

THESIS

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A STUDY OF SIMULATED WELD HEAT
AFFECTED ZONE STRUCTURES AND PROPERTIES
OF HY-80 LOW ALLOY STEEL

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SYNOPSIS

Single and double cycle simulation techniques have been employed to investigate the structure and properties of the heat affected zone in HY-80 steel due to a single submerged arc weld run and to the deposition of a tempering bead on a pre-existing edge bead heat affected zone. The thermal cycles used for simulation had peak temperatures of 1275°C, 930°C, 765°C and 650°C and corresponded to those experienced in the parent material during the submerged arc welding of 1.5 in. thick plate using a nominal heat input of 54kJ. in⁻¹ and a preheat temperature of 120°C. Double cycling utilized specific combinations of these thermal cycles. Half the total number of simulated specimens received post cycle heat treatment at 650°C. Charpy V-notch impact transition curves and tensile and hardness data have been obtained for each condition studied and structural examination has employed the carbon extraction replica technique in the electron microscope. The results indicate that the properties of the weld heat affected zone in HY-80 steel are superior to those obtained for QT-35 steel but inferior to those of the HY-80 parent material. The tempering bead technique has been shown to have, at best, only a limited tempering effect on the grain coarsened region of an edge bead heat affected zone and can also cause an increase in the susceptibility of this region to brittle fracture. Post weld heat treatment at 650°C is recommended, wherever possible, for HY-80 weldments. The calculation of weld heat affected zone thermal cycles, which correlate well with the practical situation, has been made possible by the development of a series of computer programs. These programs include functions to account for the release of latent heat from the solidifying weld pool and the variation of thermal conductivity with temperature.

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1. INTRODUCTION

Greater strength is always desired in a submarine structure and to obtain a balanced design for a new class of vessel a steel of higher yield strength is necessary. For any new steel to be worthwhile in terms of improved military characteristics its yield strength should be significantly higher than the steel already in use but such increase should be consistent with maintaining good ductility, toughness, resistance to low-cycle fatigue and weldability. In the U.S.A. up until about 1953 conventional submarines were produced in HTS fabricated in the as-rolled condition and having a yield strength of 21 tonf. in⁻². In order to obtain the higher yield strength necessary to ensure a good strength: weight ratio in the design of a new class of nuclear powered submarines of larger hull diameter, HTS steel was modified by small alloying element additions. However, this procedure did not produce a significant strength increase and further development used the quenched and tempered Krupp (1894) armour plate steel as a basis. The outcome of this research was a quenched and tempered low carbon steel ($< 0.18\%$) containing approximately 3% nickel, 1.5% chromium and 0.5% molybdenum, designated STS, which gave an attractive combination of high yield strength and good toughness. With composition variation to ensure good hardenability in different plate thicknesses this steel was referred to as HY-80 (high yield strength; 80,000 lbf. in⁻², (≈ 35 tonf. in⁻²) minimum) and produced in plate form to U.S. military specification MIL-S- 16216. HY-80 steel has been used exclusively for the production of pressure hulls, ring stiffeners, insert reinforcements etc., for the U.S. nuclear powered submarine fleet. The steel is now being used in the U.K. for the fabrication of the latest additions to the 'Dreadnought' class of nuclear powered submarines, superceding the home produced QT 35 quenched and tempered steel for this purpose. HY-80 belongs to the class of low carbon martensitic steels which have an Ms temperature of the order of 400°C so that on rapid cooling auto-tempered martensite is produced, a feature which is of advantage in ensuring good weldability. The low carbon content maintains toughness and weldability while a sufficient alloy content imparts necessary hardenability. Small amounts of elements, such as molybdenum, may be present to confer resistance to temper brittleness and vanadium may be added to provide resistance to tempering.

The environmental temperature in which a submarine operates depends both upon the submerged depth and geographical latitude,



X 4

Figure 1. Hydrogen induced crack in the heat affected zone of a multi-run weld in a low alloy steel (after Dolby,(1)).

consistent with its strategic operation, and may vary within the range -35 to 50°C . The pressure hull must withstand the hydrostatic loading of uniform external pressure associated with the depth at which it is operating and, during its operational lifetime, may experience many major depth excursions with a resultant accumulation of about 10^4 to 3×10^4 cyclic loadings. In addition, the hull may also have to withstand, in the event of enemy attack, dynamic loading in the form of a shock wave due to the underwater detonation of an explosive device.

