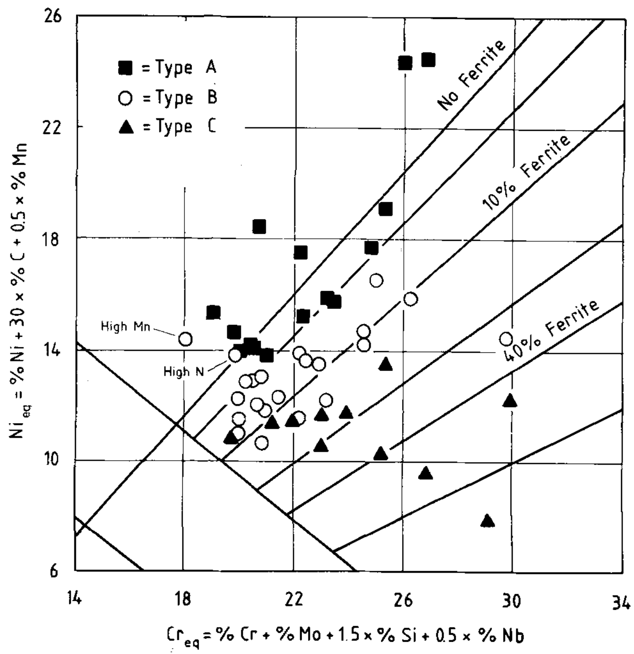

Fig. 4—Compositions and microstructural types of the experimental welds on the Schaeffler diagram. The fields of types A and C can be approximately defined by the ratio of $Cr_{eq}/Ni_{eq} \lesssim 1.48$ and $\gtrsim 1.95$ respectively. Note "high N" = 0.17 pct N and "high Mn" = 5.5 pct Mn.

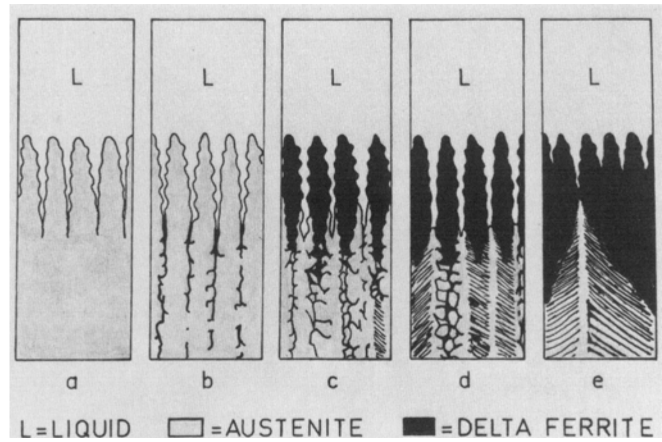

Fig. 5—Proposed solidification model for austenitic and austenitic-ferritic weld metals (schematic): (a) Type A, the weld metal solidifies completely to austenite and no further high temperature transformation takes place, (b) Type A, austenite is the leading phase and delta ferrite solidifies from the rest of the melt, (c) Type B, delta ferrite is the leading phase, austenite solidifies from the rest of the melt and a quick phase transformation $\delta \rightarrow \gamma$ takes place at high temperatures, (d) As (c), but a higher volume fraction of ferrite at room temperature is present, (e) Type C, the weld metal solidifies completely to delta ferrite and austenite forms through a solid state transformation.

looked on as consequences of different modes of solidification. In the welds of type A the austenite is the primary solidifying or leading phase and delta ferrite, if any, solidifies from the rest melt between the cells (Fig. 5(a) and (b)). Solidification in type B welds is probably the inverse of this, the delta ferrite being the leading phase and austenite solidifying from the rest melt. At lower temperatures the majority of the ferrite is transformed to austenite either by an equiaxial or acicular mechanism, depending on the supercooling of the delta ferrite (Fig. 5(c) and (d)). In type C welds only the ferrite solidifies directly from the melt and the austenite is precipitated from the solid ferrite at lower temperatures. It nucleates preferentially at the grain boundaries and grows into the interior of the grains by an acicular mechanism as a consequence of pronounced supercooling (Fig. 5(e)).

To summarize, in considering the microstructure at room temperature, one should emphasize that both solidification and the phase transformation $\delta \rightarrow \gamma$ have an effect on the structure in type B welds, whereas it is determined almost entirely by the solidification in type A welds and by the phase transformation $\delta \rightarrow \gamma$ in type C welds.

The solidification model proposed here for weld metal is supported by the work by Fredriksson[3] concerning directional solidification of an austenitic stainless steel of type 304 (Cr_{eq} = 19.8 and Ni_{eq} = 10.1). This steel solidifies primarily as ferrite, but a nitrogen addition from 0.06 to 0.18 to 0.25 pct depending on the cooling rate causes a change in the solidification sequence. This decreases the ratio of Cr_{eq}/Ni_{eq} approximately to the value, which corresponds to a transition from type B to type A microstructure in the welds (see Fig. 4).

The microstructure types present some interesting connections with welding practice and the properties of austenitic stainless welds. The majority of commercial filler metals and steels are balanced so that typically 4 to 10 vol pct delta ferrite is present at room temperature. According to DeLong the ferrite network becomes continuous at levels beginning somewhere in the range 4 to 7 vol pct.[2] These points of view agree well with features of type B welds in our classification. In addition, the range of the maximal resistance to hot cracking as determined by Hull[4] coincides excellently with the range of type B welds.

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Appendix 1. The Greek alphabet and names of the letters present in the paper

γ gamma
δ delta

1. T. Takalo, N. Suutala, and T. Moisio: *Met. Trans. A*, 1976, vol. 7A, pp. 1591-92.
2. W. T. DeLong: *Weld J.*, 1974, vol. 53, pp. 273s-286s.
3. H. Fredriksson: *Met. Trans.*, 1972, vol. 3, pp. 2989-97.
4. F. C. Hull: *Weld. J.*, 1967, vol. 46, pp. 399s-409s.

Correction to ***Met. Trans. A*****, 1979, vol. 10A**

Discussion of "Schematic Transformation Diagrams for Steel" by A. K. Cavanagh

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