Fig. 7.11 Enlarged image of Fig. 7.4, indicating the region where bainitic ferrite appears to grow directly from the allotriomorphic ferrite.

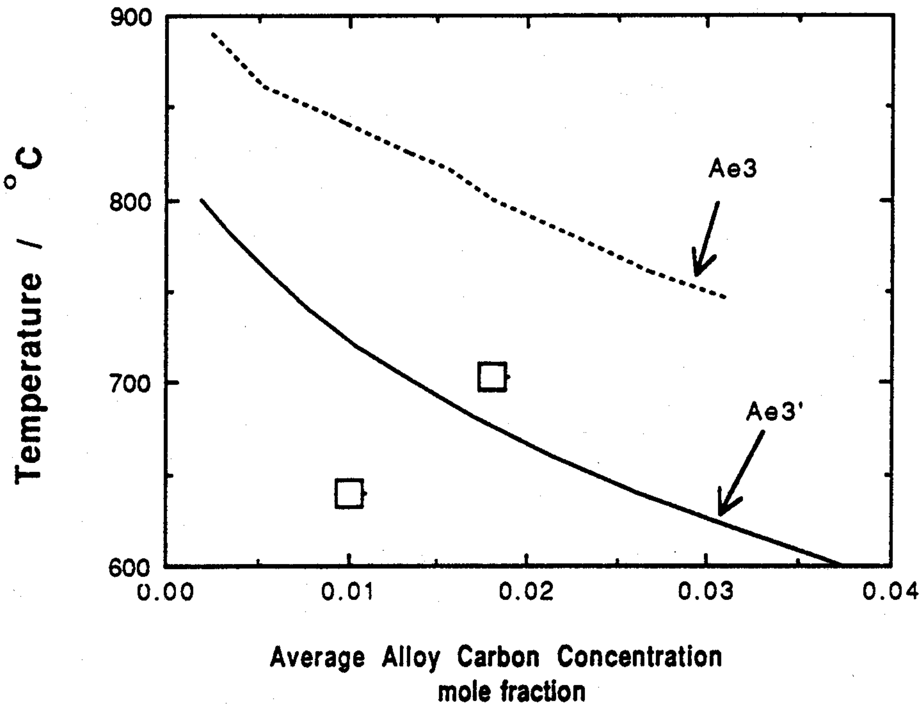

Fig. 7.12 Plot of calculated Ae3 and Ae3' phase boundaries for various average alloy carbon concentration for a Fe-2Si-3Mn (wt.%) alloy system (MTDATA, 1989, Bhadeshia and Edmonds, 1980). The square symbols indicate the composition of alloys A1 and A2 and temperature at which they were heat treated to form allotriomorphic ferrite.

Chapter 8

Stress and the Acicular Ferrite Transformation

8.1 Introduction

In spite of the microstructural differences, there is considerable evidence that acicular ferrite and bainite have similar transformation mechanisms. The microstructures differ only because of their nucleation site. Some of the similarities between bainite and acicular ferrite were summarised in section 1.2.6. In addition, results of previous chapters indicated following;

1. Both reactions stop when the austenite carbon concentration reaches a value where it becomes thermodynamically impossible to achieve diffusionless growth (chapter 4).
2. Acicular ferrite only forms below the bainite–start temperature.
3. The elimination of austenite grain surfaces by covering them with inert allotriomorphic ferrite leads to a transition from a bainitic to an acicular ferritic microstructure (chapters 4 & 5).

Transformation to both acicular ferrite and bainite causes displacements which are characterised as invariant–plane strains with large shear components (Fig. 8.1). Consequently, the growth of a plate of acicular ferrite or bainite is confined to a single austenite grain (*i.e.,* it is hindered by a grain boundary) since the coordinated movement of atoms implied by the shape change cannot, in general, be sustained across a border between grains in different crystallographic orientations. A further implication is that plates of acicular ferrite, like bainite, *always* have an orientation relationship with the parent phase which is within the Bain region. This is not necessarily the case when the transformation occurs by a reconstructive mechanism.

It is this last characteristic, the invariant–plane strain shape deformation, which forms the basis of the present investigation. Such a transformation is displacive, and can be regarded as a deformation mode which in addition to accomplishing a shape change, also alters the nature of the lattice. During transformation induced by a decrease in temperature, it is the chemical driving force which causes the deformation. However, the application of an appropriate external stress can provide a mechanical driving force which stimulates transformation without the need to undercool. Such effects are well established for bainite, where a tensile stress favours the development of crystallographic variants whose shape deformations comply most with the applied stress.

The purpose of this investigation is to verify the existence of a similar effect for acicular ferrite. As will be seen later, the specific kind of stress response investigated provides good evidence for the displacive mechanism of the transformation. Acicular ferrite is a prominent phase in steel weld metals (primarily because they contain nucleating inclusions). There is a possibility that the residual stresses associated with welding may influence the development of the acicular ferrite microstructure.

The only previous study on the relationship between stress and acicular ferrite in mixed microstructures, is by Dallam and Olson (1989). The stress was mostly generated by cooling constrained samples, the transformation being followed by monitoring the sample diameter. Dallam and Olson concluded that there is little influence on the volume fraction of acicular ferrite.

As discussed later, a much more sensitive method of following stress–affected displacive transformation is to monitor the transformation strains along more than one direction (Bhadeshia *et al.*, 1991).

8.2 EXPERIMENTAL METHOD

Alloy 77 is chosen for this study. The welding conditions and the final deposit chemistry are listed in Table 2.1 and Table 2.2. The weld metal, in the as deposited conditions transforms to a bainitic microstructure, in spite of the presence of numerous inclusions of the kind necessary to nucleate acicular ferrite. This is because, in the absence of any allotriomorphic ferrite at the austenite grain boundaries, the boundaries are free to nucleate bainite which forms at an overwhelming rate and succeeds in stifling the intragranular formation of acicular ferrite. By reheating the weld metal and then partially transforming the austenite at an intermediate temperature (T_1) to form a thin, uniform layer of allotriomorphic ferrite at the austenite grain surfaces grain boundary nucleation of bainite can be stifled. Thereby the remainder of the austenite transforms to intragranularly nucleated bainite as the alloy is held at a temperature (T_2) below the bainite–start (B_S) temperature (chapter 4).

This two–stage heat treatment (Fig. 8.2) provides a convenient way of generating an acicular ferrite microstructure in a controlled manner. The calculated transformation temperatures in chapter 5 are taken as an aid to the design of the heat treatment (Table 8.1). The calculation methods have been discussed before.

The heat treatment can easily be carried out in the *Formaster* thermomechanical simulator used in the present studies. The machine is equipped to simultaneously monitor and record the diameteral ($\frac{\Delta R}{R}$) and longitudinal strains ($\frac{\Delta L}{L}$), in addition to time, temperature and load data. It can be programmed to automatically carry out the specified thermomechanical treatments. The samples were in the form of 8 mm diameter cylinders 12 mm in length, made from the weld metal by rolling, swaging and machining (Fig. 2.4). The specimen preparation has been presented in chapter 2. The simulator experiments were carried out with the specimen chamber filled with argon. The rapid temperature changes needed between T_1 and T_2, and between T_2

Table 8.1 Calculated transformation data for alloy 77.

Ae_3	828 °C
W_S	680 °C
B_S	546 °C
M_S	427 °C

and ambient temperature were achieved using argon gas jets. Since the equipment is fully capable of monitoring the transformation during these rapid changes, it was confirmed that no reaction occurred during the gas quenching process, so that all the transformation could be attributed to isothermal reaction at T_1 and T_2. The thermomechanically treated samples were examined in both their transverse and longitudinal sections using the usual metallographic techniques. The austenite grain sizes were determined to be 151 ± 23 μm, defined as the mean lineal intercept of random test lines superimposed on the microstructure.

8.2.1 The Applied Stress

The stress was only applied the instant the sample reached T_2, in order to influence the development of acicular ferrite. It was intended in this work to limit the applied stress to a value below the yield strength of austenite. Any plastic deformation can complicate interpretation because the resulting defect structure can either assist or interfere with the progress of transformation. Published data on the yield strength of austenite as a function of temperature (Fig. 8.3) were extrapolated to estimate its yield strength to be approximately 500 MPa over the temperature range 500–550 °C. Even if the extrapolation is unjustified, the yield strength cannot be less than about 200 MPa, which corresponds to experimental data from 750 °C (Fig. 8.3). Consequently, the experiments were carried out at two load levels, both of which should be well below the austenite yield strength at T_2: the chosen loads correspond to stress levels of 116 and 174 MPa. The results of these experiments should be compared against transformation without the influence of any applied stress. However, under the conditions of the experiments, the longitudinal strain cannot be recorded unless a small stress of about 12 MPa is applied, since the cross head uses this signal to follow the dimensional changes in that direction. Consequently, the "zero stress" experiments described here in fact were carried out with the 12 MPa stress during transformation. The details of the actual experiments are listed in Table 8.2; the calculated time–temperature–transformation diagram presented in Fig. 6.1 (for alloy 77) confirms that the transformation temperatures are chosen to first cause the growth of allotriomorphic ferrite and then of acicular ferrite.

Table 8.2 Heat Treatment Schedule of the experiments. In each case the austenitisation heat treatment was carried out at 1150 °C for 10 minutes. The stress was applied immediately after the sample reached T_2. t_1 and t_2 are the time periods spent by the sample at the temperatures T_1 and T_2 respectively. The prefix "ST" on the specimen identification indicates transformation under the influence of a large stress which is nevertheless below the yield strength of the austenite and "NST" corresponds to the no stress experiment (see text).

Identification	T_1 °C/ t_1 s	T_2°C/ t_2 s	Stress MPa
NST0	660 / 120	530 / 300	12
NST1	660 / 120	500 / 300	12
NST2	660 / 120	500 / 300	12
ST1	660 / 60	530 / 300	116
ST2	660 / 120	530 / 300	116
ST3	660 / 120	530 / 300	174
ST4	660 / 120	500 / 300	116

8.3 Results and Discussion

8.3.1 The Microstructure

Throughout these experiments, it was our aim to demonstrate a clear effect of stress. The clarity of microstructural changes can be maximised by only partially transforming the austenite to acicular ferrite, and allowing the remainder to decompose by a martensitic transformation during the quench to ambient temperature. Two transformation temperatures were therefore studied, both quite close to the calculated B_S temperature (546 °C). This ensures that the amount of acicular ferrite that forms is limited, because that amount increases from zero at B_S to a quantity determined by the T_0' curve of the phase diagram as the temperature is reduced below B_S.

Specimens NST0 and NST1 which are from the zero stress experiments showed similar microstructures, NST1 having a larger quantity of acicular ferrite, as expected from its lower T_2 transformation temperature. The microstructure of NST1 is illustrated in Fig. 8.4a. It reveals the thin continuous layer of allotriomorphic ferrite which grew and covered the austenite grain boundaries at T_1; in addition, there is the intragranularly nucleated acicular ferrite which formed at T_2. The acicular ferrite microstructure is as expected in normal welds, with plates pointing along many directions. Sample NST2 represents a repeat of the treatment given to

NST1. It was used for confirmation of the experimental method during the initial work when the experimental scheme was being designed. Its microstructure was found to be identical to that of NST1.

Fig. 8.4b–e are longitudinal sections of samples ST1–4; the direction of the applied stress is indicated on each of the micrographs. All of these experiments reveal a dramatic change in the microstructure when the acicular ferrite forms under the influence of the applied stress. The number of crystallographic variants per grain is reduced drastically and there is a strong tendency for the plates that do grow to align within any given prior austenite grain.

These results are easily understood once it is accepted that acicular ferrite growth leads to an invariant–plane strain (IPS) change in the transformed region (Strangwood and Bhadeshia, 1987a). A uniaxial compressive stress will oppose the dilatational component of this IPS (Patel and Cohen, 1953), but the shear component will interact favourably. Since the shear component is much larger than the dilatational strain, the overall interaction of the shape change with the strain can be favourable. The compressive stress resolves to give maximum shear stresses on planes which lie at 45 ° to the stress axis. Given that the samples used are polycrystalline, it is only by fortuitous circumstance that the crystallographic orientation of a particular austenite grain permits acicular ferrite to form on these planes of maximum shear stress. Nevertheless, acicular ferrite variants which happen to be most parallel to those planes are expected to be favoured by the stress, and will form preferentially. This is what leads to the development of the aligned microstructure illustrated in Fig. 8.4b–e. It is noteworthy that there are in principle up to 24 different crystallographic variants of acicular ferrite possible within each austenite grain. Of these, it is likely that one or more may lie fairly close to the 45 ° orientation. Indeed, it is easy to see from Fig. 8.4b–e that the acicular ferrite plates that grow tend to lie fairly close to the optimum orientation.†

One further observation is relevant; Fig. 8.5 shows higher magnification images of samples NST1 and ST2. It is evident that acicular ferrite tends not to form in the regions of austenite adjacent to the allotriomorphs of ferrite. This is exactly as predicted by theory, that the partitioning of carbon during the growth of the allotriomorphs enriches the adjacent austenite which consequently becomes more stable to transformation when compared with unaffected austenite.

8.3.2 The Transformation Strains

Given the nature of the invariant–plane strain shape change (Fig. 8.1), the transformation strain can only be isotropic if the effects of many different randomly orientation crystallographic

---

† In fact, although the plane of maximum shear stress is at 45 degrees to the stress axis, the most favoured plane is calculated by also taking into account the dilatational component of the IPS. However, since that component is rather small, 45 degrees is a good approximation for the optimum orientation.

variants are averaged. The shear components would then tend to cancel out, and the dilatational component of the shape change would appear to be like a uniform volume expansion instead of a uniaxial strain normal to the habit plane. This kind of behaviour might be expected in a random, fine grained polycrystalline sample of austenite.

However, if the sample is crystallographically textured, or if the variants do not form at random, then the transformation strain cannot be isotropic, and the net strain along any particular direction may contain contributions from both the shear and dilatational terms. It then becomes possible to detect transformation plasticity as well as volume change.

The strain measurements are presented in Fig. 8.6. In the absence of stress or crystallographic texture, both the longitudinal and radial strains are expected to be positive since only the volume change is detected. As is seen from 8.6, this is true during the initial stages of transformation (NST1, NST2), but not in the later stages, probably because of the small stress of 12 MPa used for these samples, but possibly also because of crystallographic texture in the samples.

For all other samples (ST1–4), the longitudinal strain is always negative as the specimen transforms in a manner which tends to accommodate the applied compressive stress, whereas the radial strains are all positive. The experiments provide impressive proof of the shape deformation accompanying the displacive growth of acicular ferrite. It would be impossible to explain both the magnitude and signs of the strains detected on the basis of a reconstructive transformation mechanism.

More detailed interpretations of these data require methods of accurately deconvoluting the measured strains into the volume change and transformation plasticity terms. That problem is currently under investigation.

