CHAPTER 9

SCATTER IN WELD METAL TOUGHNESS MEASUREMENTS

9.1 INTRODUCTION

In conjunction with the other research described in this dissertation, work has also been underway to quantify the factors that determine weld metal toughness. Increasingly stringent mechanical property requirements are being imposed in the manufacture of ferritic steel constructions, and a detailed knowledge of the factors influencing weld metal toughness is consequently vital. The Charpy V-notch test is used widely in quality control for determining the toughness of steels. The test is empirical, but is popular because it is both uncomplicated and cheap to perform. In general, much less energy is required to propagate a cleavage crack in a steel than is necessary for a ductile crack to grow. This is demonstrable by carrying out impact tests over a range of temperature, when the energy absorbed by the specimen when plotted as a function of temperature usually shows a sigmoidal behaviour, as the mode of fracture changes from brittle to ductile (Figure 9.1). Though the absorbed energy measured in this test cannot be used directly in quantitative assessments of the resistance of structures to brittle fracture, it can be used in a comparative manner for quality control.

The problem of predicting the impact behaviour from a knowledge of microstructure as yet seems insurmountable, although there are certain aspects of toughness which correlate strongly with microstructure, and which seem to have a straightforward physical basis. It is clear from published data that the scatter in the measured toughness values obtained from weld metal toughness results is frequently much greater than that obtained when measuring the toughness of plain carbon steels of equivalent chemical composition (compare Figure 9.2), and it can be hypothesised that this might be connected with the constitution of the microstructure. This phenomenon has been commented on before. Neville (1985) observed that many materials show variation in the measured values of their toughness or resistance to catastrophic crack propagation, and, in his own work on ferritic steels, he noted that the introduction of microstructural inhomogeneities, such as hard pearlite islands, can lead to a significant variation in measured fracture tough-

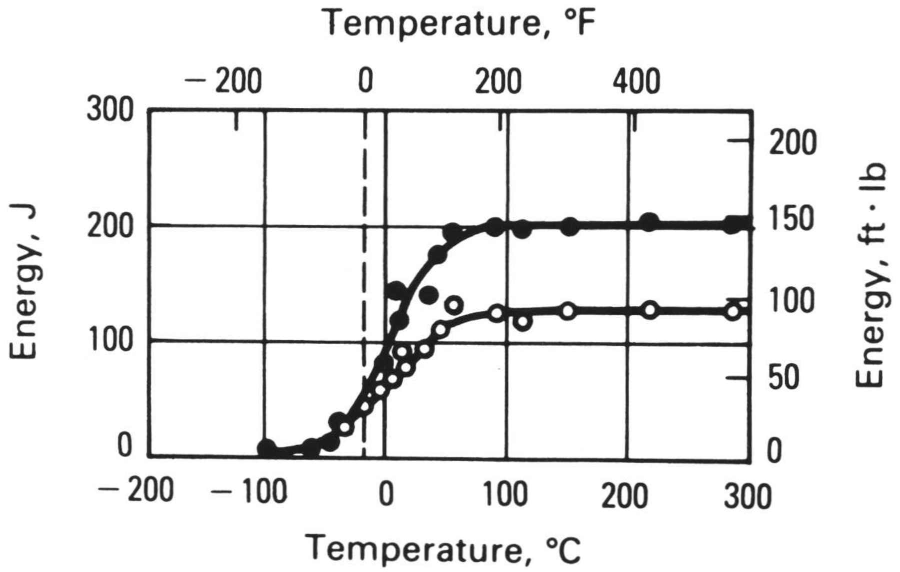

Figure 9.1: Changes in Charpy V-notch properties for conventional pressure vessel steel plate, tested in the transverse (○) and longitudinal orientation (•) with respect to the rolling direction. (After Dieter, G. E. (1978), "ASM Metals Handbook", 8, 262).