Since welding is extensively used in submarine fabrication, exclusively in the case of the pressure hull, it is important to be aware of its effect on the structure and properties of the parent material. The thermal cycles produced by the welding process in a region of the parent material adjacent to the weld can markedly alter the structure of the steel and the associated mechanical properties of this heat affected zone may show wide variations. In low alloy steels the fracture toughness of the heat affected zone may be severely impaired by welding and this can lead to cold cracking, of an underbead nature, in the heat affected zone either immediately after welding or after a period of several days has elapsed. If such cracks are present in service they may propagate as fatigue cracks under low-cycle loading. However, whether or not propagation has to some extent occurred in this manner, such cracks impose a serious limitation on brittle fracture resistance. If the environmental temperature is sufficiently low, heat affected zone cold cracks may propagate in a brittle manner under conditions of dynamic loading, leading to catastrophic failure of the pressure hull. It is, of course, feasible that, in the absence of fatigue or cold cracking, brittle failure could easily occur from a region of geometrical stress concentration, such as an undercut at the weld toe, or the line of edge bead - parent plate fusion which would, of course, be dependent upon edge bead shape. In such instances the region of the HAZ in which fracture initiation and early propagation would be favoured would be that with the poorest fracture toughness. A typical underbead crack⁽¹⁾ in a multi-run weld is shown in figure 1. It has been established that heat affected zone cold cracking can occur only when the following condition is satisfied viz., a sufficient concentration of hydrogen is present in a susceptible microstructure of a weld heat affected zone across which a stress, usually a residual welding stress, is acting. The practical methods of avoiding this type of cracking are all based on the control or elimination of one or more of these factors. By suitable electrode choice and proper design, the concentration of hydrogen and the degree of residual stress respectively, present in a weld heat affected zone may be largely controlled. However, in low alloy steels susceptible structures of poor fracture toughness are almost inevitably present due to the very nature of the welding process, even if this is carried out within specified heat

input and preheat temperature ranges.

Current U.S. Navy and Ministry of Defence (Navy) specifications call for the application of the tempering bead technique for all multi-run welds in HY-80 steel in an effort to eliminate, as far as possible, the incidence of heat affected zone cold cracking and the existence of a structure of low fracture toughness adjacent to the edge bead - parent plate fusion line. The technique involves the arrangement of consecutive weld runs such that the heat affected zone due to the run fusing with the parent material is tempered by a subsequent run. The specifications stress the importance of applying this technique to the surface layer of a weld so that the heat affected zone at the plate surface due to an edge bead is tempered by a subsequent run. If the tempering bead is misaligned or is deposited using an inappropriate heat input then its effect on the edge bead heat affected zone structure and consequent mechanical properties may be serious, as further thermal cycling above the initial range could occur.

Subcritical post weld heat treatment invariably results in heat affected zone structures with mechanical properties which approach those of the parent material. In many situations, however, it is not employed due to the problem of the inherent distortion which often accompanies the furnace treatment of large or relatively complex fabrications.

This present research, which forms part of a general programme to investigate the heat affected zone structure and properties of low alloy steels, has a four-fold object:

- (a) to establish for a given heat input condition and preheat temperature the nature of the weld heat affected zone due to a single weld run.
- (b) by employing double cycle simulation techniques, using the same welding conditions as in (a), to predict the effect of the tempering bead position on the structure and properties of a pre-existing edge bead heat affected zone.
- (c) to ascertain the effectiveness of post weld heat treatment in restoring the properties produced by (a) and (b) to those of the parent material.
- (d) to indicate the general nature of tempering in a pre-existing heat affected zone due to the heat flow from subsequent weld runs in a multi-run weld.