8.4 Conclusions

It is found that the acicular ferrite microstructure responds to stresses below the yield strength of the austenite. The response is in fact quite remarkable, the stress favouring the development of specific crystallographic variants. This leads to a destruction of the conventional acicular ferrite structure in which the plates emanating from inclusions point in many different directions. Apart from these microstructural changes, the measured strains during transformation under the influence of stress give conclusive proof for the shape deformation and displacive mechanism of acicular ferrite growth.

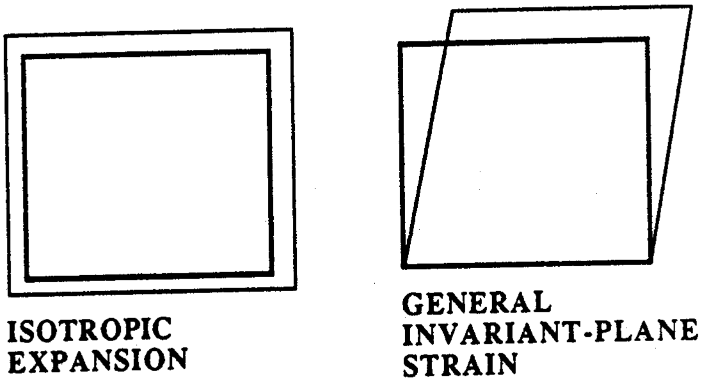

Fig. 8.1 Schematic explanation of shape changes due to isotropic volume change and a general invariant plane strain. The invariant plane strain in austenite to displacive ferrite products involve shear and a dilatational component (Bhadeshia *et al.*, 1991).

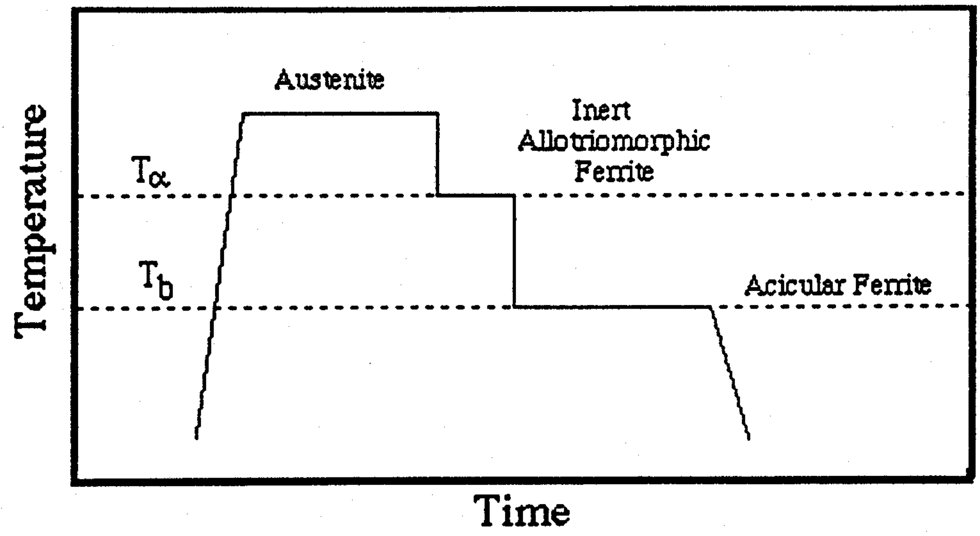

Fig. 8.2 Schematic illustration of two stage heat treatment schedule for generating acicular ferrite microstructure.

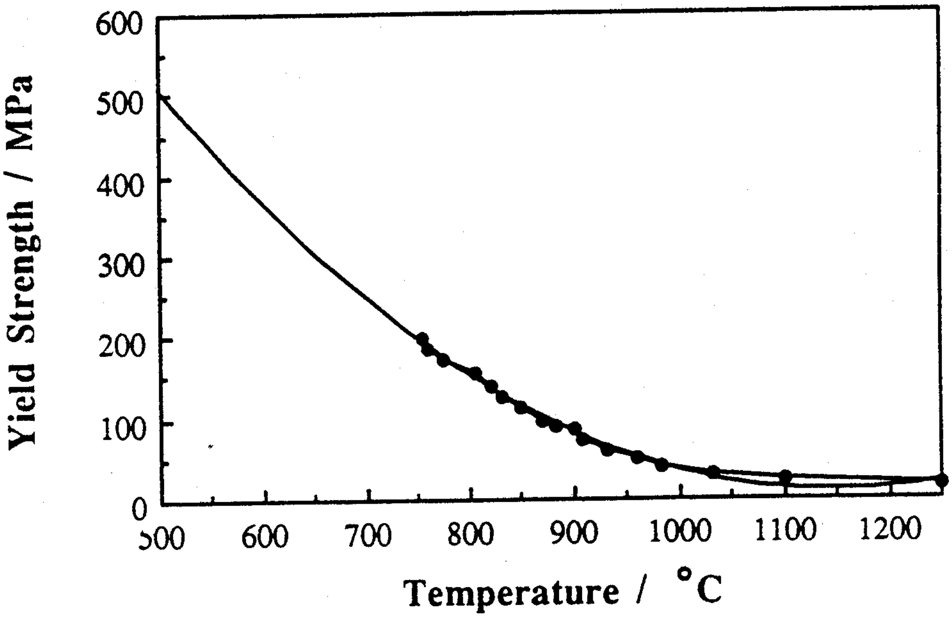

Fig. 8.3 Variation of yield strength as a function of temperature. The graph also contains two extrapolation curves in estimating the yield strength of austenite at around 500 - 550 °C. The experimental data points are for a Nb-V microalloyed steel (Weiss *et al.*, 1981).

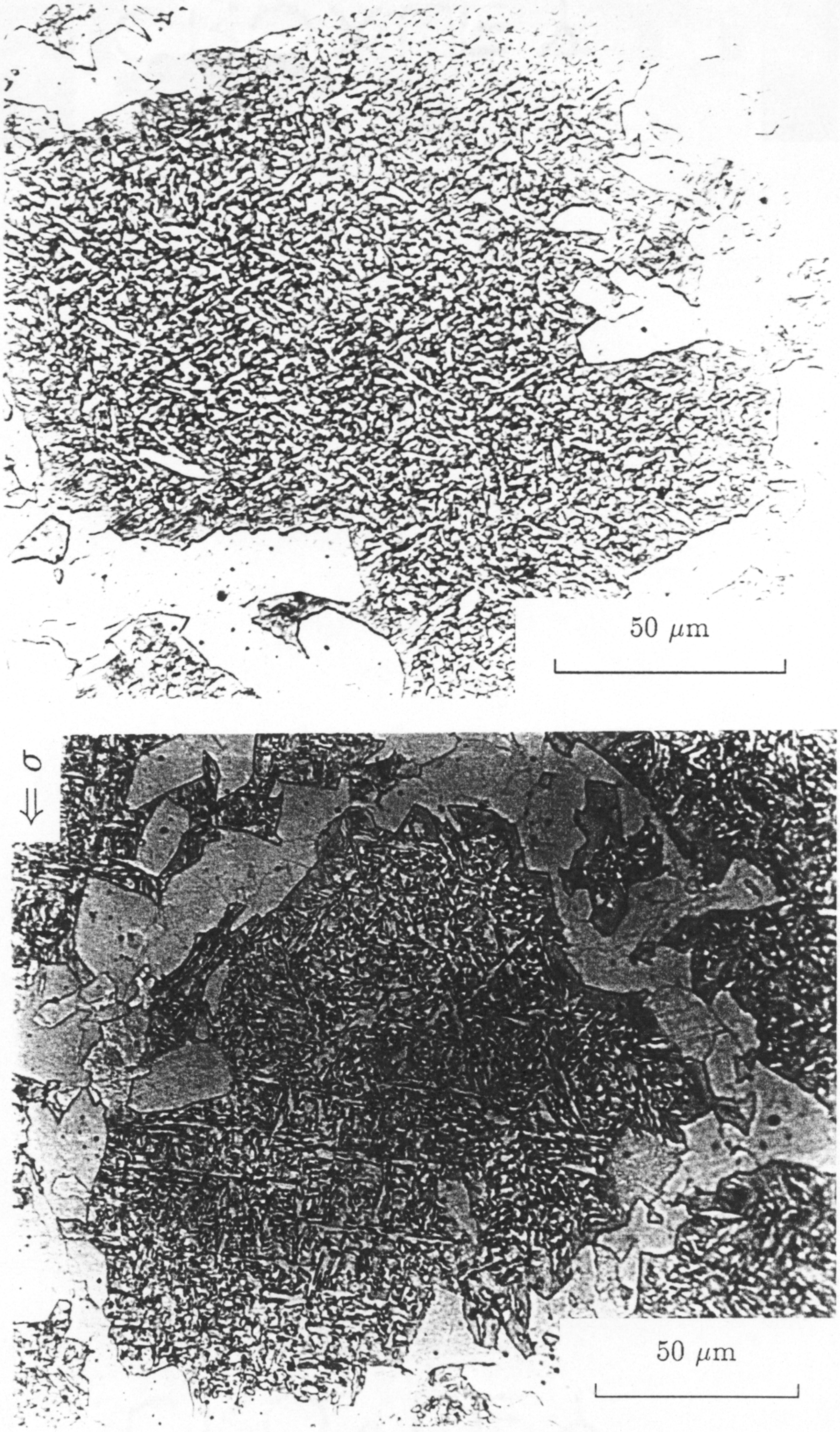

Fig. 8.4 Microstructure observed in no stress experiments and stressed samples. For each case there are two microstructures corresponding to the transverse and longitudinal directions. The arrows in the longitudinal section micrographs indicate the stress direction. The details and the heat treatments are given in the Table 8.2; (a) NST1,

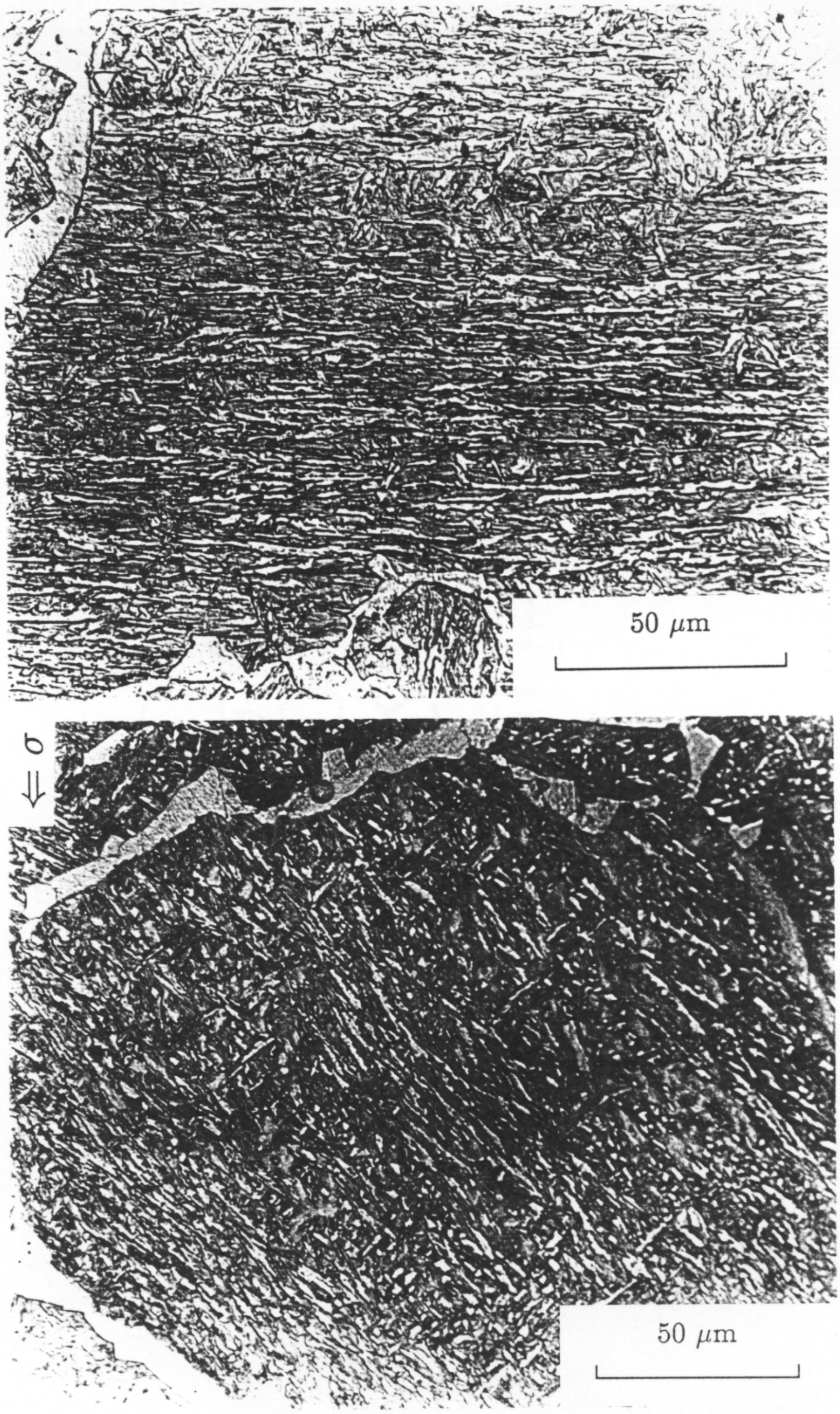

Fig. 8.4 continued...

