

The influence of alloying elements on the formation of allotriomorphic ferrite in low-alloy steel weld deposits

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The microstructure of the fusion zone of low-alloy steel weld deposits is complex [1, 2], but the formation of fairly uniform layers of allotriomorphic ferrite (sometimes called "grain boundary proeutectoid ferrite") at the boundaries between the columnar austenite grains is generally accepted to be detrimental to the toughness of the weld deposit [3]. Thick layers of such ferrite seem to offer little resistance to crack propagation.

There have been many studies on the role of alloying elements in influencing the formation of allotriomorphic ferrite in weld deposits, but these have largely been based on empirical correlations between v_α (the volume fraction of allotriomorphic ferrite) and the chemical composition. The significance of such correlations is not clear because the composition is often represented in terms of "carbon equivalents", which fail to adequately differentiate between ferrite and austenite stabilizing elements, and which take no account of the effect of alloying elements on the thermodynamics of transformation. The present work is an attempt to see whether a more fundamental parameter, the parabolic rate constant, can be used to approximately represent the effect of alloying elements on the development of allotriomorphic ferrite in weld deposits.

Allotriomorphic ferrite (α) is the first phase to form during the cooling of low-alloy steel weld deposits to temperatures below the Ae3 temperature [2, 4, 5]. The ferrite nucleates at the columnar-austenite grain boundaries, and these boundaries rapidly become decorated with a nearly uniform layer of α . Because of the very low C and substitutional (X) alloy content of low-alloy steel weld deposits, the time necessary to saturate the γ boundaries is so small that the formation of much of the α essentially involves the diffusional

thickening of layers of grain boundary ferrite. Hence it is reasonable to model this transformation solely in terms of growth, the growth occurring by the movement of essentially planar α/γ boundaries, leading to the thickening of the layers of α . If it is further assumed that growth is diffusion-controlled, then v_α should mainly depend on the parabolic rate constant α_1 , defined by the equation [6]:

$$q = \alpha_1 t^{0.5} \quad (1)$$

where q is the allotriomorphic half-thickness and t is the time, defined to be zero when $q = 0$. Most steel weld deposits are at least quaternary alloys, and the diffusion control could arise from the diffusion of carbon or of substitutional alloying elements [7]. However, for alloys higher than ternary alloys, the complexity of treating multi-component diffusion is beyond reason [8]. It is fortunate then, that the growth of α in dilute steels generally occurs without the bulk partitioning of X elements [9], especially when the growth rates involved are large [7], as is the case for weld deposits. In such circumstances, ferrite growth can occur at a rate controlled by the diffusion of carbon in the γ ahead of the interface. We assume therefore that α growth occurs by a carbon diffusion controlled paraequilibrium [10] mechanism so that X elements do not partition at all between the phases; carbon partitions to an extent which allows its chemical potential to be identical in all phases.

The parabolic rate constant is obtained by solving the equation [6]:

$$\begin{aligned} & [2(x\gamma^\alpha - \bar{x})/(x\gamma^\alpha - x_\alpha^{\alpha\gamma})] [(D/\pi)^{0.5}] \\ & = \alpha_1 \{ \exp [\alpha_1^2/(4D)] \} \{ 1 - \operatorname{erf} [\alpha_1/(2D^{0.5})] \} \end{aligned} \quad (2)$$

