AN ANALYSIS OF THE PRIMARY MICROSTRUCTURE OF Cr and Mo CONTAINING LOW-ALLOY STEELS

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Abstract: Weld metal strength can be increased by alloying with Cr and Mo, without unduly sacrificing toughness. In this work we report theoretical and experimental work on the design of Cr and Mo containing welding consumables.

Introduction: The columnar austenite (γ) grain structure of steel welds can be represented as a honeycomb of hexagonal prisms, each of side length 'a' and length 'c', with c>a (1). During cooling, at the temperature $T=T_h$, the γ begins to transform, to layers of allotriomorphic ferrite (α) which grow by a diffusional transformation mechanism, at the γ/γ boundaries. As the temperature falls, diffusional transformation becomes sluggish and Widmanstätten ferrite (α_w) plates nucleate at the α/γ boundaries and grow into the γ by a displacive transformation mechanism. Acicular ferrite (α_a) then nucleates on inclusions (and also by autocatalytic effects) within the γ grains and grows in the form of thin plates. The small amount of remaining γ then decomposes either into degenerate pearlite and/or mixtures of martensite and retained γ. The volume fraction of these phases is small; they are called "microphases".

The microstructure can be estimated as a function of weld chemical composition and welding conditions (2,3). The model needed for this is able to account for the presence of C, Si, Mn, Ni, Mo, Cr and V. Having applied the model to Fe-C-Si-Mn and Fe-C-Si-Mn-Ni manual-metal-arc (MMA) welds, our aim was to test it against Cr and Mo containing welds. The model is summarised below and this time avoids an empirical correction to the volume fraction of α.

The multicomponent phase diagram is first computed using thermodynamic data. T_h is determined by using an additive reaction rule, the cooling curve and a computed Time-Temperature-Transformation diagram for the alloy concerned; the diagram has two 'C' curves, the upper one giving the time for the isothermal initiation of diffusional transformations such as α and pearlite, while the lower C curve

represents the initiation times for displacive transformations such as α_w, α_a. The cross-over point of the two C curves is the temperature T_1 below which displacive transformations are assumed to be kinetically favoured, so that the growth of α ceases and gives way to α_w and α_a formation. During anisothermal transformation the half thickness (q) of each such layer is given approximately by numerically integrating the function (2):

$$q = \int_{t=0}^{t=t_1} 0.5\alpha_1 \, t^{-0.5} \, dt \qquad (1)$$

where t_1 is the time taken for the weld deposit to cool from T_h to T_1, α_1 is the one-dimensional parabolic thickening rate constant, a complex function of many variables (2). 't' is the time, defined to be zero when q=0. The volume fraction v_α of α is then derived as a function of a and q:

$$v_\alpha = [4qC_3(a-2qC_3)/a^2] \qquad (2)$$

where C_3=tan(30°). At $T=T_1$, α_w formation begins approximately isothermally. If the time available for unhindered α growth is t_2, then the volume fraction v_w of α_w is given by:

$$v_w = 3.34G(a-2qC_3)t_2^2/a^2 \qquad (3)$$

where G is the α_w lengthening rate. In the absence of impingement with α_a, the time available for α_w to grow right across the γ grains is t_3, given by:

$$t_3 = 2[a \sin(60^\circ) - q]/G \qquad (4)$$

If t_3 is less than a critical time t_c, α_w can grow without impingement with α_a and t_2 in eq. 4 is set to equal t_3. If, however, $t_c<t_3$, α_w growth is terminated by impingement with α_a, and t_2 is set equal to t_c. t_c is experimentally found to be 0.211s for welds containing inclusions which nucleate acicular ferrite. As is conventional, microphases are included with α_a, so that $v_a=(1-v_\alpha-v_w)$ where v_a is the volume fraction of α_a. The effect of chemical heterogeneities is taken into account as discussed in (4).

Experimental: The welds (Table 1) were deposited using MMA welding (4mm diameter electrodes, joint design: ISO2560, plate thickness 20mm, 170A, 21V,DC+, 4mm/s, interpass temperature 250°C). The microstructure was analysed as discussed in (2). The cooling curves for the welds are calculated using eq. 1b of (3), with C_1=1324.8 and C_2=1.60.

Results and Discussion: Fig. 1a shows that there is reasonable agreement (correlation coefficient R=0.9237) between calculated and experimental data. The calculations confirm the experimental observations that the variation in microstructure caused by up to 0.6wt.% Cr and Mo is small. v_α is generally underestimated; the theory assumes one-dimensional paraequilibrium growth, and this has yet to be justified for Cr and Mo containing welds.

Table 1: Compositions and microstructure of undiluted weld metals.