(b) ST1,

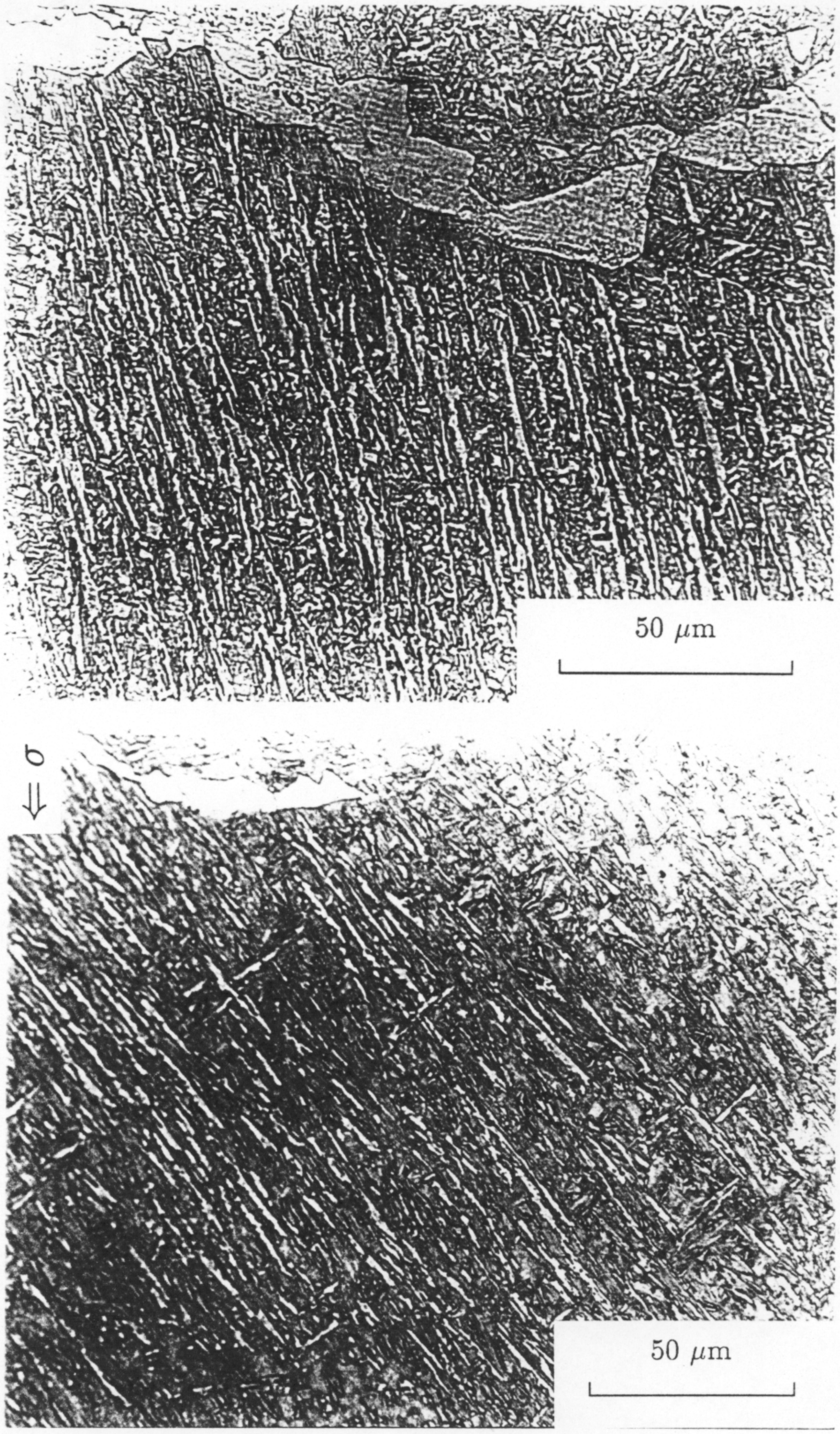

Fig. 8.4 continued...

(c) ST2,

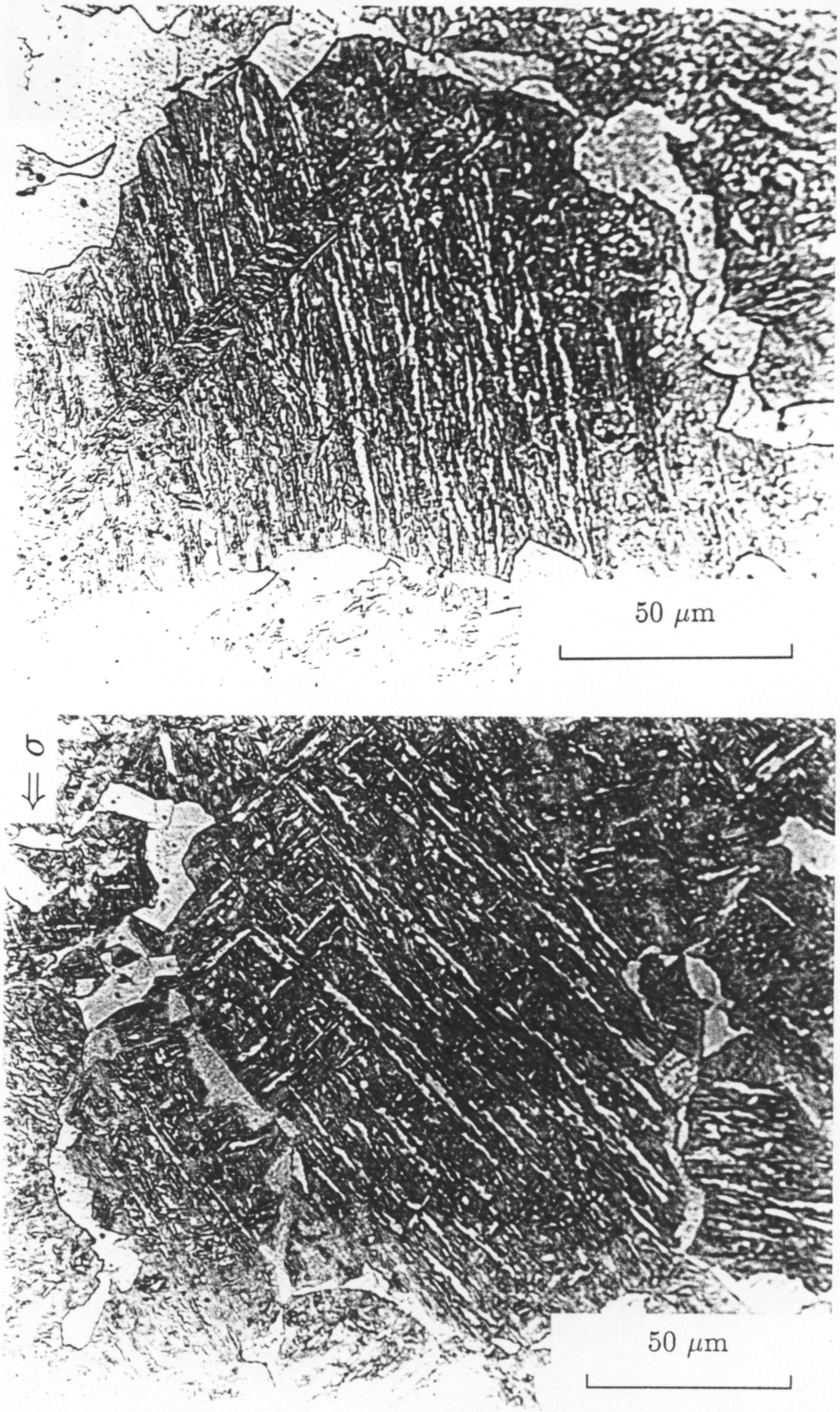

Fig. 8.4 continued...

(d) ST3,

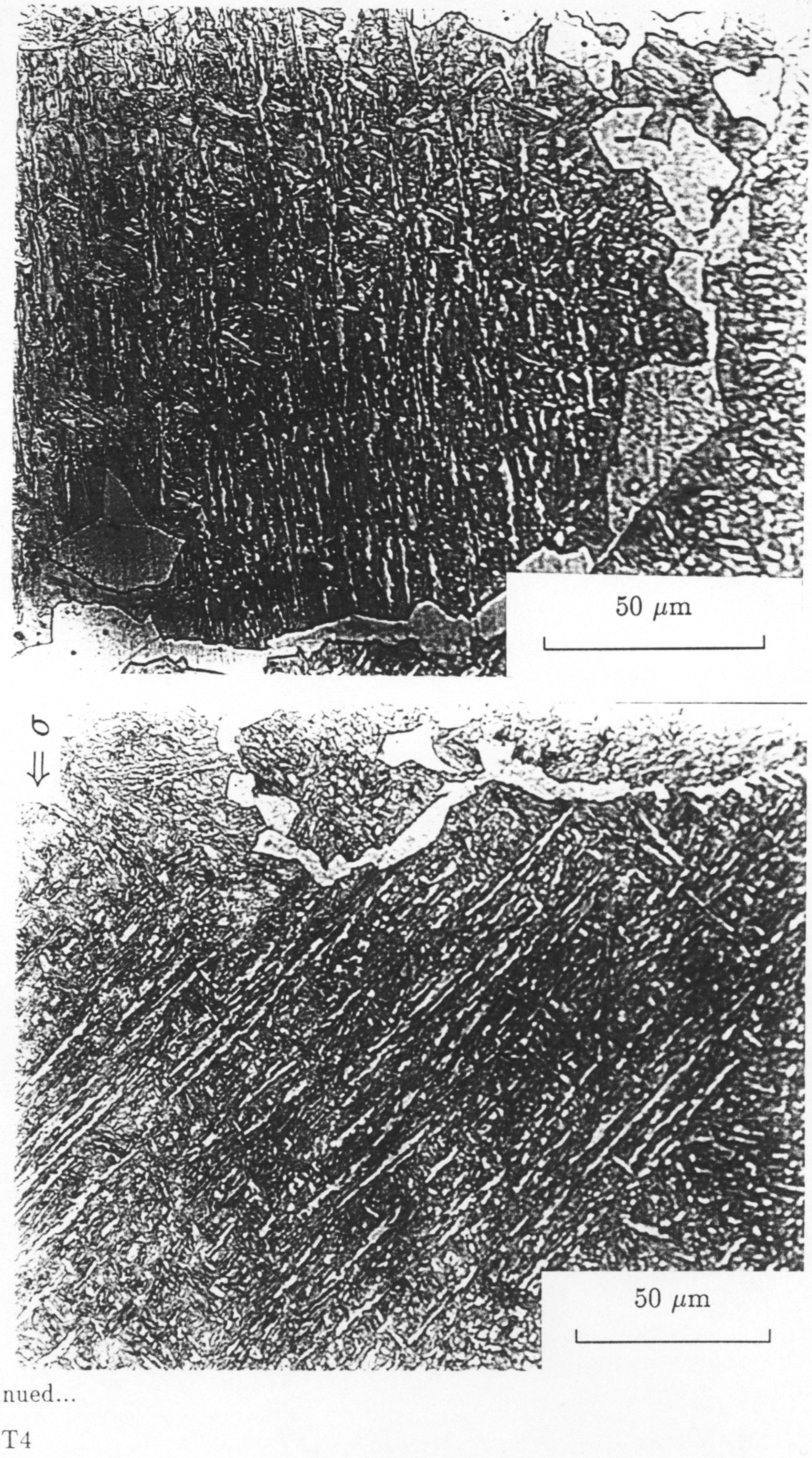

Fig. 8.4 continued...

(e) ST4

Fig. 8.5 Microstructure of (a) NST1 and (b) ST2 at higher magnification. The micrograph illustrates the transformation free zone around the allotriomorphic ferrite interface.

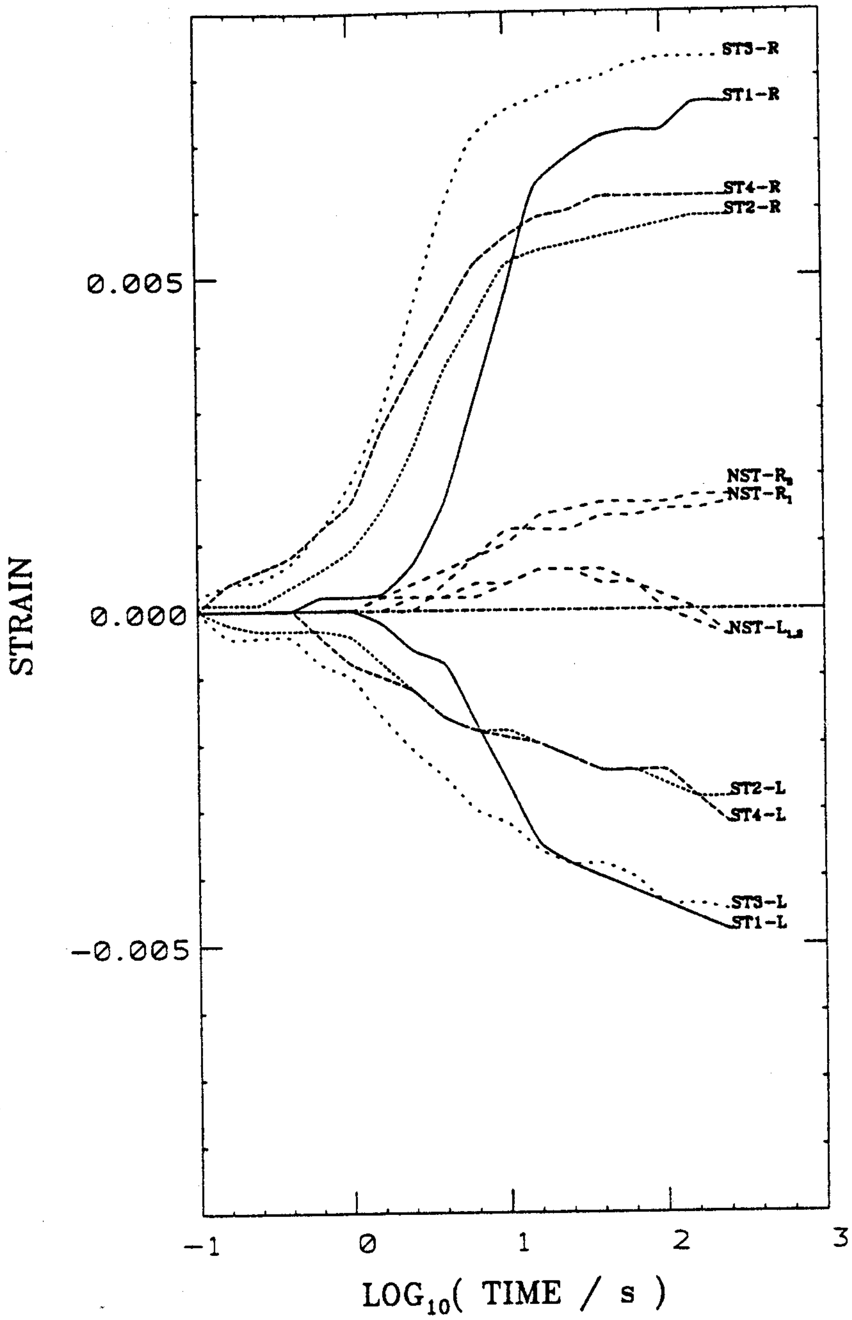

Fig. 8.6 Plot of transformation strain in the longitudinal and transverse directions as a function of time for all the experiments listed in Table 8.4. Here the letter 'R' represents the diametaral strain $\frac{\Delta R}{R}$ and 'L' represents longitudinal strain $\frac{\Delta L}{L}$.