Four thermal cycles typical of those experienced by separate regions of the heat affected zone during the submerged arc welding of 1.5 in. thick HY-80 steel plate were obtained, two by actual measurement and two by computer programming based on an analytical solution to the basic differential equation for 3-dimensional heat flow. These cycles were then used to simulate, in specimen blanks of appropriate size, using a controlled electric resistance heating apparatus, structures corresponding to the four separate heat affected zone regions. The blanks were suitable for the production of Charpy V-notch impact, metallographic and Hounsfield tensile test specimens.

Single cycled specimens were used to determine the structure and properties of the representative heat affected zone regions by a single weld run and to provide a standard by means of which the results obtained for double cycled specimens could be compared. Dual combinations of the four basic cycles were employed to reproduce conditions in the heat affected zone of an edge bead due to the deposition of a tempering bead at various positions relative to the edge bead surface. The double cycling procedure also gave an insight into the tempering processes occurring in the heat affected zone during the deposition of a multi-run weld. Approximately half the total number of single and double cycled specimens were post cycle furnace heat treated to simulate post weld heat treatment.

The Charpy V-notch impact properties of the as-cycled and post cycle heat treated simulated regions gave a positive indication of their relative susceptibilities to brittle failure. These results were correlated with the microstructure determined by electron and, to a much lesser extent, optical microscopy and the tensile properties in an effort to ascertain the role of microstructure and strength on the notch toughness properties of the various heat affected zone conditions investigated.

A necessary adjunct to this investigation, due to a material limitation, was the use of computer programming to obtain two of the required thermal cycles. By considering both the variation of thermal properties of the material with temperature and the release of latent heat by the solidifying weld pool, close correlation was obtained with cycles experienced within an actual heat affected zone during welding. An integral part of this research has, therefore, been the development of a series of computer programs which can be used to give a reasonably accurate prediction of the thermal cycle experienced by any point in the weld heat affected zone.

2. LITERATURE REVIEW

2.1. THE APPLICATION OF HEAT FLOW THEORY TO FUSION WELDING

Virtually since its inception as a joining process there has been an appreciation of the importance of the thermal cycles produced by welding in modifying the structure and properties of the parent material adjacent to the actual weld deposit. Theoretically it should be possible to calculate:

- (a) the temperature distribution at any time during or after welding.
- (b) the maximum temperatures reached at any point during welding.
- (c) the heating and cooling rates at any point in the parent structure.

If such calculation can be achieved then it should be possible to predict and control:

- (a) structural changes in the heat affected zone and perhaps even in the solidified weld pool.
- (b) distortion due to thermal cycling and restraint.

A large amount of work has been done on the theory of heat flow in welding but even in the simplest case a discrepancy exists between theory and practice. The extent of this discrepancy is too great in some instances for the theory to be of much consequence or use. Developments have improved accuracy but only at the expense of mathematical difficulty and time consuming calculation. However, the position has recently been vastly improved by the advent and use of high speed digital computers.

The following two sections review accepted theories of heat flow as applied to fusion butt welding and their correlation with the practical situation.

* 2.1.1. Basic Heat Flow Theory

The basic theory of heat flow due to a moving source originated in connection with arc welding. Many early attempts, both experimental and theoretical, were made to describe the temperature situation created by arc welding. These were approximate in nature and were reviewed by SPRARAGEN and CLAUSEN (2). It is evident from this review that there are two fundamental approaches which can be made to heat flow from a moving source:

- (a) to consider the progression of welding as due to an infinite number of point sources of heat
- (b) to consider a single, constant velocity point source of heat.

The theory of heat flow due to a moving point source was first applied to welding by ROSENTHAL (3,4), using the experimentally established principle of the 'quasi-stationary' state. This state is obtained when an observer in a frame of reference moving with the point source would fail to notice any change in the temperature field with time. In welding the 'quasi-stationary' state exists when a weld bead is deposited on the surface of a plate in the region which is removed 3 to 6 in. from the edges and when the arc moves with constant velocity and has a constant rate of energy consumption (5). The principle, therefore, does not apply at the start and finish of a weld but is otherwise correct. This situation may appear obvious but by virtue of its application the mathematical aspects of the problem were made much more tractable.