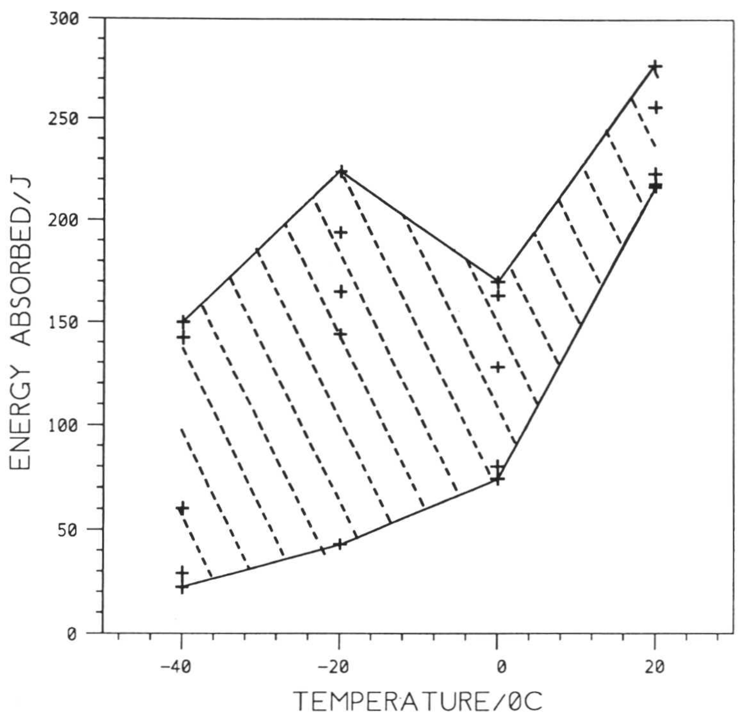

Figure 9.2: Typical Charpy toughness results for Fe-0.03C-1.08Mn-0.55Si (wt%) multipass low-alloy steel MMA weld, characterised by a large degree of scatter (shaded region). (Data: courtesy B. Gretoft, ESAB AB).

ness values during repeat tests on specimens of the same material. Garland (1975a; 1975b) observed that erratic and occasionally low as-welded toughness results have been recorded in both laboratory tests and procedural trials on a range of structural steels despite using welding materials generally approved for critical fabrications at weld heat inputs typical for these applications. He recorded that the as-welded mechanical properties achieved cannot be reconciled with either weld metal composition, or weld metal microstructure, as conventionally assessed in terms of area fraction of the major microstructural constituents. Another reason, suggested by Hayes *et al.* (1986), is that in tests on narrow welds, the crack has been observed to deviate into the adjacent material giving high absorbed energy measurements, which reflect the yielding properties of the adjacent material, rather than the toughness of the weld metal. It is the aim of this work to show that this behaviour *is* a consequence of the inhomogeneity of weld metal microstructures.

Several workers have commented that specific regions in the microstructure of weld metals are potential sources of failure. For example, Mardziej and Sleeswyk (1987) found weld metals of almost identical chemical composition, produced by the same welding procedure and consumables, differed significantly in toughness values, and attributed this to regions of local brittleness in the microstructure. Similarly, Thaulow *et al.* (1987) carried out a detailed examination of the surfaces of failed SMA weld metal fracture toughness specimens. It was found that the majority of the brittle fractures in their specimens had initiated from the primary weld metal. Widgery (1972) also found that cleavage cracks initiate preferentially in the as-deposited microstructure of low-alloy steel weld metal. Thus, it appears that the different microstructural morphologies encountered in a weld metal do provide local regions of strength and weakness, and any assessment of the factors that affect weld metal toughness should, therefore, include an analysis of this behaviour.

This work is part of a continuing project which aims for the prediction of the mechanical properties of low-alloy steel weld deposits from a knowledge of their chemical composition and detailed fabrication history. The first part of this work aims to show that a large amount of the scatter obtained in weld metal toughness experiments can be attributed to the nonuniformity of the weld metal microstructure. Secondly, it is demonstrated that the mechanical properties of regions within multirun welds can be expected to vary locally.

9.2 ANALYSIS OF SCATTER

In order to try to interpret the broad scatter that may be obtained in the impact testing of weld metals, it was first necessary to find a suitable way of representing scatter. The three most frequently used ways of rationalising scatter in results from the toughness testing of weld metals are to take an average of the Charpy readings obtained at a given temperature (*e.g.* Evans, 1980), measure the standard deviation (Drury, 1984), or plot the lowest Charpy readings obtained to focus attention on the lower ends of the scatter bands (Taylor, 1982).