Chapter 9

Effect of Cobalt on Fe-Cr-Mo-C Weld Deposits

9.1 Introduction

The present research project was initiated to explain the transition from acicular ferrite to bainite in Fe-Cr-C welds. The introduction of thin allotriomorphic ferrite layers on the austenite grain boundaries has been demonstrated to optimise the opportunity for the formation of acicular ferrite (chapters 4,5& 6). The problem with the 2.25Cr-1Mo (wt.%) welds widely used in industry is that it is important to cool them at rates which allow the formation of allotriomorphic ferrite. The aim of the work presented in this chapter was to introduce the formation of allotriomorphic ferrite by appropriate alloying. Suitable alloying additions are Al, Si and Co, all of which reduce the stability of austenite (Aaronson and Domain, 1966). Silicon is not, however, desirable from the toughness point of view, whereas a large concentration of aluminium cannot be transferred across the welding arc because of its strong oxidising potential. Consequently, cobalt was chosen to reduce the hardenability.

9.2 Design of Welds

The weld microstructure model developed by Bhadeshia *et al.,* (1985), as introduced in chapter 1, is based on the paraequilibrium thermodynamics for the calculation of phase diagrams and TTT diagrams. The alloying additions bring about the change in thermodynamics by:

1. The free energy for transformation from austenite to ferrite in alloyed iron, namely $\Delta G_{Fe}^{\gamma\to\alpha}$ is modified. The alloying additions modify the magnetic and non magnetic component of free energy *vs* temperature curves, a concept due to Zener (1955).
2. The carbon activity in austenite is influenced by substitutional alloying additions and this is accounted for, by altering the pairwise carbon-carbon interaction energy (ω_γ) in austenite.

Aaronson *et al.,* (1966b) represented the free energy change for the austenite Fe-X alloy transforming to ferrite with the same composition by the following equation (Bhadeshia, 1982a):

$$\Delta G_{Fe}^{\gamma\to\alpha} = \Delta^\circ G_{Fe}^{\gamma\to\alpha}\{T - 100\sum_{i=1}^{n} Y_i(\Delta T_{mag})_i\} + 141\sum_{i=1}^{n} Y_i(\Delta T_{mag} - \Delta T_{non.mag})_i \qquad (9.1)$$

Here the curly brackets represent a functional relation. ΔT_{mag}, $\Delta T_{non.mag}$ are the displacements in the temperature at which the respective free energy change accompanying austenite to ferrite in pure iron are calculated, in order to account for the changes in magnetic and non magnetic free energy components. The values of $(\Delta T_{mag})_i$ and $(\Delta T_{non.mag})_i$ tabulated in the work of

Table 9.1. The magnitudes of $\Delta T_{non.mag}$ & ΔT_{mag} for various substitutional alloying additions.

Element	$\Delta T_{non.mag}$ °C per at. % X	ΔT_{mag} °C per at. % X
Si	0.0	-3.0
Mn	-39.5	-37.5
Ni	-18.0	-6.0
Mo	-17.0	-26.0
Cr	-18.0	-19.0
V	-32.0	-44.0
Co	16.0	19.5
Cu	-11.5	-4.5
Al	+15.0	8.0
W	-17.0	-26.0

Zener (1955) and Aaronson *et al.,* (1966b) were taken for the calculations (Table 9.1) and Y_i is the mole fraction of ith substitutional alloying addition disregarding the presence of carbon.

The alloying elements affect the carbon-carbon (ω_γ) interaction energies in the austenite, which modifies the activity of carbon. The variations of ω_γ with atom fractions of substitutional alloying elements can be calculated as presented in the references (Aaronson *et al.,* , 1966b, Bhadeshia, 1981c) and were represented as a polynomial equation. The variation of ω_γ with atom fraction of cobalt derived by Bhadeshia (1981c) was adopted to present calculations (Figure 9.1).

The model after the above modification predicted a shift of reconstructive transformation to allotriomorphic ferrite to a shorter times, as seen in TTT diagrams. The TTT diagrams will be presented later, together with measured data. The cobalt containing alloy (AM1, 4.3 wt.% Co) was compared with one without cobalt (A1).

9.3 Experimentation

The procedure for all-weld metal test pieces has been explained in chapter 2. The welding variables and detailed weld compositions (A1 - 0.0 wt.% Co and AM1 - 4.3 wt.% Co) are presented in Tables 2.1 & 2.2. The columnar austenite grain sizes ($\bar{L}_{tn}$) were measured and were found to be 105 ± 21 and 95 ± 15 μm. This difference was considered to be insignificant. The austenitisation conditions were 1150 °Cand 10 minutes during dilatometry. The prior austenite grain size after this austenitisation treatment was found to be 80-90 μm. The isothermal treatments were carried out at 820, 800, 775, 750, 725, 700, 675, 650, 625, 600, 575, and 550 °C for time periods up to one hour.

9.4 Results and Discussion

Figures 9.2 & 9.3 show the absence of allotriomorphic ferrite and acicular ferrite in both alloys with no discernible difference in microstructure. Only bainitic sheaves could be found. The transformation temperatures for the two alloys are presented in Table 9.2. They are, as expected, higher for alloy AM1 compared with A1. Hence, with a similar cooling rate, AM1 might have a tendency to develop into higher volume fractions of martensite. The optical microstructure showed no obvious difference between A1 and AM1. The cobalt thus proved to be ineffective.

Table 9.2. Calculated transformation temperatures for alloy A1 and AM1.

ALLOY	W_s range °C	B_s °C	M_s °C
A1	660-680	555	440
AM1	660-780	644	536

The hardness values in the weld metal regions were found to be 283 ± 6 HV in alloy A1 and 289 ± 6 HV in alloy AM1. Isothermal dilatometric data were compared with the calculated TTT diagrams (Fig. 9.4). The transformation times for a constant volume fraction of transformation (observed from the relative length change) were measured to construct the TTT diagrams. There is negligible difference between the experimental transformation start times of alloy A1 and AM1 for all transformation temperatures above 660 °C. This is consistent with the absence of allotriomorphic ferrite reaction in both welds. The comparison of the experimental and calculated transformation start times for alloy A1 show a good correlation, but, a gross underestimation of the incubation time for cobalt (alloy AM1) containing alloy is evident from Fig. 9.4. The thermodynamic model predicts a order of magnitude smaller incubation time.

If the cobalt addition causes a change in the free energy of transformation, there should be corresponding change in transformation temperature. The bainitic transformation temperatures were measured during continuous cooling at 30 °Cs^{-1} (Fig. 9.5). The results agree well with the calculations (Bhadeshia, 1982a): A1; Predicted B_s=555, Measured B_s=580 & AM1; Predicted B_s= 644 , Measured B_s= 630 °C. This proves that the cobalt certainly has a thermodynamic effect on bainite.

As a further check, the martensitic transformation temperatures data given by Raghavan and Thomas (1971) were examined with the present thermodynamic model. They studied Fe-0.3C-4Cr (wt.%) with differing cobalt concentration. Their measured M_s temperatures are 348, 383 and 390 °Cfor cobalt concentrations 0, 1.4, and 5.3 wt.% respectively. Corresponding calculated martensitic transformation temperatures were 325, 383 and 435 °C. The trends are therefore correctly predicted although the discrepancy is large at high concentration. The excellent correlation in the calculated bainitic transformation temperature with the experimental points, and good correlation of TTT data points with experimental points of alloy A1 indicates that the free energy modification was indeed taken into account in the thermodynamic model. Further experiments (Fig. 9.6 and 9.7) verified that cobalt did not accelerate the transformation to allotriomorphic ferrite in Fe-Cr-Mo-C steels.

9.5 Summary

The driving force for transformation is increased on addition of cobalt as indicated by the changes in bainite transformation temperatures, although for reasons which are not clear, it was found not to influence the formation of allotriomorphic ferrite in the alloys studied.

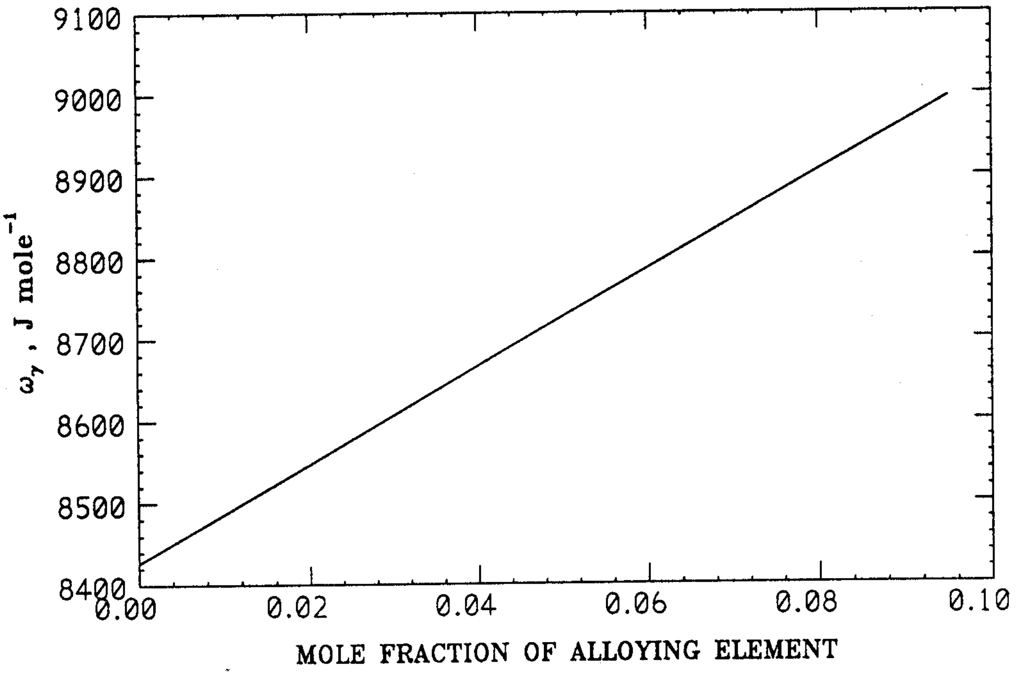

Fig. 9.1 Variation of carbon-carbon (ω_γ) interaction energy in austenite as a function of cobalt concentration (after Bhadeshia, 1981c).

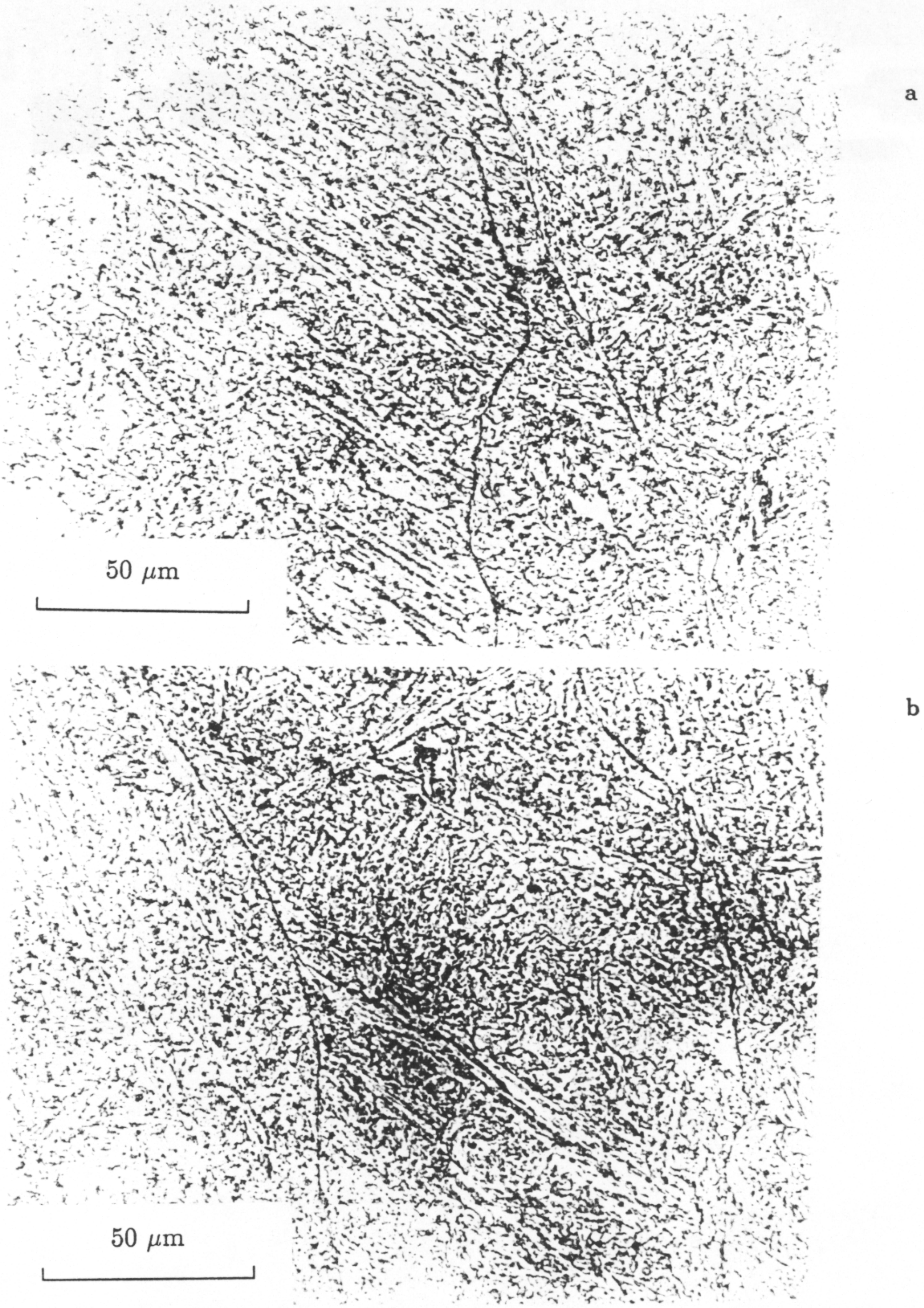

Fig. 9.2 Microstructure observed in welds A1 and AM1. The absence of grain boundary nucleated allotriomorphic ferrite and acicular ferrite is noticeable. (a) A1, (b) AM1.