Beginning with the partial differential equation for 3-dimensional heat flow expressed in rectangular co-ordinates which are referred to a fixed origin in the solid,

$$\frac{\partial^2 T}{\partial x^2} + \frac{\partial^2 T}{\partial y^2} + \frac{\partial^2 T}{\partial z^2} = 2\lambda \frac{\partial T}{\partial t} \quad 2.1.1$$

ROSENTHAL³ applied the principle of the 'quasi-stationary state' and obtained the expression

$$\frac{\partial^2 \phi}{\partial \xi^2} + \frac{\partial^2 \phi}{\partial y^2} + \frac{\partial^2 \phi}{\partial z^2} - (\lambda v)^2 \phi = 0 \quad 2.1.2$$

where the substitution $\xi = x - vt$, makes the point source the origin. Unless where otherwise stated, the symbols used in these equations and those that follow are defined and listed in Table 1, page 165.

ROSENTHAL (4,6) applied the theory to heat distribution during welding and cutting and to other metal treatments. He used several assumptions in an effort to simplify the mathematics involved, some of which could lead to significant errors. These assumptions were:

- (a) that heat flow is entirely by conduction through the plate and heat losses from the plate surface are negligible.
- (b) that, from the point of view of heat flow, the plate

is infinitely wide. This is tenable in many welding situations but cannot account for end or edge effects in the welding of relatively narrow plate.

- (c) that the welding arc is a point source of heat in 3- dimensional heat flow. This does not account for arc spread and could lead to serious error, particularly in heavy metal welding, as the point source concept is satisfactory for the electron heat input but not for the input from the positive column.
- (d) (i) that there is no change in the physical state of the metal in the vicinity of the arc, i.e. no melting occurs.
and
(ii) that the thermal properties of the material are constant throughout the temperature range considered.

These assumptions, d(i) and (ii), are incongruous and must lead to incorrect analysis. Despite the inherent shortcomings of these assumptions, the Rosenthal equations which follow, for the temperature distribution and cooling rate in thick and thin plates, have formed the basis for the majority of the study of heat flow in the vicinity of welded joints (For 2-dimensional heat flow in thin plate the heat source is assumed to have the shape of a line perpendicular to the plate surface. This form of source is used in all the equations for thin plate in this section).

For thin (2-dimensional) plate:

$$T - T_0 = \frac{Q'}{2\pi k g} \exp(-\lambda v \xi) K_0(\lambda v R) \quad 2.1.3.$$

$$\frac{dT}{dt} = 2\pi c \rho \left(\frac{g v}{Q'}\right)^2 (T - T_0)^3 \quad 2.1.4.$$

For thick (3-dimensional) plate:

$$T - T_0 = \frac{Q}{4\pi k R} \exp(-\lambda v \xi) \exp(-\lambda v R) \quad 2.1.5.$$

$$\frac{dT}{dt} = \frac{2\pi k v}{Q} (T - T_0)^2 \quad 2.1.6.$$

RYKALIN (7) making similar assumptions to those made by Rosenthal developed the following equations in which the point source was taken to be the origin of the co-ordinates. The equations which are effectively identical to Rosenthal's, have been reviewed by WESTBY (8) in a recent paper.

For thin plate:

$$T = \frac{Q'}{2\pi k g} \exp(-\lambda v x) K_0(R\lambda\sqrt{v^2 + 4ab}) \quad 2.1.7$$

where $b = \frac{2a}{c\rho g}$ = heat transfer factor
 α = coefficient of heat transfer to the surroundings.

The term $4ab$ may, therefore, be regarded as being introduced to compensate for heat losses to the surroundings.

For thick plate:

$$T = \frac{Q}{2\pi k R} \left\{ \exp[-\lambda v(x+R)] \right\} \quad 2.1.8$$

WELLS (9) indicated that the Rosenthal relationship between heat input and temperature distribution in thin plate could be simplified to:

$$Q = 8 k T \left(\frac{1}{5} + \frac{v d}{4\alpha} \right) \quad 2.1.9$$

where d is the width of the zone which has been raised T degrees centigrade above the initial plate temperature.