where $x\gamma^\alpha$ and $x_\alpha^{\alpha\gamma}$ are defined by the

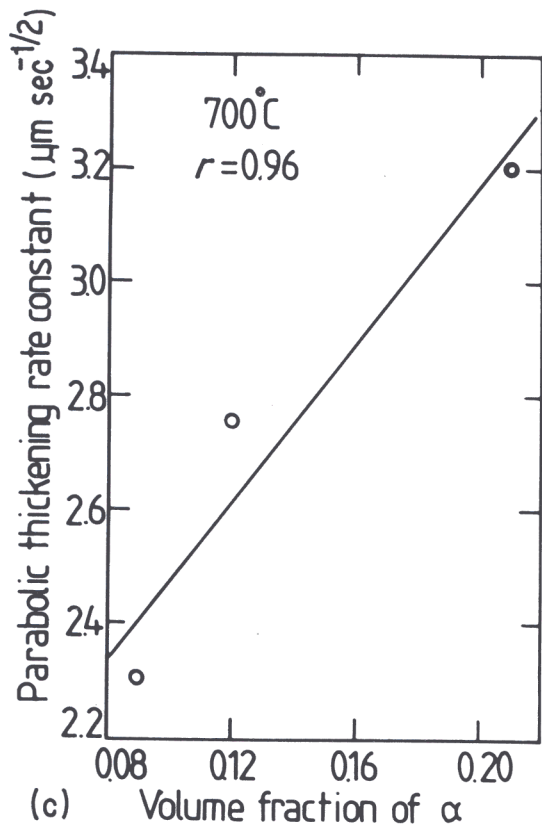
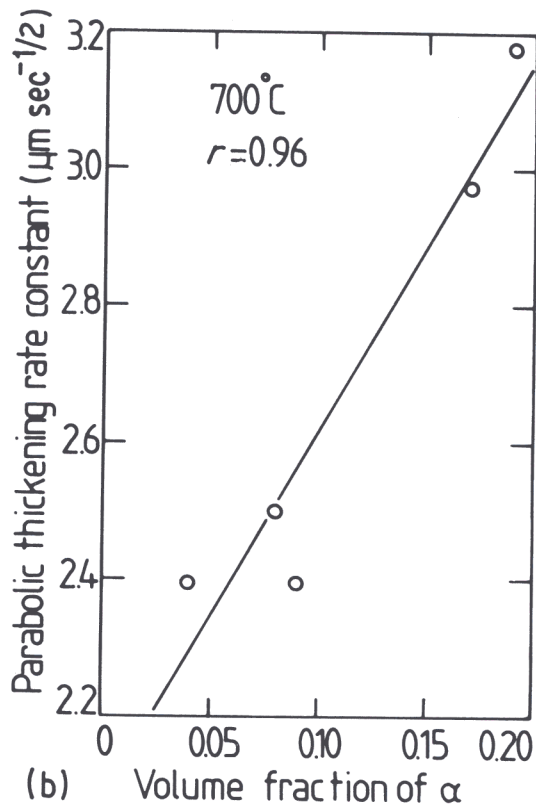
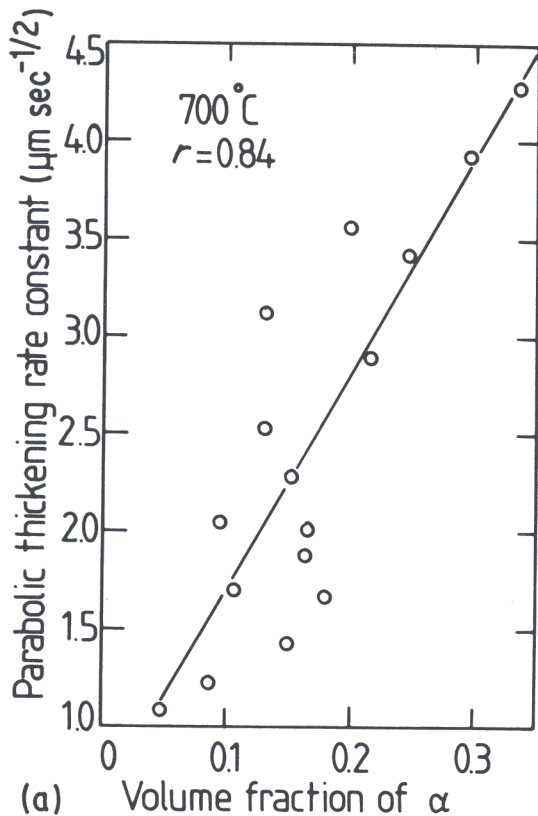


Figure 1 Plots of calculated parabolic thickening rate constant against experimental data on volume fraction of allotriomorphic ferrite. (a) Alloys 1 to 16 (data due to Evans [14]), (b) alloys 17 to 21 (data due to Abson [15]), (c) alloys 22 to 24 (data due to Abson [15]). The calculations are for a transformation temperature of 700°C , and r refers to the linear regression correlation coefficient.