An alternative to this was suggested by Smith (1983) who proposed a Scatter Factor to quantify any spread obtained in Charpy values, where

$$\text{Scatter Factor} = \frac{\text{Maximum energy} - \text{Minimum energy}}{\text{Average energy}} \times 100(\%) \tag{9.1}$$

However, such an *ad hoc* relationship cannot be used to provide statistically meaningful results. Yet, a difficulty in being more specific is that we are attempting to describe the toughness of a weld metal over a *range* of temperature, rather than simply rationalise the scatter in a set of data at one temperature. The best way round this is first to fit a curve to a given set of data.

An idealised impact energy temperature curve is sigmoidal in shape, and the curve-fitting could be done by one of three ways:

(i) the least squares method which gives equal weight to all points,

(ii) a weighted least squares method,

or (iii) by fitting the data to a logistic (log-related) curve.

In fact, the last method is the most common for a sigmoidal, rather than plain curve, and appeared to be justified over an (unweighted) least squares analysis in that the residuals between the observed and fitted log/temperature scales (discussed below) were approximately the same at all temperatures, *i.e.* at steep and shallow gradients. A weighted least squares analysis was not attempted since there was no clear way by which the weighting could be applied.

The sigmoidal curve has the form (Bronshtein and Semendyayev, 1973):

$$\ln(\frac{E}{E_{US} - E}) = \alpha + \beta T \tag{9.2}$$

where E = energy absorbed

E_{US} = upper shelf energy

T = temperature

and α and β are experimentally determined constants†.

A difficulty with fitting the sigmoidal curve to the experimental data is that the upper shelf energy needs to be defined. For this analysis, E_{US} was taken to be 2% above the maximum recorded impact value. This treatment was found to be satisfactory, and is a fair assumption since the upper shelf energy is essentially independent of temperature over the range of interest (Honeycombe, 1981a). α and β are determined by plotting the intercept and gradient respectively of a graph of $\ln\left(\frac{E}{E_{US}-E}\right)$ against temperature.

Regression analysis was performed using GLIM (General Linear Interactive Modelling) software developed by the Royal Statistical Society. The optimum values for α and β occur when the scatter of a given set of data points around a trial curve is a minimum. The scatter may be evaluated by calculating the *deviance* of the data, which is equal to the sum of the squares of the deviations of the sample observations from the mean. This may be expressed algebraically as

$$\sum_{i=1}^{n} \left[\ln(\frac{E_i}{E_{US} - E_i}) - (\alpha + \beta T_i)\right]^2$$

The best values of α and β are found, therefore, when this function is a minimum. However, in order to compare sets of data, it is necessary to consider the scale

†Perhaps the most common use of an equation of this form is in the quantitative description of reaction kinetics, (*e.g.* Johnson and Mehl (1939)).

parameter, where

$$\text{Scale Parameter} = \frac{\text{Deviance}}{\nu} \tag{9.3}$$

where ν is the number of degrees of freedom and illustrates the excess amount of data points available to be used in the regression analysis. It is defined as the number of data, n, minus the number of independent constraints on that set of data (Duckworth, 1968). The equation has two unknown constants, α and β, and so $\nu = (n - 2)$.

The scale parameter allows for the fact that the deviance of a large set of data will necessarily be greater than that of a smaller set of equally scattered data. The attraction of this method is that it quantifies scatter irrespective of the actual shape, and absolute magnitudes of the data, of the curve. It should be noted that this technique will give a false indication of the scatter associated with a given Charpy curve if only a few readings have been taken, and, irrespective of the proportions of various phases in the microstructure, if only three pairs of data are provided, the deviance will be zero! To guard against this, it is suggested that a minimum number of, say, ten readings per curve should be taken.

A computer listout of the program used for the evaluation of scatter is given in Appendix 6.

9.3 QUANTIFICATION OF HETEROGENEITY

Since it is believed that the variation in Charpy results obtained from similar welds at the same temperature depends upon the phases present in the weld, the inhomogeneity of a given weld metal microstructure would also need to be quantified. This was can be done by calculating the entropy, H, of a given microstructure (Large 1967; Karlin and Taylor, 1975).

Let X be a random variable assuming the value i with probability p_i, $i = 1, ..., n$. The entropy of X, as a logarithmic measure of the mean probability, is computed according to

$$H(X) = -\sum p_i \ln(p_i) \tag{9.4}$$

It should be noted that for $p_i = 1$, $H(X) = 0$. Conversely, the entropy is a maximum value, $\ln(n)$, when $p_1 = ... = p_n = \frac{1}{n}$.