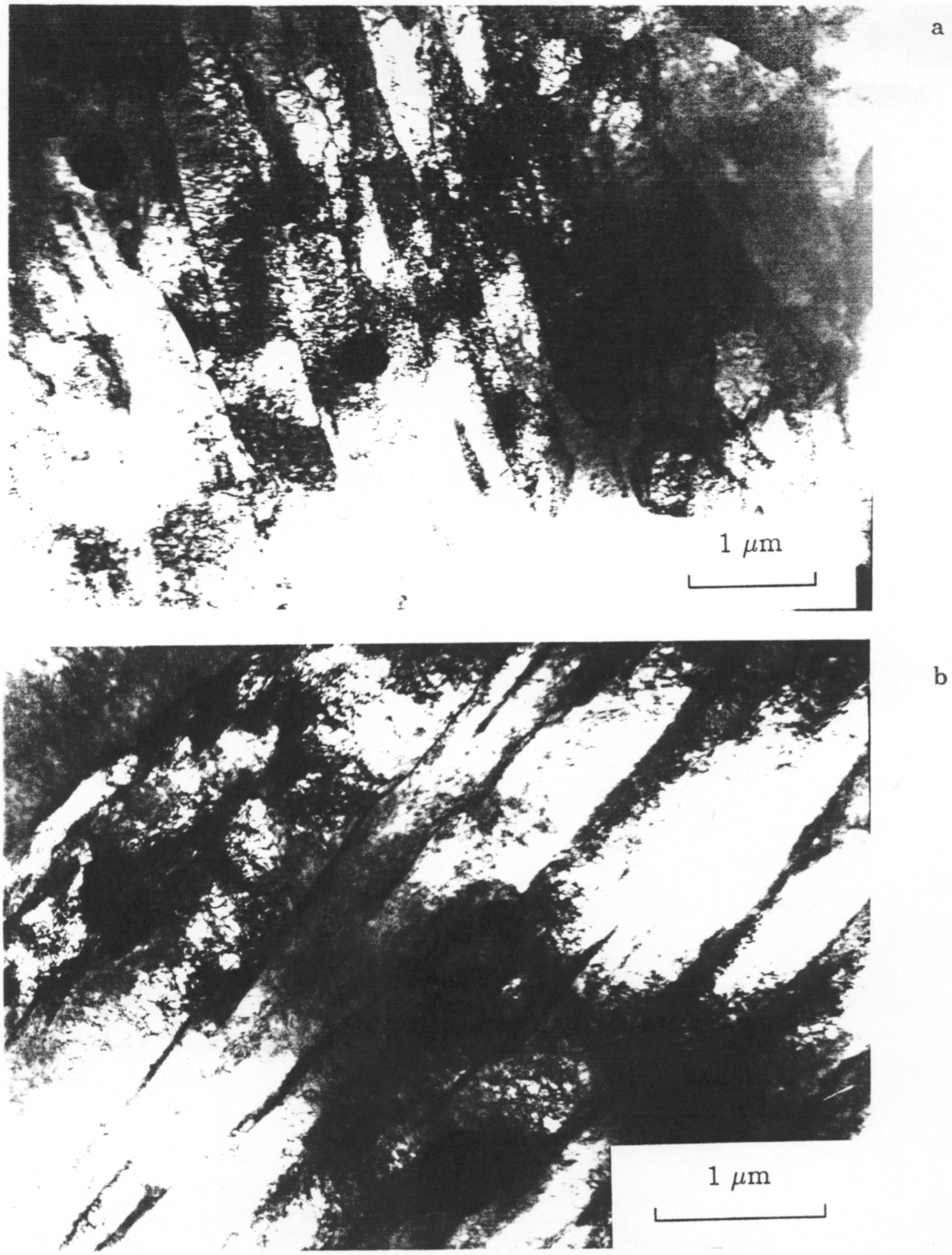

Fig. 9.3 Comparison of the primary weld metal microstructures of alloy A1 and alloy AM1. (a) A1, (b) AM1.

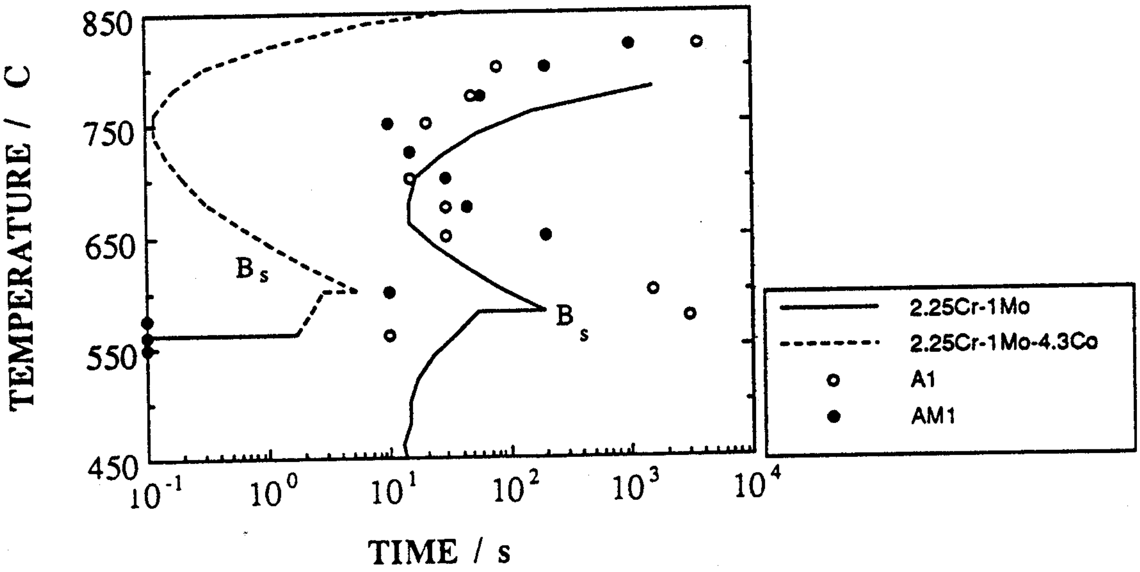

Fig. 9.4 Comparison of calculated TTT curves for alloys A1 and AM1. The points were measured using dilatometry. The open circles are for weld A1 and filled circles for weld AM1. Note the similarity of the TTT curves at transformation temperatures above 660 °C for both alloys.

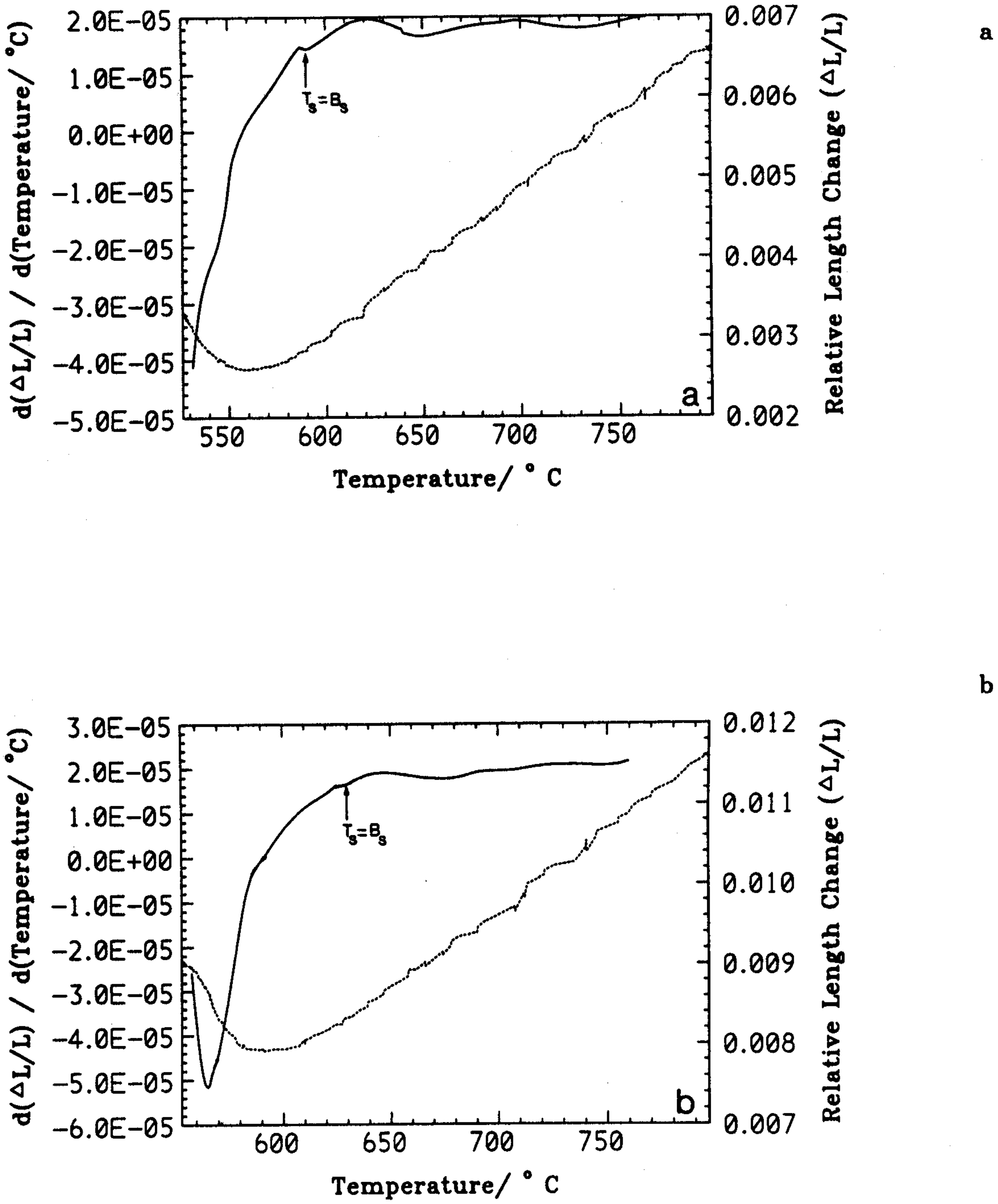

Fig. 9.5 Relative length change as a function of temperature for alloys A1 and AM1. The solid line is $d(\frac{\Delta L}{L})/d(T)$ vs T. The Dotted lines correspond to the original $\frac{\Delta L}{L}$ *vs* T. The onset of bainitic transformation is marked with the arrow (a) A1, (b) AM1.

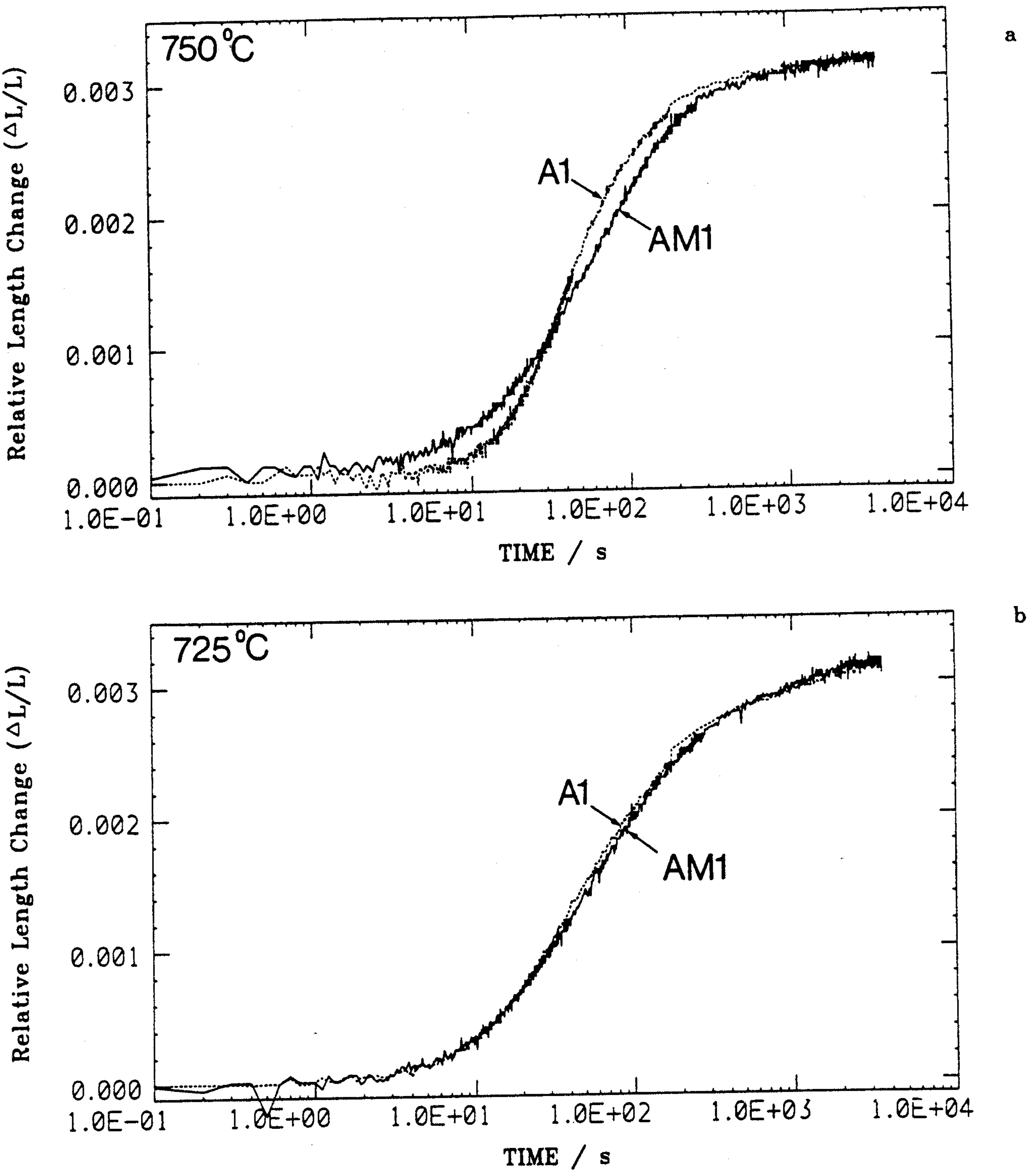

Fig. 9.6 Allotriomorphic ferrite kinetics for the alloys A1 and AM1, for two transformation temperatures, (a) 750 °C, (b) 725 °C.