paraequilibrium $\gamma/(\gamma + \alpha)$ and $\alpha/(\alpha + \gamma)$ phase boundaries, respectively, and represent the compositions of the γ and α at the interface. \bar{x} is the average alloy carbon concentration and T is the isothermal transformation temperature \mathbf{D} is a weighted average diffusivity of carbon in austenite, defined by the equation [11].

$$\mathbf{D} = \int_{x_\gamma^\alpha}^{\bar{x}} D \, dx / \bar{x} - x_\gamma^\alpha \quad (3)$$

The use of \mathbf{D} instead of D (the diffusivity of carbon in austenite) is necessary because D is strongly dependent on the C concentration, so that its magnitude is expected to vary as a function of the distance ahead of the interface [11].

Although substitutional alloying elements are assumed not to diffuse during transformation,

they influence growth rate by altering the thermodynamic driving force for transformation. This in turn affects x_{γ}^{α} and hence α_1 . X elements also alter the activity of carbon in austenite and so influence D . Since x_{α}^{γ} is always very small, it is taken to be the same as that for plain carbon steels (see Fig. 1b of [12]).

The allotriomorphic ferrite formed during welding develops over a range of temperatures. However, for dilute alloys, the trend in values of α_1 for different alloys is not affected [13] by T , so that the use of a constant T to obtain α_1 is justified for the purposes of comparing alloys. This procedure implies that any such comparisons must be for welds deposited under identical conditions (i.e. having identical cooling curves).

The experimental data used to test whether α_1 correlates well with v_{α} is listed in Table I, and is due to Evans [14] (alloys 1 to 16) and Abson [15] (alloys 17 to 24). The deposits are all Fe–C–Si–Mn alloys so that all the alloying elements can be expected to be in solid solution during α formation. The data are analysed in three sets, alloys 1 to 16, 17 to 21 and 22 to 24, each set being deposited via the manual metal arc welding tech-

nique but using different conditions. The detailed welding conditions are given in the original references together with the concentrations of various impurity elements.

Allotriomorphic ferrite in weld deposits usually forms over the range 800 to 600°C, and an average temperature of 700°C was chosen for the α_1 calculation. The paraequilibrium $\gamma/(\gamma + \alpha)$ phase boundary, necessary to obtain x_{γ}^{α} as a function of X element content, was calculated as in [16] and D was obtained using Siller and McLennan's [17, 18] model taking full account of the influence of X elements on the activity of C in γ .

Fig. 1 illustrates the results; it is clear that for low-alloy steel weld deposits, most of the variation in v_{α} as a function of composition can be rationalized in terms of the parabolic thickening-rate constant α_1 . The results also confirm the main assumption that in dilute weld deposits, it is the growth stage which is most important in determining v_{α} , and not the nucleation stage.

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TABLE I Experimental data (due to Evans [14] and Abson [15]) on v_{α} as a function of composition (wt %)

Alloy	C	Si	Mn	v_{α}
1	0.045	0.30	0.65	0.34
2	0.044	0.32	0.98	0.30
3	0.044	0.32	1.32	0.20
4	0.045	0.30	1.72	0.13
5	0.059	0.33	0.60	0.25
6	0.063	0.35	1.00	0.22
7	0.066	0.37	1.35	0.13
8	0.070	0.33	1.77	0.10
9	0.099	0.35	0.65	0.15
10	0.098	0.32	1.05	0.17
11	0.096	0.30	1.29	0.16
12	0.093	0.33	1.65	0.11
13	0.147	0.40	0.63	0.18
14	0.152	0.41	1.00	0.15
15	0.148	0.38	1.40	0.09
16	0.141	0.36	1.76	0.05
17	0.06	0.72	1.15	0.17
18	0.06	0.76	1.70	0.08
19	0.06	0.71	1.81	0.04
20	0.06	0.74	0.92	0.19
21	0.07	0.66	1.40	0.09
22	0.06	0.71	1.91	0.09
23	0.06	0.70	0.89	0.21
24	0.06	0.71	1.40	0.12

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