It has been shown in earlier work that the microstructure of a weld metal can be taken as having three principal constituents: acicular, allotriomorphic, and Widmanstätten ferrite (Bhadeshia *et al.*, 1985; Abson and Pargeter, 1986; see also Chapter 5). It is important to emphasize that although α_a and α_w have similar strengths (Sugden and Bhadeshia, 1988), the weld metal microstructure cannot be treated as a two-phase microstructure (with α_a and α_w grouped together), since the *toughnesses* of the two phases are quite different. Therefore, the entropy of a given weld metal microstructure

$$H = -\left[V_\alpha \ln(V_\alpha) + V_a \ln(V_a) + V_w \ln(V_w)\right] \tag{9.5}$$

where V_α, V_a, and V_w are the volume fractions of allotriomorphic, acicular, and Widmanstätten ferrite respectively.

The entropy of the distribution quantifies the heterogeneity of the microstructure. H will vary from zero for an homogeneous material to ln3 (*i.e.* 1.099) for a weld with equal volume fractions of acicular, allotriomorphic and Widmanstätten ferrite. By multiplying by (1/ln3), the heterogeneity of the three phase microstructure of a weld may be defined on a scale from zero to unity. *i.e.*

$$\mathrm{Het}_3 = H \times 0.910 \tag{9.6}$$

A listout of the computer program used for the calculation of Het_3 is included in Appendix 6.

As a secondary experiment, it was also decided to see if the primary and secondary regions of multipass welds could be treated similarly. Here, the secondary region is taken to comprise that part of the microstructure consisting of partially reaustenitised and significantly tempered regions (Svensson *et al.*, 1988).

It follows that the heterogeneity of the assumed two-phase microstructure

$$\mathrm{Het}_2 = -\left[V_p \ln(V_p) + V_s \ln(V_s)\right] \times \left(\frac{1}{\ln 2}\right) \tag{9.7}$$

where V_p and V_s are the volume fractions of the primary and secondary regions respectively.

9.4 RESULTS

Initially, this work aimed to concentrate on analysing the primary (unrefined) regions of the weld metal. Data were taken from Watson *et al.* (1981) (Figure 9.3), and Bailey (1985) for two pass SA and triple arc SA welds respectively, and results for the estimation of scatter, and calculation of heterogeneity are given in Tables 9.1 and 9.2. Although Watson *et al.* (1981) referred to one of the phases observed as proeutectoid ferrite, this is a popular misnomer, and their description of this phase shows they meant *allotriomorphic* ferrite.

Figure 9.4 shows the relationship between the scatter observed in Charpy toughness values for the all-weld metal specimens and their microstructural heterogeneity.

Data for the calculation of Het_2 for the primary and reheated regions of multipass MMA low-alloy steel weld metals were taken from Abson (1982), and Taylor (1982). The work due to Taylor (1982) was particularly convenient since the Charpy data had been published numerically, rather than on a graph, and this facilitated the analysis. The percentage primary microstructure for Taylor's welds, which were in accordance with ISO-2560, could be estimated from a knowledge of the compositions and the heat inputs of the welds (Svensson *et al.*, 1988). It should be noted that the Charpy curves for W15SS and W15R (Abson,1982) could not be included in this analysis, because the upper shelf energies for these welds were unevaluated. The various steps involved in the calculation of the scale parameter, and Het_2 for these data are summarised in Tables 9.3 and 9.4.

Figure 9.5 shows calculated values for the scatter obtained in Charpy toughness experiments on multirun weld metal specimens, as a function of microstructural heterogeneity, treating the weld as a two-phase microstructure.

