite and ferrite have the same substitutional alloying element concentrations.  $x_{\alpha}^{\alpha\gamma}$  is the corresponding paraequilibrium carbon content in the  $\alpha$ , and is always a very small quantity, and has been taken to be 0.02 wt.pct., irrespective of the temperature – the Ae1' lines are thus not drawn on Fig. 2.

The term  $\bar{D}$  is a weighted average diffusivity of carbon in  $\gamma$ , given by<sup>8,9</sup>

$$\bar{D} = \frac{\bar{x}}{x_y^{\gamma\alpha}} \int_{x_y^{\gamma\alpha}}^{\bar{x}} D dx / (\bar{x} - x_y^{\gamma\alpha}) \quad \dots(4)$$

where  $D$  is the diffusivity of carbon in austenite, a function of the amount of carbon in austenite and of the substitutional alloying element concentration<sup>9</sup>.

The results are summarised in Fig. 3, for a range of isothermal transformation temperatures. In reality, the specimens undergo continuous cooling rather than isothermal transformation; Fig. 3 fortunately shows that the trend in growth rate variation is maintained for all the temperatures studied, and that the lower carbon steels always transform faster. Hence, because there are no drastic changes in trends or growth rates with temperature, it is felt that the calculation of isothermal growth rates is justified.

The growth rate differences are very large and explain the data of Table 2. For example, at 700°C (an average temperature in the transformation range quoted earlier), Fig. 1 indicates that allotriomorphic ferrite should nucleate almost immediately, and should continue growing for at least 100s before the formation of Widmanstätten or acicular ferrite can begin (see the shear C curve). Hence, if we take  $t=100s$ , then  $q=40,26,20,18 \mu m$  for steels 1,2,3 and 4 respectively. If the  $\gamma$  grain size is taken to be about 200 $\mu m$ , a rough estimate of the volume fractions of allotriomorphic  $\alpha$  to be expected, before the onset of Widmanstätten ferrite or acicular ferrite formation, would be given by  $v=2q/\text{grain size}$ , or  $v=0.4,0.26,0.20,0.18$  for steels 1-4 respectively, in good agreement with the data of Table 2. Notice that the choice of 700°C for the calculation is not critical, since the differences in growth rates (Fig. 3) are similar for all the temperatures examined.

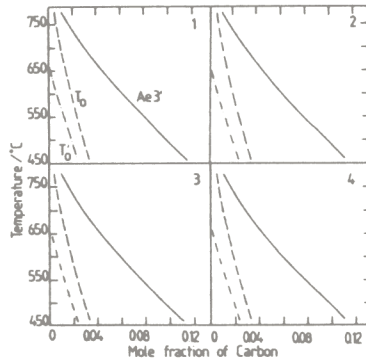


Fig. 2 Phase diagrams for the alloys of Table 1. The  $T_0$  and  $T_1$  curves are included just for information.

### 3.3 Effect of Oxygen and Sulphur Content

The growth rate analysis discussed above explains the decrease in volume fraction of allotriomorphic ferrite with increasing carbon content. It follows that the amount of Widmanstätten and acicular ferrite must increase with carbon content, because of the higher quantity of residual austenite left for further transformation. Microstructural variations like this are sometimes attributed to the effect of non-metallic inclusions on transformation kinetics<sup>10,11</sup>. However, the present work demonstrates that the microstructural variations are independent of inclusion content (proportional to O+S content, Ref.12). Alloys 2 and 3 have identical O and S concentrations but different microstructures. In fact there is little variation in (O+S) over the whole range of alloys examined; further work on these aspects is in progress.

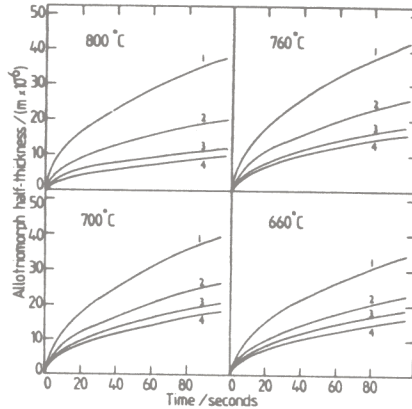


Fig. 3 Allotriomorph thickening analysis

#### 4 CONCLUSIONS

It appears that the changes in microstructure as a function of carbon content arise because it is the allotriomorphic ferrite growth rate that controls microstructural development in the low-alloy steel weld deposits examined. The lower the volume fraction of allotriomorphic  $\alpha$ , the higher is the probability of forming other constituents, such as acicular and Widmanstatten ferrite. The microstructural variations in the alloys examined seem to be independent of the oxygen and sulphur contents of the weld deposits.

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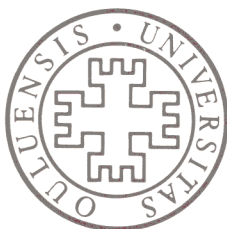
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