CHAPTER 11

A PROGRAMME FOR FUTURE RESEARCH

This work tackled a broad range of issues. Throughout this project, however, the central aim has been to further our ability to quantitatively predict weld metal properties.

The discovery that the inclusions in weld deposits are not uniformly distributed is an important step towards being able to model the inclusion distribution in steel weld deposits. The precise mechanism, and the factors which influence this behaviour could be found using transparent organic media, seeded with inert powders to act as inclusions. In this way, the process of inclusion redistribution could be observed as it occurred, during solidification. Similarly, the effect of inclusion size distribution could also be investigated using low melting point alloys with carefully-characterised particle sizes. With a knowledge of the initial size distribution of the inclusions in the melt, and the dimensions of the primary columnar grains, the final distribution of these inclusions could be estimated. The quantity of secondary indigenous inclusions in the weld is also calculable. Their formation is a consequence of solute accumulation at the grain boundaries, and will be largely dependent upon alloy composition and the cooling rate of the melt. Together these give an overall picture of the total distribution of inclusions in the weld deposit from which many important parameters, such as the spacing, distribution and size distribution of the inclusion may be determined.

Detailed analysis of the inclusion distributions of a wide variety of welds is required, so that those factors that determine their ultimate volume fraction, size distribution, and mean particle size can be understood, since these are now recognized to be so important in controlling the strength and toughness of the weld. As precise information is the dispersion characteristics is needed, this investigation could probably best be done using carbon replica techniques in the transmission electron microscope. The experimental procedures involved are well-documented (A. S. T. M., 1973), and their usefulness in weld metal analysis is being recognized (Pacey *et al.*, 1982; Keville, 1983; Liu, 1987). The deposition of an amorphous layer of carbon allows the study of large specimen areas. Also, microanalysis of individual inclusions is facilitated since the matrix is absent, although an image of the microstructure is retained on the replica. Finally, replicas guarantee the extraction of large inclusions, which can fall out of foils.

It has been shown (Gretoft *et al.*, 1986) that, on a microscopic scale, the concentration of alloying elements in a weld is not constant. It would, therefore, be of particular interest to see how fluctuations in the local concentration of alloying elements in a given weld deposit can be related to its microstructure. This research would involve obtaining detailed composition profiles across the cell and columnar grain boundaries of a weld, and, on a larger scale, from the fusion boundary to the centre of the weld, since centreline segregation would tend to cause solute to accumulate ahead of the solidifying interface to give a slightly higher substitutional alloying element concentration in the centre (Davies and Garland, 1975). The influence of chemical segregation in relation to the position of the inclusions also merits investigation, but has yet to be studied.

A systematic study is required of the solidification behaviour of steel welds so that the identity of the solidifying phase can be predicted for a given alloy as a function of cooling rate. The general characteristics of weld metal solidification need to be established. For example, it would be interesting to see if the cooling curves for ferritic and austenitic solidification are visibly different in terms of the rate at which heat is removed from the weld. These results would then be combined with diffusion theory to allow the calculation of the growth rates of δ and γ dendrites as a function of alloy composition. Combined with the thermodynamic work described in this dissertation, the results would be represented in the form of isothermal solidification (solid/liquid) temperature-time-transformation curves. The appearance of the curves would alter depending upon what nuclei were present. For example, in a low carbon weld, nuclei of δ are present, and solidification as austenite is correspondingly difficult. If nuclei of austenite are present, however, then an austenitic solidification mode is likely to predominate.

A particularly interesting experiment in this line of research might be to weld a low-carbon electrode onto a high carbon steel plate, connected to a low carbon base plate. As with Weld 3.1, on the high carbon base plate, the weld metal would solidify as austenite. In such circumstances, it is possible that the austenitic mode of solidification would survive the transition across to the low-carbon base plate to give a weld solidification structure consisting of primary austenite grains on a δ-ferrite base plate. The details of such a structure would be of great general interest.

Practical measurement of the partitioning that occurs in the microstructure during solidification, perhaps by quenching in of weld deposits during welding would also be of interest.

The model for the strength of weld metals originated in this thesis should be developed in two ways. First, research must be undertaken to quantify how the microstructural strengthening contribution alters with temperature so that the variation of weld metal strength with temperature can be calculated. This could be done by measuring the strengths of a set of welds with carefully characterised microstructures. Ideally, testing should be carried out not only over a wide temperature range, but at a variety of strain rates, since testing at higher strain rates is equivalent to testing at lower temperatures. An ability to predict how yield strength varies with temperature has many potential applications; one example would be in the analysis of thermal stresses in welds. Large local stresses are known to exist in weld metals around inclusions due to differences in the coefficients of thermal contraction, and the elastic moduli of the inclusion and the matrix (Farrar and Harrison, 1987). However, the lack of directionality in the distribution of acicular ferrite, despite the strong temperature gradients, implies that thermal stresses are not a major factor in influencing inclusion nucleation.

Secondly, a series of systematic experiments to model the reheated regions of multirun weld deposits is needed, so that the microstructural changes that occur in the fusion zone during multilayer deposition can be analysed. A good way to do this would be to anneal a set of welds for increasing times in the temperature range of, say, 150-1000°C, and record how simple mechanical properties of the weld metals (*e.g.* hardness) vary with heat treatment. For temperatures above the eutectoid temperature, the effect of different cooling rates should also be measured, since these will affect the development of the reheated microstructure. (Such experiments might readily be carried out in a dilatometer, wherein cooling rates can be controlled very accurately). This work would provide data on how the microstructure of a weld changes for a given initial microstructure, heat treatment, and cooling rate, and, in this way, a detailed model could be constructed which would allow the strength of multirun welds to be predicted.