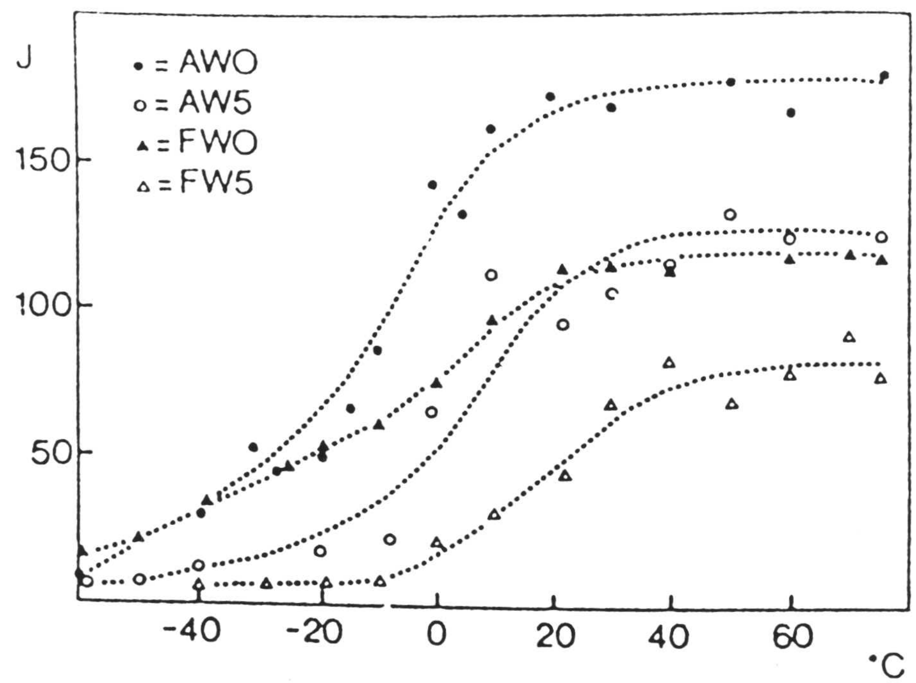

Figure 9.3: Charpy-V/Temperature curves, used for one scatter analysis. (After Watson, M. N., Harrison, P. L., and Farrar, R. A. (1981), *Weld. Met. Fab.*, **49**, (3), 101-108).

Weld	Reference	E_{US}/J	Deviance	ν	Scale Parameter
AWO	Watson *et al.*, 1981	183	6.54	13	0.503
FWO	"	122	2.29	12	0.191
AW5	"	134	6.96	11	0.632
FW5	"	94	5.34	11	0.485
W1	Bailey, 1985	107	6.65	10	0.665
W2	"	139	4.63	10	0.463
W4	"	123	1.89	10	0.189

Table 9.1: Estimation of scatter for all-weld metal specimens.

Weld	V_α	V_a	V_w	H	Het_3
AWO	0.29	0.67	0.04	0.756	0.688
FWO	0.09	0.89	0.02	0.380	0.346
AW5	0.50	0.47	0.03	0.788	0.717
FW5	0.25	0.08	0.68	0.817	0.744
W1	0.20	0.54	0.26	1.005	0.915
W2	0.18	0.69	0.13	0.830	0.755
W4	0.13	0.86	0.01	0.441	0.401

Table 9.2: Calculation of heterogeneity for all-weld metal specimens.