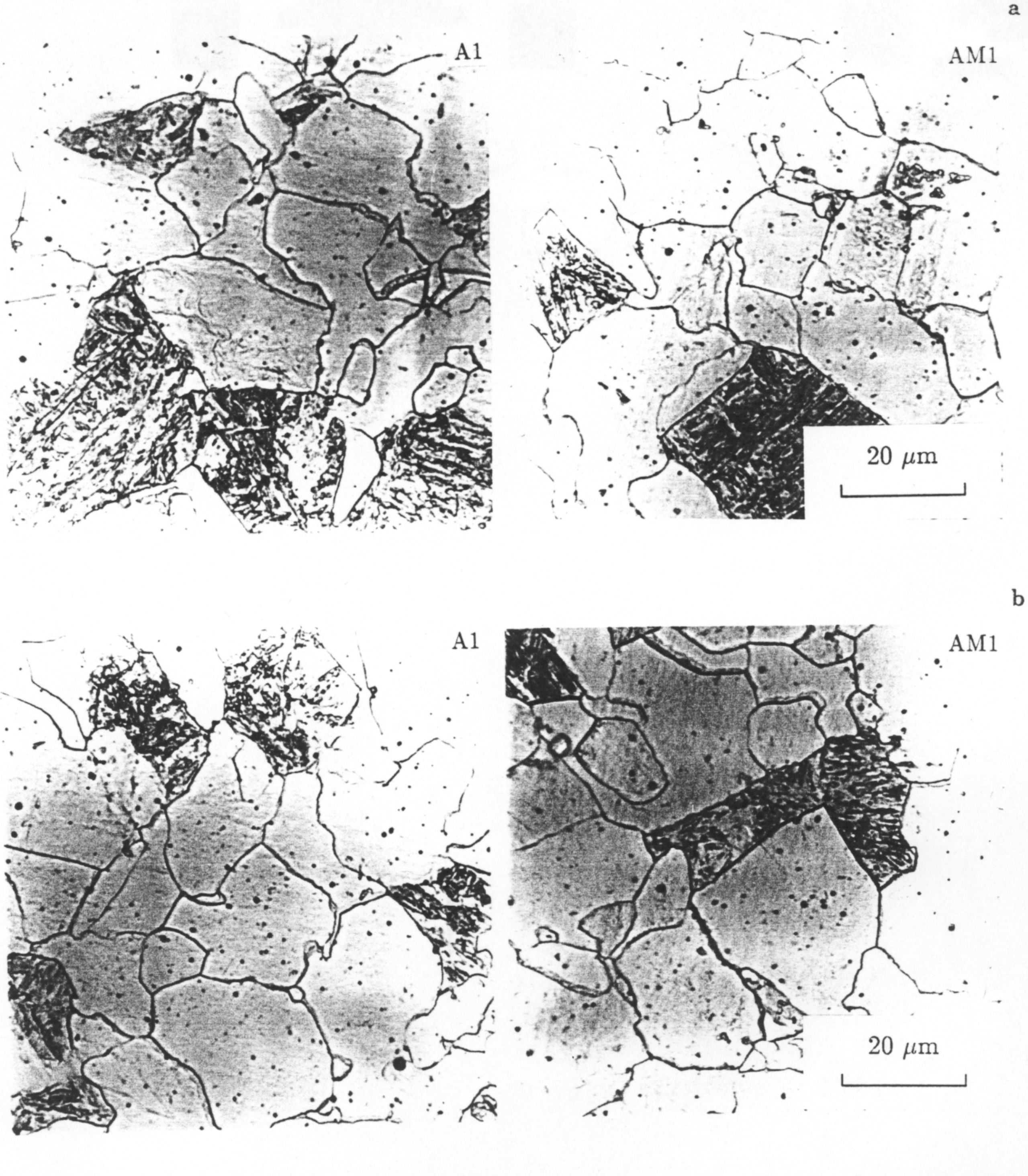

Fig. 9.7 Corresponding microstructures for the transformation conditions of Fig. 9.6. (a) 750 °C, (b) 725 °C.

Chapter 10

Future Research

The present research has illustrated a simple way of rendering the austenite grain boundaries ineffective in nucleating bainite by decorating them with thin layers of allotriomorphic ferrite. This allows the inclusions within the residual austenite an opportunity to nucleate acicular ferrite. The inertness of allotriomorphic ferrite is explained on the basis of crystallography and solute partitioning.

The columnar austenite grain parameters in weld control the subsequent microstructure development. In single run welds the grain size is controlled by weld composition and the base plate texture. In multirun weld it is further complicated by the solidification segregation patterns created by previous weld passes. Hence, a fundamental model has to be developed using solidification theories, crystallography and thermodynamics of the transformation to δ ferrite and austenite from liquid steel.

The allotriomorphic ferrite model now needs to be improved to predict crucial parameters such as the extent to which the austenite grain boundaries are covered by layers of allotriomorphic ferrite and the transition from local to para-equilibrium mode of growth. This will require a detailed and extensive re-examination of allotriomorphic ferrite kinetics in both the nucleation and growth stages. The kinetic model should include isothermal and anisothermal transformation conditions.

The second and most promising need is to model the kinetics of transformations under the influence of stress. Further stress/crystallography data ha to be collected, data of importance in variant selection and hence in calculating mechanical driving force. The measured transformation strains with stress will form a basis for calculating the residual stress in welds as they cool. Alloy modification or a change in weld cooling rate conditions can be considered to modify the residual stress development. The application of this research lies in the design of welded joint of high strength steels, which experience thermal, applied and shrinkage stresses.

REFERENCES

Aaronson, H. I. (1962a)

On the partition of chromium between austenite and proeutectoid ferrite, *Trans. Met. Soc - AIME*, Vol. 224, pp. 870-871.

Aaronson, H. I. (1962b)

Decomposition of austenite by diffusional processes, Interscience, New York, USA.

Aaronson, H. I. and Domain, H. A. (1966)

Partition of alloying elements between austenite and proeutectoid ferrite or bainite, *Trans. Met. Soc - AIME*, Vol. 236, pp. 781-796.

Aaronson, H. I., Domain, H. A. and Pound, G. M. (1966a)

Thermodynamics of the austenite → proeutectoid ferrite Transformation I, Fe-C alloys, *Trans. Met. Soc - AIME*, Vol. 236, pp. 753-767

Aaronson, H. I., Domain, H. A. and Pound, G. M. (1966b)

Thermodynamics of the austenite → proeutectoid ferrite Transformation II, Fe-C-X alloys, *Trans. Met. Soc - AIME*, Vol. 236, pp. 768-781

Aaronson, H. I. and Russell, K .C. (1981)

Nucleation – mostly homogeneous and in solids, Solid → Solid phase transformations, TMS-AIME, Warrendale, Pennsylvania, USA, pp. 371-397.

Abson, D, J. (1978).

The role of inclusions in controlling weld metal microstructure in C - Mn steels, The Welding Institute, Cambridge, England.

Abson, D. J. (1982)

Influence of Manganese and preheat on the microstructure and mechanical properties of vertical up MMAW welds in 38 mm thick C - Mn - Nb steel plate - Part I, *Welding Institute Report*, 194/1982, October

Abson, D. J. (1986)

Influence of Manganese and preheat on the microstructure and mechanical properties of vertical up MMAW welds in 38 mm thick C - Mn - Nb steel plate - Final report, *Welding Institute Report*, 309/1986, August.

Abson, D, J., Dolby, R. E., and Hart, P. H. M. (1978).

The role of non metallic inclusions in ferrite nucleation in carbon steel weld metals. Proc. Intl. Conf. on *Trends in Steels and consumables for welding*, London, England, The Welding Institute. p. 75

Abson, D. J., and Pargeter, R. J. (1986)

Factors affecting as deposited strength, microstructure and toughness of manual metal arc welds suitable to C-Mn steel fabrications, *Int. Met. Reviews*, Vol. 31, No. 4, p. 141

Ali, A., and Bhadeshia, H. K. D. H. (1989)

Growth rate data on bainite in alloy steels, *Materials Science and Technology*, Vol. 5, April, pp. 398-402.

Alok Singh., Ramakrishna, D. and Gupta, S. P. (1988)

Kinetics of Ferrite transformation in HSLA steels, *Z. Metallkde*, pp. 180-188.

Andrews, K. W. (1956)

The calculation of transformation temperatures and austenite-ferrite equilibria in steels, *J.I.S.I*, pp. 414-427.

Ashby, M. F. (1987)

Technology of the 1990s: advanced materials and predictive design, *Phil. Trans. R. Soc. London.*, A 322, pp. 393-407.

Ashby, M. F., and Easterling, K. E. (1982)

A first report on diagram for grain growth in welds, *Acta Metallurgica*, Vol. 30, pp. 169-198.

Atkinson, C. (1982)

The growth kinetics of ledges during phase transformations: multistep interactions, *Proc. R. Soc. (London)*, A384, p. 107.

AWS - Welding Handbook, Vol. 1

Macmillan press, 1981.

AWS - Welding Handbook, Vol. 4

Macmillan press, 1972.

Aziz, M. J. (1982)

Model for solute redistribution during rapid solidification, *J. Appl. Physics*, Vol. 53, pp. 1158-1168.

Bain, E. C. and Paxton, H. W. (1966)

Alloying elements in steel, 2nd Edition, American Society for Metals, Metals Park, Ohio, USA.

Babu, S. S. and Bhadeshia, H. K. D. H. (1990)

Transition from bainite to acicular ferrite in reheated Fe-Cr-C weld deposits, *Materials Science and Technology*, Vol.6, October, pp. 1005-1020.

Babu, S. S. and Bhadeshia, H. K. D. H. (1991)

A direct study of allotriomorphic ferrite crystallography, *Materials Science and Engineering*, Vol. A142, pp. 209-219.

Babu, S. S. and Bhadeshia, H. K. D. H. (1991)

Mechanism of the transition from bainite to acicular ferrite, accepted for publication in the *Materials Transactions*, Japan Institute of Metals.

Babu, S. S., Bhadeshia H. K. D. H., and Svensson, L.-E. (1991)

Crystallographic texture and the austenite grain structure of low alloy steel weld deposits, *Journal of Materials Science Letters*, Vol. 10, pp. 142-144.

Baker, R. G. (1975)

Weldability and its implications for materials requirements, Proc. *Rosenhain Centenary Conference*, a joint symposium held 22-24, Sep, 1975 of the metals society, National Physical Lab and Royal Society, London., pp. 207-222.

Baker, R. G., and Nutting, M. A. (1959)

The tempering of $2\frac{1}{4}$%Cr-1%Mo steel after quenching and normalising, *Journal of the Iron and Steel Institute*, p. 266.

Ban-ya, S., Elliot, J. F. and Chipman, J. (1969)

Activity of carbon in Fe-C alloys at 1150 °C, *Transactions of the Met. Soc. AIME*, Vol. 245, pp. 1199-1206.

Ban-ya, S., Elliot, J. F. and Chipman, J. (1970)

Thermodynamics of austenite in Fe-C alloys, *Metallurgical Transactions*, Vol .1, pp. 1313-1320.

Bhadeshia, H. K. D. H. (1980a)

Thermodynamics of steels: carbon - carbon interaction energy, *Metals Science*, 1980, Vol. 14, pp.230-232.

Bhadeshia, H. K. D. H. (1980b)

The Lower Bainite Transformation and the Significance of Carbide Precipitation, *Acta Metallurgica*, Vol. 28, pp. 1103-1114

Bhadeshia, H. K. D. H. (1981a)

A rationalisation of shear transformations in steels, *Acta Metallurgica*, Vol. 29, pp. 1117 - 1130.

Bhadeshia, H. K. D. H. (1981b)
Diffusion of carbon in austenite, *Metal Science*, Vol. 15, pp. 477-479.

Bhadeshia, H. K. D. H. (1981c)
Thermodynamic extrapolation and martensite-start temperature of substitutionally alloyed steels, *Metal Science*, Vol. 15, pp. 178-180.

Bhadeshia, H. K. D. H. (1981d)
Driving force for martensitic transformation in steels, *Metal Science*, Vol. 15, pp. 175-177.

Bhadeshia, H. K. D. H. (1982a)
A Thermodynamic analysis of isothermal transformation diagrams, *Metal Science*, Vol. 16, pp. 159 -165.

Bhadeshia, H. K. D. H. (1982b)
Bainite: Overall transformation kinetics, *Journal de Physique, Colloque C4*, No. 12, Vol. 43, p. C4-435.

Bhadeshia, H. K. D. H. (1982c)
Application of first-order quasichemical theory to transformations in steels, *Metal Science*, Vol. 16, pp. 167 - 169.

Bhadeshia, H. K. D. H. (1982d)
An Aspect of the Nucleation of "Burst" Martensite in Steels, *Journal of Materials Science*, Vol. 17, 1982, pp. 383-386.

Bhadeshia, H. K. D. H. (1985)
Diffusional formation of ferrite in iron and its alloys, *Progress in Materials Science*, Vol. 29, 1985, pp. 321-386.

Bhadeshia, H. K. D. H. (1987a)
Bainite in Steels, Presented at *Phase Transformations '87.*

Bhadeshia, H. K. D. H. (1987b)
4 4 Worked examples in the geometry of crystals, Institute of Metals publication.

Bhadeshia, H. K. D. H. (1989)
Modelling of microstructure in the fusion zone of steel weld deposits, Keynote, Second International Conference on *Trends in Welding Research*, 14 - 18, May, Gatlingburg, Tennessee, U.S.A.

Bhadeshia, H. K. D. H. (1990)
Control of Weld Metal Microstructure and Properties, Proceedings of and International Conference on Metallurgy, Welding and Qualification of Microalloyed (HSLA) Steel Weldments, November pp. 1-35.

Bhadeshia, H. K. D. H., and Christian, J. W. (1990)
Bainite in Steels, *Metallurgical Transactions A*, Vol. 21A, pp. 767-797.

Bhadeshia, H. K. D. H., David, S. A., Vitek, J. M. and Reed, R. W. (1991)
Stress-induced transformation to bainite in a Fe-Cr-Mo-C pressure vessel steel *Materials Science and Technology,*, Vol. 7, pp. 686-698.

Bhadeshia, H. K. D. H., and Edmonds, D. V. (1979)
The Bainite Transformation in a Silicon Steel, *Metallurgical Transactions A*, 1979, Vol 10A, pp. 895-907.

Bhadeshia, H. K. D. H., and Edmonds, D. V. (1980)
The mechanism of bainite formation in steels. *Acta Metallurgica*, Vol. 28, pp. 1265 - 1273.

Bhadeshia, H. K. D. H., Svensson, L. E., and Gretoft, B. (1985a)
A Model for development of microstructure in low-alloy steel (Fe-Mn-Si-C) weld deposits. *Acta Metallurgica*, Vol. 33, pp. 1271 - 1283.

Bhadeshia, H. K. D. H., Svensson, L. E., and Gretoft, B. (1985b)
The influence of alloying elements on the formation of allotriomorphic ferrite in low-alloy steel weld deposits, *Journal of Materials Science Letters*, Vol. 4, pp. 305-308.

Bhadeshia, H. K. D. H., Svensson, L.-E., and Gretoft, B. (1986a)

Prediction of Microstructure of Submerged - arc line pipe welds., International Conference on *welding and performance of pipelines*, 1986, published by the Welding Institute, Abington, U.K.

Bhadeshia, H. K. D. H., Svensson, L. E., and Gretoft, B. (1986b).

The austenite grain structure of low-alloy steel weld deposits, *Journal of Materials Science*, Vol. 21, pp. 3947 - 3951.

Bhadeshia, H. K. D. H., Svensson, L. E., and Gretoft, B. (1986c).

An analysis of the primary microstructure of Cr and Mo containing low alloy steel welds, Proceedings of the 4th Scandinavian Symposium on Materials Science, Norwegian Institute of Technology, Norway.

Bhadeshia, H. K. D. H., Svensson, L. E., and Gretoft, B. (1987)

Theory for allotriomorphic ferrite formation in steel weld deposits, Proc. Intl. Conf.on *Welding Metallurgy of Structural Steels.* AIME, Warrendale, Pennsylvania. Edited by J. Y. Koo, pp. 517 - 530.

39

Bhadeshia, H. K. D. H., and Waugh, A. R. (1982)

Bainite: An atom probe study of the incomplete reaction phenomenon, *Acta Metallurgica*, Vol. 30, pp. 775 - 784.