The upper shelf energy can sometimes be correlated with weld metal oxygen content (Devillers *et al.*, 1984). However, this is only an indirect measure of the inclusion content, more precise details of which will have to be taken into account.

Many excellent data are available in the literature (Kayali *et al.*, 1984; Bellrose, 1985; Thewlis, 1986) giving data for the Charpy toughness of welds together with detailed analyses of inclusion populations. Although, recent work has shown there to be a relationship between upper shelf energy and the ratio between the mean size of inclusions, and their mean spacing (Roberts *et al.*, 1982), a widely-applicable satisfactory model has yet to be produced. The actual advance of a crack tip in a matrix with a given inclusion distribution should be modelled using finite element analysis.

Particular notice should also be taken of the effect of nickel additions on impact toughness. Nickel is known to have a beneficial effect on weld metal toughness (Pokhodnya *et al.*, 1986), and this might be due to the softening effect which nickel was observed to noted in Chapter 6.

Also, since it is now possible to predict the flow stress and tensile strength of a weld as a function of temperature, an exciting advance from this would be to combine this work with established fracture theory to produce a model which would allow the calculation of the toughness transition temperature of a weld. Although, because the critical temperature at which, on cooling, cleavage failure becomes dominant is a function of, *inter alia*, the grain size and inclusion population of the weld (Bowen *et al.*, 1986), a more detailed model of the microstructure will be necessary, and stereological measurements of the grain structure of weld metal fracture specimens would also have to be made.

In the interest of applicability, the work on elongation and reduction in area should become more generalised. For example. the factors which control the strain-hardenability of a weld should be considered, since this affects percentage elongation. From a practical point of view, the work-hardening rate will vary with a variety of welding parameters, particularly preheat, since this will affect cooling rates and the degree to which stresses can be annealed out during cooling. A model which would allow reduction in area to be predicted from the chemical composition of the weld (wt% [O], [S], [Mn], &c.) rather than a measured inclusion volume fraction, should also be adopted. It would also be desireable to try to model ductility, terms of elongation and reduction in area, as a function of temperature.

The observed relationship between scatter in Chapy results and the uniformity of the microstructure is a particularly interesting finding, and the next step should be the application of this model to the calculation of the heterogeneity of multirun welds, which could be done most simply by treating the as-weld and reheat regions as hard and soft phases in a two-phase microstructure. This model could then be refined if found to be too simplistic. However, an important step in the development of this model, should be to take account of the *mechanical* heterogeneity of the microstructure. This is because a multiphase microstructure with phases of roughly equivalent toughness would be expected intuitively to exhibit less scatter during impact testing than one containing phases with vastly different properties, although the calculated heterogeneities of the microstructures might still be the same.

Finally, the discovery that lower bainite can form in low-alloy steel weld deposits, and its possible nucleation in inlcusions, is of enormous interest. The microstructure is so unusual that a comprehensive TEM investigation is called for.

REFERENCES

A. S. T. M. (1973), "Manual on Electron Metallography Techniques," Coordinators, G. N. Maniar and Albert Szirmae, A. S. T. M. Special Technical Publication 547, American Society for Testing & Materials, Philadephia, Pa., 20-23.

BELLROSE, A. (1985), *Aust. Weld. Res.*, (12), 4-10.

BOWEN, P., DRUCE, S. G. and KNOTT, J. F., *Acta Metall.*, **34**, (6), 1121-1131.

DAVIES, G. J. and GARLAND, J. G. (1975), *Int. Met. Rev.*, **20**, 83-106.

DEVILLERS, L., KEPLAN, D., MARANDET, B., RIBES, A. and RIBOND, P. V. (1984), "The Effects of Residual, Impurity and Microalloying Elements on Weldability and Weld Metal Properties", *[Proc. Conf.]*, Welding Institute, Abington, U.K., paper 1.

FARRAR, R. A. and HARRISON, P. L. (1987), *J. Mat. Sci.*, **22**, 3812-3820.

GRETOFT, B., BHADESHIA, H. K. D. H., and SVENSSON, L.-E. (1986), *Acta Stereol.*, **5**, 365-371.

KAYALI, E. S., PACEY, A. J. and KERR, H. W. (1984), *Can. Metall. Q.*, **23**, (2), 227-236.

KEVILLE, B. R. (1983), *Weld J.*, **63**, (9), Res. Supp., 253s-260s.

LIU, S. (1987), "Metallography and Interpretation of Weld Microstructure", *[Proc. Conf.]*, A. S. M. International, Metal Park, Ohio 44073, 87-131.

PACEY, A. J., KAYALI, S., and KERR, H. W. (1982), *Can. Metall. Q.*, **21**, (3), 309-318.

POKHODNYA, I. K., KORSUN, A. O., MESHKOV, Yu. Ya., PAKHARENKO, G. A. and SHEVCHENKO, A.V. (1986), *Avtom. Svarka*, **3**, (10), 1-5.

ROBERTS, W., AHLBLOM, B., and HANNERZ, N.-E. (1982), "Advances in the Physical Metallurgy and Applications of Steels", *[Proc. Conf.]*, 162-168. Discussion, 177.

THEWLIS, G. (1986), "3rd International Conference on Welding and Performance of Pipelines", *[Proc. Conf.]*, Welding Institute, Abington, U.K., paper 9, in press.