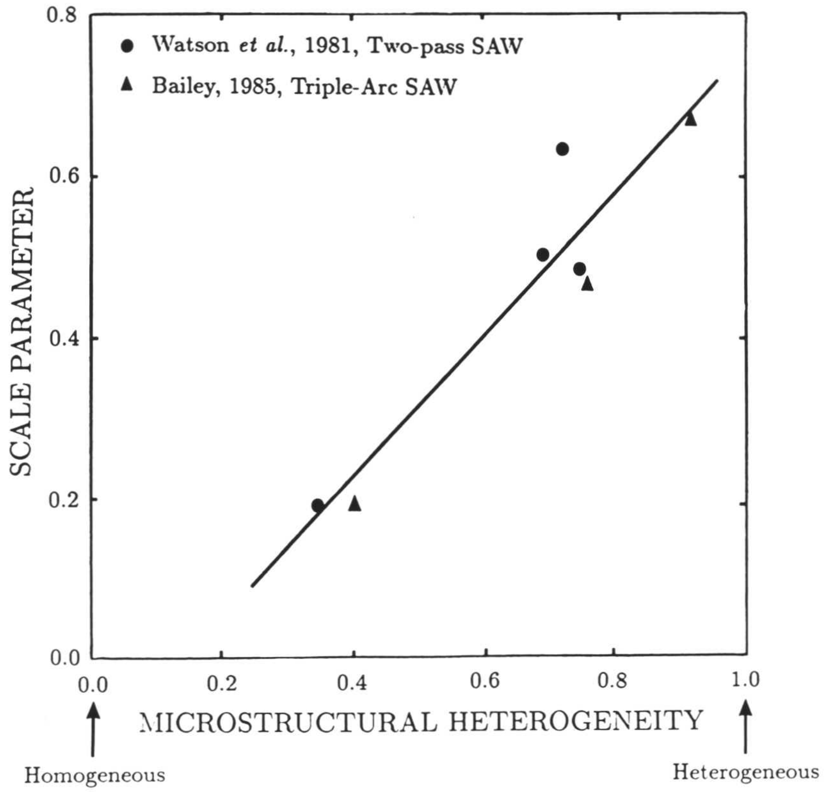

Figure 9.4: Showing the relationship between microstructural heterogeneity and scatter, as measured by the scale parameter of calculated Charpy curves. Each point corresponds to a complete set of Charpy results. The correlation coefficient is 0.94.

Weld	Reference	E_{US}/J	Deviance	ν	Scale Parameter
W18SS	Abson, 1982	184	1.10	8	0.138
W18R	"	181	2.07	8	0.259
W19SS	"	196	1.43	8	0.178
W20SS	"	200	3.41	8	0.426
W20R	"	199	2.86	8	0.358
W22R	"	197	4.60	8	0.575
E7016	Taylor, 1982	205	16.6	18	0.922
E7016-1	"	221	18.2	15	1.212
E7016-2	"	195	27.1	18	1.503
E7016-3	"	192	10.5	18	0.582

Table 9.3: Estimation of scatter for multirun welds.

Weld	V_p	V_s	H	Het_2
W18SS	0.38	0.62	0.664	0.958
W18R	0.32	0.68	0.627	0.905
W19SS	0.33	0.67	0.634	0.915
W20SS	0.35	0.65	0.647	0.934
W20R	0.43	0.57	0.683	0.985
W22R	0.24	0.76	0.551	0.795
E7016	0.30	0.70	0.611	0.881
E7016-1	0.37	0.67	0.636	0.918
E7016-3	0.42	0.58	0.680	0.981
E7016-3	0.48	0.52	0.692	0.999

Table 9.4: Calculation of heterogeneity for multirun welds.

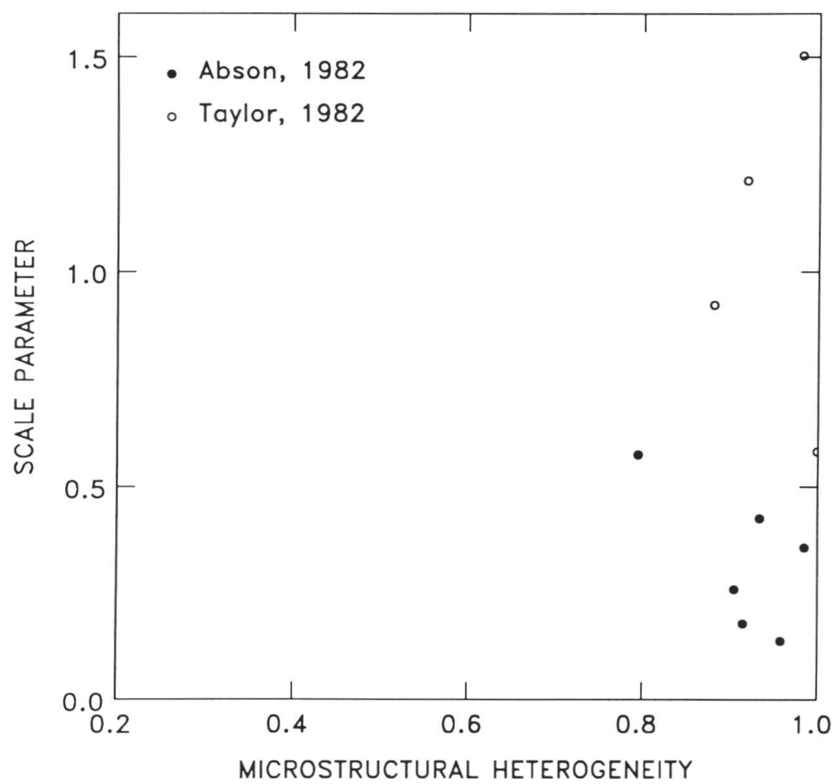

Figure 9.5: Microstructural heterogeneity versus scatter for the MMA multipass welds analysed.

