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DEVELOPMENT OF MICROSTRUCTURE IN THE FUSION ZONE OF STEEL WELD DEPOSITS

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ABSTRACT

A recent computer model for the prediction of microstructure in low-alloy steel weld deposits is modified to account for the presence of solidification-induced alloying element segregation. The new model is tested on Fe-Mn-Ni-Si-C alloys and the results indicate that segregation can significantly increase the amount of allotriomorphic ferrite in the microstructure by making its nucleation relatively easy.

INTRODUCTION

The design of electrodes and procedures for the welding of low-alloy steels is complicated by the large number of variables which can influence the final microstructure of the fusion zone. This microstructure is itself complex, consisting of allotriomorphic ferrite (α) which grows diffusionally at austenite (γ) grain boundaries, Widmanstätten ferrite (α_w) which nucleates at α/γ boundaries and grows by a displacive mechanism in the form of thin, wedge-shaped plates at a rate approximately controlled by the diffusion of C in the γ ahead of the interface, acicular ferrite (α_a) which nucleates intragranularly at inclusions but whose detailed transformation mechanism is not clear, and finally, microphases consisting of mixtures of martensite, degenerate pearlite and retained austenite, all resulting from the residual austenite remaining untransformed after α, α_w and α_a have formed. In this work we present a model which allows the quantitative prediction of weld microstructure and which includes effects due to the segregation of alloying elements during weld solidification.

MICROSTRUCTURE MODEL

A model which allows the approximate prediction of such microstructures for low-alloy steel weld deposits has recently been

lie along the centre of the prior-δ grains; the solute-depleted regions will therefore be at such γ/γ boundaries. If the first phase to solidify is γ on the other hand, then solute segregation should make α nucleation more difficult. In the present study, the alloys used should, on the basis of the carbon concentration alone, solidify with δ as the primary phase.

EXPERIMENTAL DETAILS

To design an experimental alloy which emphasizes the influence of segregation effects, calculations using the microstructure model (Bhadeshia et al., 1985a) were carried out on a series of hypothetical Fe-Si-Mn-Ni-C alloys in which Mn and Ni were systematically varied. The calculations were based on the cooling conditions appropriate for a 30 run manual metal arc weld deposited at a rate of 4mm/s with 180A, 23V, an interpass temperature of 240^{o}C, arc transfer efficiency equal to 0.775 and a joint geometry compatible with ISO-2560-1973. These are typical conditions for manual metal arc welds. The austenite grain size used for the calculations is 2a=70μm and is constant for all the alloys. The calculations all refer to the as-deposited microstructure only.

The alloy compositions and those of the corresponding depleted regions are presented in Table 1; calculated microstructures are given in Table 2. As expected, segregation seems to have the largest influence on highly alloyed steels. Alloy 5 should exhibit a pronounced change in microstructure due to segregation; an experimental alloy similar to alloy 5 was therefore designed and welded according to the above conditions.

Optical microscopy was carried out on cross-sections of the welded specimens. The quantitative metallography all refers to the as-deposited microstructure only, the measurements being confined to the last run which is unaffected by heat or dilution. The microstructure was quantified at a magnification of x700 using a Swift point counter with a sample of at least 500 points; the results thus suffer from a statistical error of the order of 2%.

RESULTS AND DISCUSSION

Its actual composition and measured microstructure are presented in Table 3 and Fig.2 illustrates its calculated TTT curves and paraequilibrium phase diagram. The TTT curves are calculated both for the homogeneous alloy and for the solute-depleted region and indicate that α formation should be accelerated by segregation. Table 3 shows that the results are in good agreement with the calculations based on the assumption that segregation aids the nucleation of α by raising T_h. Fig. 3 shows results from microanalysis experiments on the experimental alloy, obtained using scanning electron microscopy and a 'LINK-EDAX' microanalysis system. These results simply illustrate the presence of segregation, but cannot be quantitatively compared with the above theory since the experiments would have to be done on the scale of the α nucleus.

Table 1. Compositions (wt.%) of alloys whose microstructures were calculated, and of their respective solute-depleted regions.

	Alloy composition				Depleted region composition			
No.	C	Si	Mn	Ni	C	Si	Mn	Ni
1	0.07	0.40	0.60	0.60	0.07	0.18	0.47	0.39
2	0.07	0.40	0.60	1.20	0.07	0.18	0.47	0.78
3	0.07	0.40	0.60	1.80	0.07	0.18	0.47	1.17
4	0.07	0.40	1.00	0.60	0.07	0.18	0.78	0.39
5	0.07	0.40	1.00	1.20	0.07	0.18	0.78	0.78
6	0.07	0.40	1.00	1.80	0.07	0.18	0.78	1.17
7	0.07	0.40	1.50	0.60	0.07	0.18	1.17	0.39
8	0.07	0.40	1.50	1.20	0.07	0.18	1.17	0.78
9	0.07	0.40	1.50	1.80	0.07	0.18	1.17	1.17

Table 2. Calculated microstructures for alloys 1-9 (small amount of microphases, $V_v^m \equiv 0.02$ ignored). α layers are assumed to be continuous at the γ/γ boundaries. V_v^{α} refers to a corrected volume fraction, based on the calculated α volume fraction $V_v^{\alpha c}$ (Bhadeshia et al., 1985a).