BISRA (1956)

'Atlas of isothermal Transformation Diagrams of BS En steels', special report no. 56, 2 ed., London, The Iron and Steel Institute.

Bodsworth, C., Davidson, I. M., and Atkinson, D. (1968)

Thermodynamic Properties of Interstitial Solutions of Iron base Alloys, *Transactions of the Met. Soc. of AIME*, Vol. 242, pp. 1135-1143.

Bowles, J. S. and Mackenzie, J. K. (1954)

The Crystallography of Martensitic Transformations I, *Acta Metallurgica,* Vol. 2, pp. 129-137.

Brooksbank, D., and Andrews, K. W. (1972)

Stress fields around inclusions and their relation to mechanical properties, *Journal of the Iron and Steel Institute*, April, p. 248.

Cahn, J. W. (1956)

The kinetics of grain boundary nucleated reactions, *Acta Metallurgica*, Vol. 4, pp. 449-459.

Cahn, J. W. (1980)

Obtaining inferences about relative stability and metastable phase sequences from phase diagrams, *Bulletin of Alloy Phase diagrams*, Vol. 1, No.2, pp. 27-32.

Christensen, N., Davies, V. de L. and Gjermundsen, K. (1965)

Distribution of Temperatures in arc welding, *Brit. Weld. Journal*, pp. 54-75.

Christian, J. W. (1975)

Theory of Transformations in Metals and Alloys, 2nd Edition, Pt. 1, Pergamon Press, Oxford.

Christian, J. W. and Edmonds, D. V. (1984)

The Bainite transformation, Proc. Intl. Conf. on *Phase Transformation in Ferrous alloys*, Ed. Marder, A. and Goldstein, J. I., Metals Park, OHIO.

Clark, R. A. and Thomas, G. (1975)

Design of strong tough Fe/Mo/C steels and the effects of cobalt, *Metallurgical Transactions A*, Vol. 6A., pp. 969-978.

Coates, D. E. (1973a)

Diffusion controlled precipitate growth in ternary system: II, *Metallurgical Transactions*, Vol. 4, p. 1077

Coates, D. E. (1973b)

Precipitate growth kinetics in Fe-C-X alloys, *Metallurgical Transactions*, Vol. 4, p. 395.

Coates, D. E. (1973c)

Diffusional growth limitation and hardenability, *Metallurgical Transactions*, Vol. 4, p. 2313.

Cochrane, R. C. and Kirkwood, P. R. (1978)

The Effect of oxygen on weld metal microstructure, Proc. Intl. Conf. on *Trends in Steels and consumables for welding*, London, England, The Welding Institute. p. 103.

Court, S. A. and Pollard, G. (1987)

The effects of Mn and Si on the microstructure and properties of SMA steel weld metal, Proceedings of the International Symposium on *Welding Metallurgy of Structural Steels* held in Denver, Colorado, February 22 - 26, pp. 335-347.

Crank. J. (1975)

'The Mathematics of Diffusion', 2nd edition, Clarendon Press, Oxford.

Crosky, A., McDougall, P. G. and Bowles, J. S (1980)

The crystallography of the precipitation of α rods from β Cu-Zn alloys, *Acta Metallurgica*, Vol. 28, pp. 1495-1504.

Cullity, B. D. (1978)

'Elements of X-ray diffraction', p. 356, Addison-Wesley, Reading, 1978, N. Y.

Dallam, C. B. and Olson, D. L. (1989)

Stress and grain size effects on weld metal ferrite formation, *Welding Journal*, May, pp. 198s-205s.

David, S. A. and Vitek, J. M. (1989)

Correlation between solidification parameters and weld microstructures, *International materials reviews*, Vol. 34, No. 5, pp. 213-245.

Davies, G. J., and Garland, J. G. (1975)

Solidification structures and properties of fusion welds, *Int. Met. Reviews*, Vol. 20, pp. 83-104.

Dolby, R. E. (1981)

Advances in the welding metallurgy of steel, Proc. of an Intl. Conf. on Advances in the physical metallurgy and applications of steel, organised by The Metals Society, University of Liverpool, 21-24 September.

Dowling, J. M., Corbett, J. M., and Kerr, H. W. (1986)

Inclusion Phases and the nucleation of acicular ferrite in submerged metal arc welds in high strength low alloy steels, *Metallurgical Transactions A*, Vol. 17A, pp. 1611-1623.

Dyson, D. J. and Holmes, B. (1970)

Effect of alloying addition on the lattice parameter of austenite, Journal of the Iron and Steel Institute, Vol. 208 pp. 469-474.

Easterling, K. E. (1983)

Introduction to physical metallurgy of welding, *Butterworths press.*

Easterling, K. E. (1985)

Solidification features of weld metal in *Metallography and interpretation of weld microstructures*, edited by Mccall, J. L and Olson, D. L., pp. 1 - 22.

Ecob, R. C and Ralph, B. (1981)

A model of the equilibrium structure of FCC / BCC interfaces, *Acta Metallurgica*, Vol. 29, pp. 1037-1046.

Eldis, G. T. (1977)

A critical review of data sources for isothermal transformation and continuous cooling transformation diagrams, in Symposium on *Hardenability Concepts with Applications to Steels*, AIME, U.S.A.

Enomoto, M. (1988)

Comparison of alloy element partition behaviour and growth kinetics of proeutectoid ferrite in Fe-C-X alloys with diffusion kinetic theory, *Trans. Iron Steel Institute Japan*, Vol. 28, pp. 826-835.

Evans, G. M. (1986a)

Effects of silicon on the microstructure and properties of C-Mn all-weld-metal deposits, *Metal Construction*, Vol. 18, pp. 438R-444R

Evans, G. M. (1986b)

The effect of molybdenum on the microstructure and Properties of C-Mn all-weld metal deposits, IIW-Doc. II-A-666-86.

Evans, G. M. (1988)

The effect of chromium on the microstructure and Properties of C-Mn all-weld metal deposits, IIW-Doc. II-A-739-88.

Farrar, R. A., and Watson, M. N. (1979)

Effect of oxygen and manganese on submerged arc weld metal microstructures, *Metal Construction*, June, pp. 285 - 286.

Farrar, R. A., and Harrison, P. L. (1987)

Review-Acicular ferrite in carbon-manganese weld metals: an overview, *Journal of Materials Science*, Vol. 22, 1987, pp. 3812 - 3820.

Ferrante, M., and Farrar, R. A. (1982)

The role of oxygen rich inclusions in determining the microstructure of weld metal deposits, *Journal of Materials Science*, Vol. 17, pp. 3293 - 3298.

Fick, I. J. and Rogerson, J. H. (1977)

Low Carbon Structural Steels for the Eighties, Institute of Metals, London, 1977, IIIB, pp. 41-50.

Fleck, N. A., Grong, Φ., Edwards, G. R., and Matlock, D. K. (1986)

The role of filler metal wire and flux composition in submerged arc weld metal transformation kinetics, *Welding Journal*, 113s - 121s.

Garland, J. G. and Kirkwood, P. R (1975)

Towards improved submerged arc weld metal, *Metal Construction*, Vol. 7, pp. 275-283.

Gilmour, J. B., Purdy, G. R. and Kirkaldy, J. S. (1972)

Partition of manganese during the proeutectoid ferrite transformation in steel, *Metallurgical Transactions*, Vol. 3, pp. 3213-3222.

Gray, T. G. F. and Spence, J. (1982)

Rational Welding Design, second edition, Butterworths, London.

Gretoft, B., Bhadeshia, H. K. D. H., and Svensson, L.-E. (1986)

Development of Microstructure in the fusion zone of weld deposits, *Acta Stereol.*, 5 / 2 : pp 365 - 371.

Grong, Φ., Siewart, T. A., and Edwards, G. R. (1986)

Effects of deoxidation practice on the transformation behaviour and toughness of steel welds, *Welding Journal*, pp. 279s - 288s.

Grong, Φ. and Matlock, D. K. (1986)

Microstructural development in mild and low alloy steel weld metals, *Int. Met. Reviews*, Vol. 31, No. 1, pp. 27-48.

Hall, B. (1990)

The microstructure and properties of laser welds in steel, Ph.D thesis, University of Cambridge.

Harrison, P. L., and Farrar, R. A. (1981)

Influence of oxygen - rich inclusions on the γ to α phase transformation in high strength low alloy (HSLA) steel weld metals, *Journal of Materials Science*, Vol. 16, pp. 2218 -2226.

Harrison, P. L., and Farrar, R. A. (1987a)
Microstructural development and toughness of C-Mn and C-Mn-Ni weld metals, Part - 1 - Microstructural development, ***Metal Construction***, pp. 392R - 399R.

Harrison, P. L., and Farrar, R. A. (1987b)
Microstructural development and toughness of C-Mn and C-Mn-Ni weld metals, Part - 2 - Toughness, ***Metal Construction***, p. 448R.

Harrison, P. L., and Farrar, R. A. (1989)
Applications of continuous cooling transformation diagrams for welding of steels, ***International Materials Reviews***, Vol. 34, no.1, pp. 35 - 51.

Hillert, M. (1951)
Jernkontorets Ann, 1951, Vol. 135, p. 403.

Hillert, M. (1953)
Nuclear composition - a factor of interest in nucleation, ***Acta. Metallurgica***, Vol. 1, pp. 764-766.

Hillert, M. (1981)
An analysis of the effect of alloying elements on the pearlite reaction, ***Solid → Solid phase transformations,*** TMS-AIME, Warrendale, Pennsylvania, USA, 1981, p. 789.

Hippsley, C. A. (1981)
Precipitation sequences in heat-affected zone of 2.25 Cr-1 MO steel during stress-relief heat treatment, *Metal Science*, 1981, pp. 137-147.

Hippsley, C. A., and Lewandowski, J. J. (1988)
The nucleation of high temperature brittle intergranular fracture in 2.25Cr-1 Mo steel, *Metallurgical Transactions A*, Vol. 19A, pp. 3005-3011.

Homma, H., Ohkita, S., Matsuda, S., and Yammamoto, K. (1987)
Improvement of HAZ toughness in HSLA steel by introducing finely dispersed Ti-oxide, *Welding Journal*, pp. 301s - 309s.

Horri, Y., Ohkita, S., Wakabayashi, M. and Namura. M. (1988)
Study on the toughness of large-heat input weld metal for low temperature-service TMCP steel, ***Nippon Steel Technical Report***, no. 37., April, 1988, p. 3.

Imagumbai, M., Chijiiwa, R., Aikawa, N., Nagumo, M., Homma, H., Matsuda, S. and H. Mimura (1985)
HSLA Steels: Metallurgy and Applications, eds. J. M. Gray, T. Ko, Z. Shouhua, W. Baorong and X. Xishan, ASM International, Ohio.

Ion, J. C., Easterling, K. E. and Ashby, M. F. (1984)
A second report on diagrams of microstructure and hardness for heat-affected zones in welds, *Acta Metallurgica*, Vol. 32, No. 11, pp. 1949-1962.

Ishiguro, T., Murakami, Y., Ohnishi, K. and Watanabe, J. (1982)
A $2\frac{1}{4}$Cr- 1 Mo pressure vessel steel with improved creep rupture strength, *ASTM Special Technical Publication 755, 1982*, pp. 129-147.

Ito, Y., and Nakanishi, M. (1976)
Study on Charpy impact properties of weld metals with submerged arc welding, *The Sumitomo Search*, 15, May, p. 42.

Ito, Y., Nakanishi, M., and Komizo, Y. (1982)
Effects of oxygen on low carbon steel weld metal, ***Metal Construction***, September, pp. 472 - 478.

Josefsson, B., Kvist, A. and Andrén, H. O. (1987)
Atom probe microanalysis of weld metal in a submerged arc welded chromium-molybdenum steel, *Journal de Physique*, C6-435.

Josefsson, B., Kvist, A. and Andrén, H. O. (1989)
Microstructure and thermodynamic behaviour of a chromium molybdenum submerged arc weld metal in the as welded state, Second International Conference on *Trends in Welding Research*, 14 - 18, May, Gatlingburg, Tennessee, U.S.A.

Kar, R. J., and Todd, J. A. (1982)

Alloy modification of thick-section $2\frac{1}{4}$Cr-1Mo steel, *ASTM Special Technical Publication 755, 1982*, pp. 208-228.

Kaufman, L., Clougherty, E. V. and Weiss, R. J. (1963)

The lattice stability of Metals - III: Iron, *Acta Metallurgica*, Vol. 11, p. 323.

King, A. D and Bell, T. (1975)

Crystallography of grain-boundary proeutectoid ferrite, *Metallurgical Transactions A*, Vol. 6A, pp. 1419-1429.

Kinsman, K. R., Eichen, E., and Aaronson, H. I. (1975)

Thickening kinetics of proeutectoid ferrite plates in Fe-C alloys, *Metallurgical Transactions a*, Vol. 6A, pp 303-317.

Kirkaldy, J. S. (1991)

Diffusion-controlled phase transformations in steels - Theory and applications, *Scandinavian Journal of Metallurgy*, Vol. 20, pp. 55-61.

Klueh, R. L. and Swindeman, R. W. (1986)

The microstructure and Mechanical properties of a Modified 2.25 Cr-1 Mo steel, *Metallurgical Transactions A*, Vol. 17A, June.

Kluken, A. O. and Grong, Φ. (1990)

Mechanisms of inclusion formation in Al-Ti-Si-Mn deoxidized steel weld metals, *Metallurgical Transactions A*, Vol. 20A, pp. 1335-1349.

Kluken, A. O., Grong, Φ. and Rϕrvik, G. (1990)

Solidification microstructures and phase transformations in Al-Ti-Si-Mn deoxidized steel weld metals, *Metallurgical Transactions A*, Vol. 21A, pp. 2047-2058.

Kluken, A. O., Grong, Φ. and Hjelen, J. (1991)

The origin of transformation textures in steel weld metals containing acicular ferrite, *Metallurgical Transactions A*, Vol. 22A, pp. 657-663.

Kou, S. (1987)

Welding Metallurgy, John wiley and sons press.

Krishtal, M. A. (1963)

Diffusion processes in iron alloys, Israel program for scientific translations, Jerusalem, 1970.

Lange, W. F., Enomoto, M. and Aaronson, H. I. (1988)

The kinetics of ferrite nucleation at austenite grain boundaries in Fe-C alloys, *Metallurgical Transactions A*, Vol. 19A, March, pp 427-440.

Lee, J. and Hon, M. (1987)

The continuous cooling transformation of carbon-manganese steel, *Scripta Metallurgica*, Vol. 21, pp. 521-526.

Leslie, W. C. (1982)

'The Physical Metallurgy of Steels', p. 111, International Student edition, 1982, McGrawhill, Inc.