This simplified solution may be applied to the isotherm width anywhere in the heat affected zone and with the addition of APPS and MILNER'S (10) term,

$$\rho L v d$$

where L is the latent heat of fusion, it is possible to take into account the latent heat of fusion of the weld pool. However, the work (10) was limited by the fact that difficulty was encountered in the choice of suitable thermal constants, for substitution in the theoretical formulae employed.

ADAMS (11) also recognised the problem of latent heat and

suggested that the weld pool was not amenable to analysis. His solution was to rewrite Rosenthal's equations using the molten zone boundary as the parameter.

For thin plate:

$$\frac{1}{T_p - T_o} = \frac{v r' t \rho c \sqrt{2\pi e}}{Q'} + \frac{1}{T_m - T_o} \quad 2.1.10$$

For thick plate:

$$\frac{1}{T_p - T_o} = \frac{2\pi k \alpha e}{Q v} \left[2 + \left(\frac{v r'}{2\alpha} \right)^2 \right] + \frac{1}{T_m - T_o} \quad 2.1.11$$

where, r' = distance of a point from the fusion line.

T_m = melting point at the fusion line.

T_p = peak temperature at a distance r' from the fusion line.

This modification gives an improvement in the accuracy of predicted peak temperatures when compared with Rosenthal's results. It is, however, ineffective as regards the prediction of cooling rates.

GROSH et al., (12) obtained analytical solutions for the basic differential equation of heat flow, 2.1.1, which took into account certain variations of the thermal properties of a solid when heated by a moving source. The variations considered were of the type where thermal conductivity and the product of density and specific heat are similar functions of temperature; or, where these properties vary but their ratio, the thermal diffusivity, is constant. These workers assumed the following relationships:

$$k = k_o f'(T) \quad 2.1.12$$

and $\rho c = (\rho c)_o f'(T) \quad 2.1.13$

where k_o and $(\rho c)_o$ are positive constants and $f'(T)$ is the temperature derivative of some function of temperature, $f(T)$. It was also assumed that $f(T)$ and $f'(T)$ were differentiable and that $f'(T)$ was greater than zero to avoid violation of the Second Law of Thermodynamics i.e. to avoid negative values of thermal conductivity and product of density and specific heat

in the temperature ranges encountered. The latent heat of metallurgical transformations was neglected as were heat losses from the surface of the solid by convection and radiation. For the purpose of solution it was assumed that:

$$f'(T) = 1 + mT \quad 2.1.14$$

where m is a constant.

The solutions obtained were,
for thin plate:

$$T = \frac{1}{m} \left\{ \left[\frac{Q'm}{\pi c k_o} \exp(-\lambda \nu x) K_o(\lambda \nu r) + (1 + mT_o)^2 \right]^{\frac{1}{2}} - 1 \right\} \quad 2.1.15$$

and for thick plate:

$$T = \frac{1}{m} \left\{ \left[\frac{Qm}{2\pi k_o R} \exp(-\lambda \nu x) \exp(-\lambda \nu R) + (1 + mT_o)^2 \right]^{\frac{1}{2}} - 1 \right\} \quad 2.1.16$$

Equation 2.1.16, is further considered in the Appendix to this thesis.

WESTBY (8) found difficulty in solving, by numerical methods (finite difference technique), the differential equation, 2.1.1 for three-dimensional heat flow since it is non-linear. Also, using such methods the heat of fusion at constant temperature cannot be taken into account since the specific heat will be infinitely great. In order to avoid these difficulties he employed the following two variables.

$$\phi = \int_0^T \frac{k}{k_o} dT \quad 2.1.17$$

and

$$H = \int_0^T c dT$$

where ϕ is a modified temperature scale and, H , the heat content of the material. The method of introducing the modified temperature scale, ϕ , was earlier illustrated by SARJANT and SLACK (13) in connection with the temperature distribution in steel ingots

during cooling. Westby's work was carried out using mild steel as a basis and solid state transformations during cooling to ferrite and pearlite structures were assumed to occur at equilibrium values. The thermal conductivity and specific heat were allowed to vary with temperature. The assumption of equilibrium transformation temperatures, with neglect of austenite undercooling and associated subsequent transformation to martensitic and/or bainitic structures, imposes a rigorous restriction on the method and its general application to low alloy steels. The work was made possible by the use of a UNIVAC 1107 computer and Fortran programming.