9.5 DISCUSSION

It can be seen from Figure 9.4 that there is a strong relationship between the scale parameter, and microstructural heterogeneity for low-alloy steel all-weld metals. This work implies that a significant part of the observed scatter in weld metal Charpy results is attributable to the inhomogeneity of the microstructure, with larger scatters being associated with more heterogeneous microstructures. This result can be compared with the fracture toughness experiments of Thaulow *et al.*, (1987) who, for similar reasons, postulated that the most important factor in the COD testing of weldments is the positioning of the fatigue precrack.

The poor correlation for the multipass welds (Figure 9.5) highlights a limitation of this technique. Although, good results were obtained when the as-deposited microstructure was considered, the calculation of the heterogeneity of the microstructure of a given set of multipass welds must be carried out with caution. This is because the toughnesses of the two regions cannot be taken as independent. As was seen earlier (Figure 6.8), the strength of the secondary region is heavily dependent upon that of the primary from which it was formed. Of equal importance is the fact that the *difference* in the strengths of the as-deposited and reheated regions will depend upon alloy content, and will vary between steels, and it will therefore be necessary to take account of this in future work. Finally, Abson's welds had a comparatively small number of readings per Charpy curve, and this might have introduced a further discrepancy into the equation.

9.6 THE EFFECT OF TEMPERING ON WELD METAL HARDNESS

In order to model the mechanical properties of multirun welds, it will be necessary to understand more fully the nature of the mechanical inhomogeneities in the microstructure, for which *observable* differences in microstructure are only a guide. For example, in multipass arc welds, the superheated zone in which the metal is reheated to just below its melting point is believed to be potentially very weak, giving lower Charpy and CTOD values than would otherwise be expected (Gretoft and Svensson, 1986). Similarly, it has also been suggested (Svensson, 1986) that the double-renormalised region in multirun weld deposits is potentially a very weak region. A possible reason for this could be strain ageing. For example, strain ageing is believed to cause localised hardness in weld deposits, making root regions of MMA weld deposits harder and stronger than subsurface regions (Abson, 1982). If strain ageing were to be found to cause a small region of high hardness in a weld, this would then be a potential source of weakness.

To illustrate this point, samples from the top beads of three ISO-2560 welds, used elsewhere within this dissertation (Welds 6.2, 6.3 and 10.1), and whose compositions are given in Table 9.5, were extracted.

Weld ID.	C	Mn	Si	P	S	Cr	Ni	Mo	Ti	Al	N (ppm by wt.)	O (ppm by wt.)
	Composition, wt%											
6.2	0.10	1.56	0.42	0.015	0.007	0.04	0.04	0.01	0.013	0.015	119	262
6.3	0.15	1.57	0.45	0.012	0.007	0.04	0.03	0.01	0.014	0.015	96	193
10.1	0.32	1.65	0.48	0.015	0.005	0.03	0.03	0.01	0.018	0.015	64	141

Table 9.5: Concentrations of alloying additions in Welds 6.2, 6.3, and 10.1.

The welds were then tempered at temperatures up to 600°C for one hour. The specimens were quenched upon removal from the furnace to obviate any diffusion during cooling. (It should be emphasized that quenching would not cause any change in microstructure, because tempering was carried out below the Ae_1 temperature). Twenty hardness measurements (Vickers 10kg) were then made of the top bead of each of the weld metal specimens. The results obtained are summarised in Table 9.6, and plotted in Figure 9.6.

1 hour at T°C	VHN(10) Weld 6.2	VHN(10) Weld 6.3	VHN(10) Weld 10.1
21	232	252	299
290	277	262	308
420	271	254	304
502	238	253	297
605	239	251	281

Table 9.6: Hardness readings (VHN(10)) (with 95% confidence limits) for Weld 6.2, 6.3, and 7.1 after 1 hour at four different temperatures.