Liu, S., and Olson, D. L. (1986),

The role of inclusions in controlling HSLA steel weld microstructures, *Welding Journal Res. Suppl.*, pp. 139s-149s.

Lundin, C. D., Kelly, S. C., Menon, R., and Kruse, B. J. (1986)

Stress rupture behaviour of post weld heat treated 2-1/4 - 1 Mo steel weld metal, *WRC Bulletin*, 315, June.

McGrath, J. T., Chandel, R. S., Orr, R. F., and Gianetto, J. A. (1989)

A review of factors affecting the structural integrity of weldments in heavy wall reactor vessels, *Canadian Metallurgical Quaterly*, Vol. 28, No. 1, pp. 75-83.

Mclellan, R. B., and Dunn, W. W. (1969)

A quasichemical treatment of interstitial solid solution: its application to carbon austenite, *Journal of Phy. Chem. Solids*, Vol. 30, p. 2631.

Mittemeijer, E. J., Van Gent. A. and Van Der Schaff, P. J. (1986)

Analysis of transformation kinetics by non isothermal dilatometry, *Metallurgical Transactions A*, Vol. 17A, pp. 1441-1445.

Mori, N., Homma, H., Okita, S., and Asano, K. (1980)

The behaviour of B and N in notch toughness improvement of Ti-B bearing weld metals. IIW DOC IX-1158-80.

MTDATA (1989)

Metallurgical and Inorganic Thermodynamic bank, National Physical Laboratory, Teddington, U.K.

Nishioka, K. and Tamehiro, H. (1988)

Microalloying '88, International Symposium on Applications of HSLA Steel", Chicago, Illinois, USA pp. 1-9

Ohmori, Y., Ohtani, H. and Kunitake, T. (1971)

The bainite in low carbon low alloys high strength steels, *Transactions, ISIJ*, Vol. 11., pp. 250-259.

Oldland, R. B. and McPherson, R. (1984)

A direct method for determination the $\gamma \rightarrow \alpha$ transformations in welds, *Metals Forum*, Vol. 7, No. 1, pp. 55-60.

Olson, G. B. and Cohen, M. (1976)

A general mechanism of martensitic nucleation: Part II. FCC → BCC and other martensitic transformations, *Metallurgical Transactions A*, Vol. 7A., pp. 1905-1914.

Olson, G. B., Bhadeshia, H. K. D. H. and Cohen, M. (1990)

Coupled Diffusional/Displacive Transformations, Part II: Solute Trapping, *Metallurgical Transactions. A*, Vol. 21A, pp. 805-809.

Pargeter, R. J. (1983)

Acicular ferrite and inclusions in C-Mn steel weld metals, *The welding institute research bulletin*, July, pp. 215-220.

Patel, J. R. and Cohen, M. (1953)

Criterion for the action of applied stress in the martensitic transformation, *Acta Metallurgica*, Vol. 1, pp. 531-538.

Purdy, G. R. (1978)

The dynamics of transformation interfaces in steels - I, The ferrite-austenite interface in Fe-C-Mo alloys, *Acta Metallurgica*, Vol. 26, pp. 477-486.

Raghavan, M. and Thomas, G. (1971)

Structure and mechanical properties of Fe-Cr-C-Co steels, *Metallurgical Transactions,* Vol. 2, pp. 3433-3439.

Reed, R.C., and Bhadeshia, H. K. D. H. (1989)

Complete reaustenitisation in multirun steel weld deposits, presented to the 2nd international conference on *Trends in Welding Research*, Gatlingburg, Tennessee, USA, 14-18, May, 1989.

Reed, R. C. (1990)

The characterisation and modelling of multipass steel weld heat - affected zones, Ph.D thesis, University of Cambridge, 1990.

Rees, G. I. (1990)

Unpublished Research.

Ricks, R. A., Barrite, G. S., and Howell, P. R. (1981)
The influence of second phase particles on diffusional phase transformation in steels, Proc. Intl. Conf. on *Solid State Phase Transformation* , Pittsburgh, Pa., AMIE, pp. 463 - 465.

Ricks, R. A., Howell, P. R., and Barrite, G. S. (1982)
The nature of acicular ferrite in HSLA steel weld metals, *Journal of Materials Science*, Vol. 17, 1982, pp. 732 - 740.

Rosenthal, D. (1941)
Mathematical theory of heat distribution during welding and cutting, *Welding Journal Res. Suppl.*, Vol. 20, pp. 220s-234s.

Rudberg. E (1952)
Jernkontorets Ann, 1952, Vol. 136, p. 91.

Russell, K. C. (1969)
Grain boundary nucleation kinetics, *Acta Metallurgica*, Vol. 17, August, pp. 1123-1131.

Ryder, P. L. and Pitsch, W. (1966)
The crystallographic analysis of grain-boundary precipitation, *Acta Metallurgica*, Vol. 14, pp. 1437-1448

SAS software user's guide (1987)
Non linear procedure, *SAS institute Inc.,*, SAS, Cary, NC 27511-8000.

Savage, W. F. (1980)
Solidification, segregation and weld imperfections, *Welding in the world*, Vol. 18, No. 5/6, pp. 89-114.

Savage, W. F. and Aaronson, A. H. (1966)
Welding Journal, Vol. 45, pp. 85-90.

Savage, W. F., Nippes, E. F., and Erickson, J. S. (1976)
Solidification mechanisms in fusion welds, *Welding Journal Res. Suppl.*, August. pp. 213s-221s.

Shiflet, G. J., Bradley, J. R., and Aaronson, H. I. (1978)
A re-examination of the thermodynamics of the proeutectoid ferrite transformation in Fe-C alloys. *Metallurgical Transactions A*, Vol. 9A, pp. 999 - 1008.

Siller, R. H. and Mclellan, R. B. (1969)
The variation with composition of the diffusivity of carbon in austenite, *Trans.Met.Soc.-AIME*, Vol. 245, pp. 697-700.

Siller, R. H. and Mclellan, R. B. (1970)
The application of first order mixing statistics to the variation of the diffusivity of carbon in austenite, *Metallurgical Transactions*, Vol. 1, pp. 985-988.

Snieder, G. and Kerr, H. W. (1984)
Effects of chromium additions and flux type on the structure and properties of HSLA steel submerged arc weld metal, *Canadian Metallurgical Quarterly*, Vol. 23, pp. 315-325.

Smith, R. P. (1946)
Equilibrium of Iron-Carbon alloys with mixtures of $CO - CO_2$ and $CH_4 - H_2$, *J. American Chem. Soc.*, Vol. 68, pp.1163-1175.

Smith, C. S. (1948)
Grains, Phases and Interfaces: An interpretation of microstructure, *Trans. American Institute of Mining and Metallurgical Engineers*, Vol. 175, pp. 15-51.

Srinivasan, N. K. (1985)
Welding Technology, First edition, Khanna Publishers, Delhi, India.

Stark, I., Smith, G. D. W., and Bhadeshia, H. K. D. H. (1987)
The element redistribution associated with 'the incomplete reaction phenomenon' in bainitic steels: An atom probe investigation, *Phase Transformations '87*.

Steven, W., and Haynes, A. G. (1956)

The temperature of formation of martensite and bainite in low alloy steels, some effects of chemical composition, *Journal of the Iron and Steel Institute*, Vol. 183, pp. 349-359.

Strangwood, M., and Bhadeshia, H. K. D. H. (1987a)

The Mechanism of acicular ferrite transformation in steel weld deposits, Proc. Conf. on *Advances in Welding Science and Technology*, ASM international, Ohio, U.S.A, 1987, pp. 187 - 191.

Strangwood, M., and Bhadeshia, H. K. D. H. (1987b)

Nucleation of ferrite at ceramic/steel interfaces, presented at at *Phase Transformations, '87.*

Strangwood, M., and Bhadeshia, H. K. D. H. (1987c)

Solute-segregation, Oxygen content and the transformation start Temperature of steel welds, Proc. Intl. Conf.on *Welding Metallurgy of Structural Steels.* AIME, Warrendale, Pennsylvania. Edited by J. Y. Koo, pp. 495-504.

Strangwood, M. (1987)

The prediction and assessment of weldmetal microstructures, Ph.D Thesis, University of Cambridge, 1987.

Sugden, A. A. B. (1988)

Towards the prediction of weld metal properties, Ph.D Thesis, University of Cambridge.

Sugden, A. A. B., and Bhadeshia, H. K. D. H. (1988a)

A Model for the strength of as deposited regions of steel weld metals, *Metallurgical Transactions A*, Vol. 19A., June, pp. 1597 - 1602.

Sugden, A. A. B., and Bhadeshia, H. K. D. H. (1988b)

The nonuniform distribution of inclusions in low-alloy weld deposits, *Metallurgical Transactions A*, Vol. 19A, pp. 669-674.

Sugden, A. A. B., and Bhadeshia, H. K. D. H. (1989a)

Microstructural Entropy and the Scatter in Toughness, presented to the 2nd international conference on *Trends in Welding Research*, Gatlingburg, Tennessee, USA, 14-18, May, 1989.

Sugden, A. A. B., and Bhadeshia, H. K. D. H. (1989b)

Lower Acicular Ferrite, *Metallurgical Transactions A,* Vol. 20A, 1989, pp. 1811-1818.

Sugden, A. A. B., and Bhadeshia, H. K. D. H. (1989c)

Thermodynamic Estimation of the Liquidus, Solidus and Ae3 Temperatures and Phase Compositions for Low-Alloy Multicomponent Steels, *Materials Science and Technology*, 1989, Vol. 5, pp. 977-984.

Svensson, L.- E., Gretoft, B., and Bhadeshia, H. K. D. H (1986)

An analysis of cooling curves from the fusion zone of steel weld deposits, *Scand. Journal of Met.*, Vol. 15, 97-103.

Svensson, L.-E., and Bhadeshia, H. K. D. H (1988)

The design of submerged arc weld deposits for high strength steels, Proc. of the Vienna Conference of the International Institute for Welding (IIW), 1988.

Takahashi, M., and Bhadeshia, H. K. D. H. (1989)

The interpretation of dilatometric data for transformations in steels, *Journal of Materials Science Letters*, Vol. 8, pp. 477 - 478.

Takahashi, M., and Bhadeshia, H. K. D. H. (1990)

Transition from upper to lower bainite, *Materials Science and Technology*, Vol. 6, July, pp. 592-603.

Trivedi, R and Pound, G. M. (1967)

Effect of concentration dependent diffusion coefficient on the migration of interphase boundaries, *J. Appl. Physics*, 1967, Vol. 38, pp. 3569-3576.

Trivedi, R. (1970)
The role of interfacial energy and interface kinetics during the growth of precipitate plates and needles, *Metallurgical Transactions*, Vol. 1, April, 921-927.

Uhrenius, B. (1973)
Optimisation of parameters describing the carbon metal interaction in ternary austenite, *Scandinavian Journal of Metallurgy*, Vol.2, pp. 177-182.

Underwood, E. E. (1970)
Quantitative Stereology, Addison Wesley, Reading MA. USA.

Vandermeer, R. A., Vold, C. L. and King, Jr. W. E. (1989)
Modelling of diffusional growth during austenite decomposition to ferrite in polycrystalline Fe-C alloys, presented to the 2nd international conference on *Trends in Welding Research*, Gatlingburg, Tennessee, USA, 14-18, May, 1989.

Vandermeer, R. A. (1990)
Modelling of diffusional growth during austenite decomposition to ferrite in poly crystalline Fe-C alloys, *Acta Metallurgica*, Vol. 38, pp. 2461-2470.

Wada, T., Wada, H., Elliot, J. F. and Chipman, J. (1972)
Activity of carbon and solubility of carbides in the FCC Fe-Mo-C, Fe-Cr-C, and Fe-V-C alloys, *Metallurgical Transactions*, Vol. 3, pp. 2685-2872.

Wada, T., and Eldis, G. T. (1982)
Transformation characteristics of $2\frac{1}{4}$Cr-1Mo steel. *ASTM Special Technical Publication 755, 1982*, pp. 343-360.

Watson, J. D. and McDougall, P. G. (1973)
The crystallography of Widmanstätten ferrite, *Acta Metallurgica*, Vol. 21, pp. 961-973.

Weiss, I., Jonas, J. J. and Ruddle, G. E. (1981)
Process Modelling Tools, Ed. J. F. Thomas, Jr., American Society for Metals, P. 97.

Wever . F (1928)
Zeitschrift fur Metallkde, Vol. 20, pp. 363-370.

Yamamoto, K., Matsuda, S., Haze, T., Chijiiwa, R. and Mimura, H. (1987)
Residual and Unspecified Elements in Steel, ASM International, Ohio, USA, pp. 1-24.

Yang, J. R. (1987)
Development of Microstructure in High strength Steel Weld Deposits, Ph.D Thesis, University of Cambridge.

Yang, J. R. and Bhadeshia, H. K. D. H. (1987a)
Thermodynamics of the acicular ferrite transformation in alloy-steel weld deposits, Proc. Conf. on *Advances in Welding Science and Technology*, ASM international, Ohio, U.S.A, pp. 209 - 213.

Yang, J. R., and Bhadeshia, H. K. D. H. (1987b)
Reaustenitisation from Bainite, presented at *Phase Transformations '87*.

Yang, J. R., and Bhadeshia, H. K. D. H. (1987c)
Reaustenitisation in Steel Weld Deposits, Proceedings of an International Conference on *Welding Metallurgy of Structural Steels*, published by The Metallurgical Society of the AIME, Warrendale, Pennsylvania. Edited by J. Y. Koo, pp. 549-563.

Yang, J. R., and Bhadeshia, H. K. D. H. (1989a)
Orientation relationships between adjacent plates of acicular ferrite in steel weld deposits, *Materials Science and Technology*, Vol. 5, pp. 93-97.

Yang, J. R., and Bhadeshia, H. K. D. H. (1989b)
Reaustenitisation Experiments on Some High-Strength Steel Weld Deposits, *Materials Science and Engineering*, Vol. A118, pp. 155-170.

Yang, J. R., and Bhadeshia, H. K. D. H. (1990)
The Dislocation Density of Acicular Ferrite in Steel Welds, *Welding Journal,* Vol. 69, pp. 305s-307s.

Yurioka, N., Okumura, O., Kasuya, T., Cotton, H. J. U. (1987)
Prediction of HAZ hardness of transformable steels, *Metal Construction*, April, P 217R-223R.

Zener, C. (1946)
Kinetics of the decomposition of austenite, *Amer. Inst. for Mining and Metallurgical Engg.*, Vol. 167, pp. 550-595.

Zener, C. (1949)
Theory for growth of spherical precipitates from solid solutions, J. of App. Physics, Vol. 20, pp.950-952.

Zener, C. (1955)
Impact of magnetism upon metallurgy, *Transactions of AIME*, May, pp. 619-630.

- - - - - - - - - - - - - - - - - - - -