2.1.2 Correlation between Theory and Experimental Results

Techniques used to measure the temperature distribution in metals during welding have ranged from the use of temperature sensitive lacquers (10,11), which change colour when heated above a critical temperature, crayons and wax pellets, which have fixed melting points, to thermocouples welded to the plate and connected to automatic recording instruments (14, 15, 16, 17). The accuracy of the latter method is considerably better than the others which measure temperatures to only $\pm$ ten per cent., since it is difficult to establish exactly the time at which they indicate a specific temperature. BEEVERS (18) has shown that one of the major difficulties in the measurement of temperatures in welding by means of thermocouples is the elimination of voltage pick-up from extraneous sources. COWARD and APPS (17) in a series of determinations showed the problem to be of negligible magnitude and indicated that temperatures could be measured to within $\pm 10^{\circ}\text{C}$ and $\pm 5^{\circ}\text{C}$ by employing chromel-alumel and rare metal thermocouples, respectively.

PORTEVIN and SEFERIAN (19) measured the temperature distribution during metallic arc welding of mild steel, aluminium and copper. Their results of temperature distribution and cooling rate showed little correlation with PINCZON's (20) constant velocity point source theory reviewed in literature reference 2. ROSENTHAL and SCHMERBER (21) compared experimental results for the arc welding of steel with Rosenthal's theory and found, in general, that the theory gave too high maximum temperatures and excessive cooling rates. These discrepancies were attributed to the molten weld pool acting as a distributed source of heat. There was, in addition, the problem of selecting a suitable thermal conductivity value.

HESS et al., (14) measured the cooling rates in the final runs of multi-run arc welds in low carbon and low alloy steel plates of varying thickness. They suggested that the values of cooling rates were essentially determined by heat input per unit length of weld, initial plate temperature, plate thickness and

dimensions and shape of the plate, Difficulty in correlation between experimental cooling curves and curves calculated using the Rosenthal theory was, in part, attributed to the necessity to employ constant values of the thermal properties over a wide temperature range.

JHAVERI et al., (22) employed dimensionless values for temperature and time which gave useful equations for plates of intermediate thickness. Discrepancies between experiment and theory were attributed to variation of the thermal properties with temperature and to the fact that the weld pool acted as a distributed source of heat.

GROSH and HAWKINS (23) compared experimental data obtained by NIPPES et al., (15) on thick (1.5 in) type 347 stainless steel plate with theoretical data based on equation 2.1.16. They likewise made a comparison with experimental data obtained by GROSH (24) on thin (0.25 in) type 304 stainless steel plate. In both cases thermal property variation was in accordance with data due to POWELL (25). Good correlation was obtained between theory and practice and it was considered that the constant value of thermal diffusivity employed was of secondary importance for the cases considered.

WESTBY (8) compared his theoretical results, based on the thermal properties and transformation characteristics of mild steel, with practical data obtained by CHRISTENSEN and GJERMUNDSEN (26) on the same material. Fairly good agreement was obtained for thermal cycles occurring in the parent plate for large arc power ($Q \simeq 8,800 \text{ cal. sec.}^{-1}$) and it is claimed that the temperature distribution within the molten weld pool can be calculated. This latter computation requires, however, the choice of a suitable value for thermal conductivity.

CHRISTENSEN et al., (27) deposited single beads on heavy pieces of mild steel and on aluminium by various welding techniques and these were examined with regard to the net heat transfer, to the dimensions and shape of the weld pool and to the temperature distribution inside and outside the pool. The results were presented in dimensionless measure for comparison with curves derived from a generalised point source model. Reasonably reliable dimensionless graphs were developed for a general prediction of weld metal cross sections for a wide range of welding conditions. Prediction of peak temperatures and corresponding centreline distances, and maximum cooling rates was not very accurate and the use of constant thermal properties and the neglect of latent heat effects may be responsible for this. However, the study did show that the distribution of liquid metal temperatures and the shape and dimensions of the weld pool were dependent upon the presence of directed currents within the weld pool.