It can be seen that in all three cases, clear evidence of strain ageing, in terms of an increase in recorded Vickers hardness, has been obtained. The increases in hardness, as a result of the short tempering treatment, correlate with the nitrogen

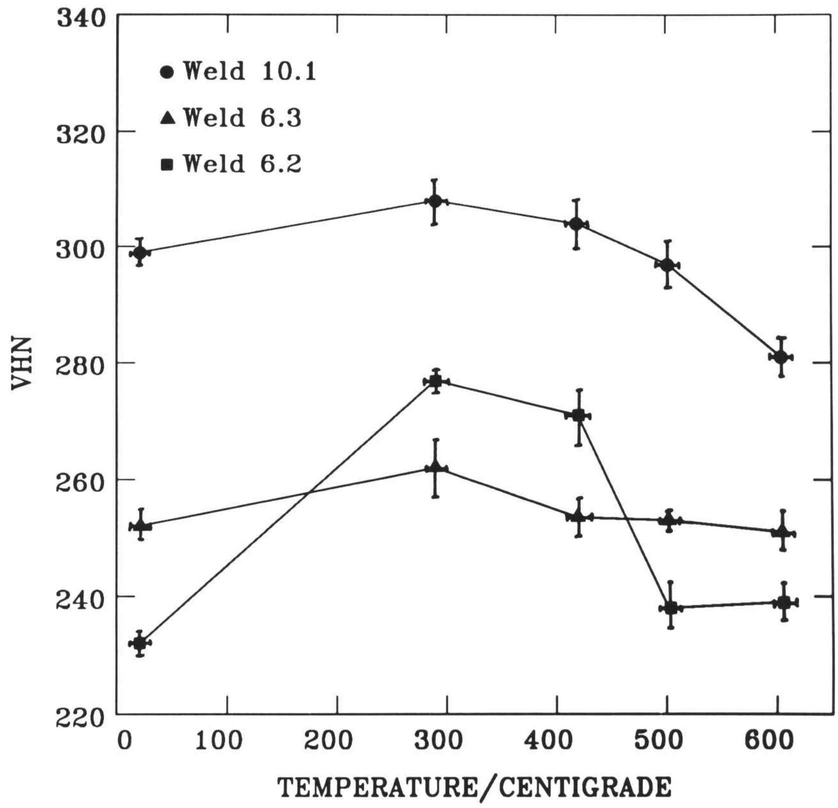

Figure 9.6: The hardnesses of Welds 6.2, 6.3, and 10.1 (with 95% confidence limits) after tempering for one hour at a temperature T. (The thermocouple accuracy has been taken as ±10°C, although, in reality, this probably underestimates its accuracy).

contents of the three welds (see Table 9.5), and can be construed to be due to the migration of nitrogen atoms to dislocations in the weld metal (Honeycombe, 1981b). Thus, Weld 6.2 increased more in strength than Weld 6.3, and Weld 10.1 increased in strength only slightly. Whatever the mechanism, these results imply that the strength of those regions of a multirun weld metal immediately below the fusion boundary which experience an equivalent tempering treatment during welding (*i.e.* equivalent in terms of the combination of tempering temperature and time) will be greater than that of the as-deposited weld metal. Thus, regions of local hardness will exist within the microstructure of multirun weld deposits where they will be liable to influence the fracture behaviour of that weld metal.

9.7 SUMMARY

A new method of interpreting weld metal toughness data characterised by wide scatter over a range of temperatures has been proposed. The microstructure of all-weld metal specimens has been treated as consisting of three independent phases: allotriomorphic ferrite, acicular ferrite, and Widmanstätten ferrite. Comparison with experimental data from the literature has shown that for all-weld metal specimens the scatter in weld metal toughness results can be related to the composition of the microstructure, and that the scatter observed is not wholly due to experimental error, but is a quantifiable function of the microstructure. For multirun weld metal specimens, however, agreement was poor. This can be attributed to two reasons. Firstly, that the model is only suitable when the toughnesses of the phases comprising the weld metal microstructure are non-interdependent, and secondly, as has also been shown, the microstructure of multirun welds will be likely to contain areas of localised hardness within regions of the same microstructure, and these will influence the toughness values recorded.

This work should permit the better design of experiments for the investigation of impact transition curves. It is also possible to estimate the error inherent in Charpy toughness results as a function of microstructure, and to plot a theoretical scatter band corresponding to scatter for mixed and homogeneous weld microstructures. This method could also be applied to aid the interpretation of weld metal COD results. Note, however, that at this stage the correlation between scatter and microstructure is empirical.

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