	Homogeneous alloy			Heterogeneous alloy		
No.	V_v^{α}	V_v^w	V_v^a	V_v^{α}	V_v^w	V_v^a
1	0.44	0.36	0.20	0.47	0.35	0.18
2	0.25	0.30	0.45	0.39	0.27	0.34
3	0.00	0.26	0.74	0.29	0.22	0.49
4	0.28	0.25	0.47	0.40	0.23	0.37
5	0.00	0.27	0.73	0.28	0.24	0.48
6	0.00	0.15	0.85	0.00	0.15	0.85
7	0.00	0.16	0.84	0.24	0.14	0.62
8	0.00	0.20	0.80	0.00	0.21	0.79
9	0.00	0.13	0.87	0.00	0.13	0.87

Table 3. Calculated and measured microstructure of experimental weld deposit of composition Fe-0.07C-0.38Si-1.20Mn-0.94Ni-0.03920, wt.%. $\bar{L}_{tn}$ was measured to be 60μm, so that 2a=88μm. The composition of the solute-depleted regions was calculated to be Fe-0.07C-0.25Si-0.94Mn-0.42Ni wt.% since $k_{Si}=0.65$, $k_{Mn}=0.78$ and $k_{Ni}=0.44$ at 1450°C (assumed value of the liquidus temperature). Note that $T_h \rightarrow T_l = 705\text{-}600^{\circ}C$ for the solute-depleted regions, $T_h \rightarrow T_l$ being zero for the homogeneous alloy.

	V_v^{α}	V_v^w	V_v^a
Calculated for homogeneous alloy	0.00	0.16	0.84
Calculated for heterogeneous alloy	0.26	0.14	0.60
Actual microstructure	0.23	0.23	0.54

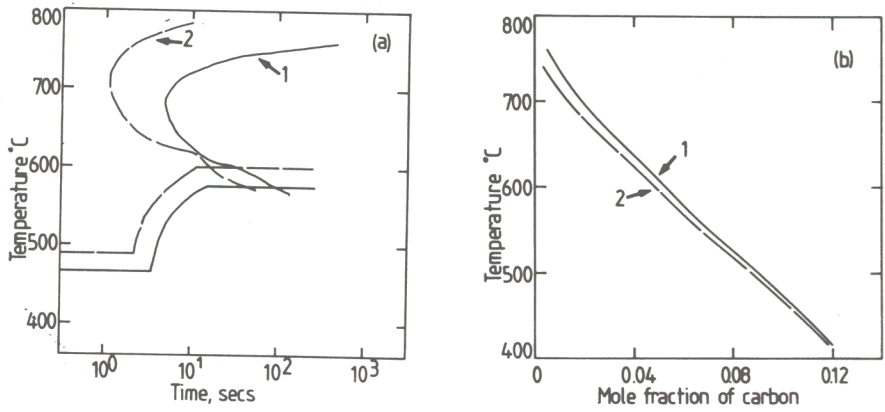

Fig.2 : a) Calculated TTT curves (Bhadeshia, 1982) for the initiation of transformation; curve 1 is for Fe-0.07C-0.38Si-1.20Mn-0.94Ni (wt.%), curve 2 for the solute-depleted region (Fe-0.07C-0.25Si-0.94Mn-0.42Ni wt.%). In each case, the upper 'C' curves represent diffusional transformation products (α, pearlite), the lower 'C' curves the displacive transformations (Bainite, α_w and perhaps α_a). b) Phase diagram for Fe-0.07C-0.38Si-1.20Mn-0.94Ni (wt.%); curve 1 represents the Ae3' $\gamma/(\gamma+\alpha)$ paraequilibrium phase boundary, curve 2 is the Ae3' boundary when ferrite contains 50J/mol of stored energy (used for the calculation of α_w growth rate - Bhadeshia et al., 1985a).

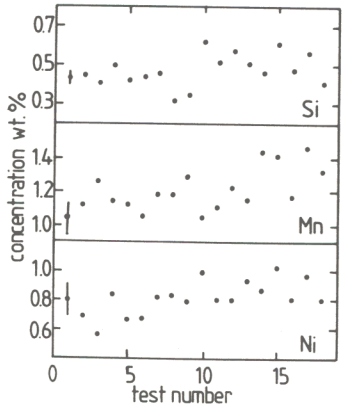

Fig.3: SEM microanalytical data (ZAF corrected) from tests done at random positions within the fusion zone of the top layer.

CONCLUSIONS

The results indicate that a major effect of substitutional element segregation during the welding of low-alloy steels is on enhancing the nucleation of α, and hence on the temperature range ($T_h \rightarrow T_l$) over which α can grow during cooling below Ae3. α can nucleate relatively easily in regions which are solute-depleted, especially if these coincide with the position of austenite grain boundaries. The effect is particularly pronounced if the alloy content of the steel is high enough to prevent the formation of α in the absence of segregation. The role of diffusion during cooling to T_h, on the exent of segregation, needs more detailed investigation; this in turn requires a method of predicting whether the primary phase to solidify is δ-ferrite or γ, since solute diffusivity in the former can be orders of magnitude higher than in the latter. Also, the position of the γ/γ boundaries with respect to the solute-depleted regions is expected to depend on the nature of the primary phase.

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