Weld	1	2	3	4	5	6
C wt.%	0.049	0.053	0.058	0.053	0.052	0.050
Si wt.%	0.45	0.47	0.44	0.44	0.44	0.43
Mn wt.%	1.58	1.61	1.60	1.59	1.58	1.54
Cr wt.%	0.23	0.36	0.57	0.03	0.03	0.03
Mo wt.%	0.00	0.00	0.00	0.21	0.36	0.64
O ppm by wt.	358	397	400	363	369	338
N ppm by wt.	73	74	91	64	84	72
Volume fraction α	0.22	0.20	0.19	0.17	0.11	0.09
Volume fraction α_w	0.04	0.06	0.07	0.07	0.10	0.05
Volume fraction α_a	0.74	0.74	0.74	0.75	0.79	0.87
γ grain size '2a', μm	131	111	174	144	139	179

v_w is on the other hand consistently overestimated. Table 2 indicates that the growth of α_w is limited by hard impingement with α_a (since $t_2=t_c=0.211$) and this was confirmed metallographically; the α_w plates were in all cases found not to grow right across the γ grains, their growth being impeded by the intragranular formation of α_a. The discrepancy may arise because the nucleation rate of α_w may be reduced or the value of t_c may be reduced if α_a formation becomes easier as a consequence of the addition of Cr and Mo. The latter reason is inconsistent with the fact that the temperature difference T_1-B_s is larger for the present alloys than for the alloys used originally (2) in fixing t_c. It is tentatively concluded that Cr and Mo reduce the nucleation rate of α_w. This reduction can be accounted for by multiplying v_w by 0.289, and agreement between experiment and theory then improves (correlation R=0.9987), as illustrated in Fig. 1b.

Table 2: Calculated quantities for welds 1-6.

Weld	1	2	3	4	5	6
G_w, μm/s	129	100	76	118	120	126
T_h, ^oC	741	731	719	748	750	755
T_1, ^oC	609	609	609	609	609	609
t_2, s	0.211	0.211	0.211	0.211	0.211	0.211
Bainite Start B_s, ^oC	587	582	573	588	587	591

Because the microstructural differences between the welds are not very large, their mechanical properties (Table 3) are found to be similar and in all cases acceptable. For service at -60^{o}C, it may be advisable not to use high concentration of Mo.

Despite its limitations, the model has for the first time been shown to give fairly accurate estimates of microstructure in Fe-C-Si-Mn-Cr,Mo weld deposits. It can therefore be used to theoretically design such welds; Table 4 presents

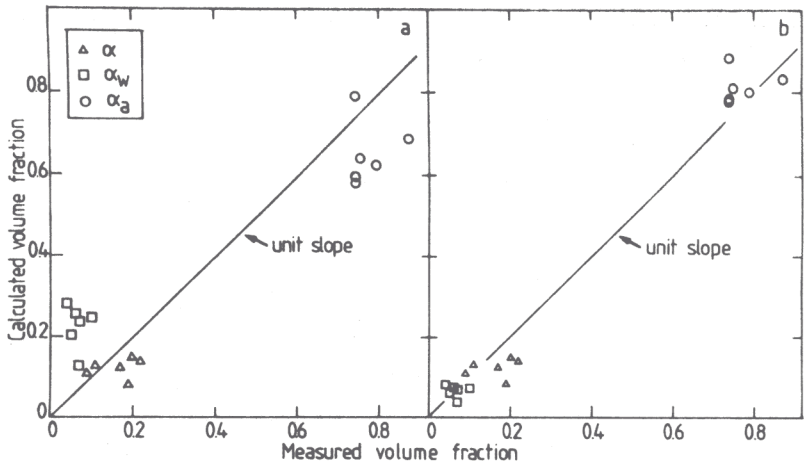

Fig. 1 Comparison between experimental and calculated results. In (b), the nucleation rate of α_w is reduced by a factor of 0.289.

some calculations of this kind.

Apart from showing the effect of combined additions of Cr and Mo, the calculations emphasize the strong effect of small variations in C, and the influence of lower Mn deposits. The calculations (Table 4) assume 2a=140μm, and include the 0.289 reduction in v_w.

Table 3: Mechanical properties. Each result is a mean of 5 tests.

Weld	1	2	3	4	5	6
Charpy toughness (20°C) J	209	213	205	206	216	194
Charpy toughness (-20°C) J	195	180	182	187	158	162
Charpy toughness (-40°C) J	165	150	134	161	114	69
Charpy toughness (-60°C) J	67	38	90	74	48	29
Vicker's Hardness, HV5	192	202	202	196	210	228

Table 4: Calculations of weld microstructure.

Weld	7	8	9	10	11	12
C wt.%	0.05	0.05	0.07	0.07	0.05	0.05
Si wt.%	0.45	0.45	0.45	0.45	0.45	0.45
Mn wt.%	1.60	1.60	1.60	1.60	1.00	1.00
Mo wt.%	0.20	0.40	0.20	0.40	0.20	0.40
Cr wt.%	0.40	0.20	0.40	0.20	0.40	0.20
v_α	0.12	0.13	0.09	0.10	0.16	0.17
v_w	0.06	0.07	0.04	0.04	0.10	0.11
v_a	0.82	0.80	0.87	0.86	0.74	0.72

Conclusions: The primary microstructure of Cr and Mo containing low-alloy steel welds can be estimated using phase transformation theory (2). There are, however,

some systematic discrepancies which need further investigation. In particular, Cr and Mo additions seem to reduce the nucleation rate of Widmanstätten ferrite. There is also a need to investigate whether Cr and Mo remain in solid solution during the variety of transformations which occur during cooling from the liquid state.

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