In the context of the above discussion the major problems in the application of heat flow theory to the welding situation are:

- (a) The selection of real values of the thermal properties and the consideration of their variation with temperature.
- (b) The consideration of the latent heat of fusion in the solidifying weld pool and the heat evolution or abstraction due to solid state metallurgical transformations.
- (c) In practice heat is not supplied via a point or line source.
- (d) In many situations it is difficult to ascertain whether heat flow is purely 2- or 3-dimensional or a combination of these flow types.

Problems (a) and (b) have received some attention (8, 10, 12) but thermal property variation with temperature and the latent heat of fusion from the solidifying weld pool have not been treated together in an attempt to obtain better correlation with practical thermal cycles in the weld heat affected zone.

X 2.2. THE STRUCTURE OF LOW ALLOY STEELS

The mechanical properties of a low alloy steel are dependent upon its microstructure and this is basically a function of chemical composition. However, seldom are the so called 'equilibrium' structures found to exist throughout a low alloy steel plate or section since the ultimate structure is markedly dependent upon the conditions of austenitization, the heating rate to and the cooling rate from the austenitizing temperature, the presence of externally applied and inherent transformation stresses, the microstructure which existed prior to austenitization and microsegregation resulting from ingot solidification. It is the purpose of this section to correlate the individual effects and the influence of interaction of these variables in producing the range of structures found in the weld heat affected zone of a low alloy steel.

2.2.1. Austenitization

In low alloy steels austenite is formed, on cooling, by transformation of the γ - phase. On heating austenite is formed from ferrite but may also be produced from this phase by an increase in pressure.

The formation of austenite on heating merits attention due to its great practical importance but, somewhat surprisingly,

only a small amount of work has been published on this subject (28-31). ALBUTT and GARBER (28) have shown that, on heating, pure iron transforms from the b.c.c. α - phase to the f.c.c. γ - phase by a diffusional shear process at about 920°C. When carbides are present this reaction does not occur and, instead, austenite is nucleated at the carbide-ferrite interface within the range of temperatures between Ac_1 and Ac_3 . In low alloy steels containing chromium and molybdenum CONTRACTOR et al., (32) have found the carbides in the quenched and tempered structure to be alloy carbides and it has been shown (31, 33, 34) that such carbides resist solution in austenite to a greater extent than unalloyed iron carbide. This is of major importance if the thermal excursion above Ac_1 is of short duration since it can result in the production of inhomogeneous austenite and, possibly, lead to only partial carbide dissolution. Structural heterogeneity of this type can produce quite serious anomalies in the correlation of apparent structure with mechanical properties and is discussed in section 2.2.5.2.

Once nucleated, the austenite grains grow, become enriched with carbon and consume the ferrite until impingement occurs. PAXTON (29) claims that this mode of transformation persists even in the case of rapid heating rates. Once impingement has occurred the austenite grains start to coarsen in the normal manner. Some experimental work has been carried out (30) in this field in attempts to produce fine austenitic grain sizes on heating which should lead to fine ferrite grain sizes on cooling with a consequent improvement in mechanical properties.

For a given carbon content it has been shown (29) that the distribution and shape of the carbide in the original steel condition influences the degree to which the Ac_3 temperature is elevated by fast heating rates; the Ac_1 temperature is claimed to be insensitive to this. It is possible, however, that several other factors such as initial homogeneity, chemical composition and grain size may affect the values of the Ac_1 and Ac_3 temperatures. COWARD and APPS (17), while measuring thermal cycles during the welding of mild steel, found that both the Ac_1 and Ac_3 temperatures were increased by the fast heating rates involved. The Ac_3 temperature was considerably more elevated than the Ac_1 .

2.2.2. Austenite Transformation

When a steel is cooled from a temperature at which austenite is the stable phase to temperatures at which this phase becomes unstable it transforms ultimately to mixtures of ferrite and carbide or to martensite, or, to aggregates of all three structures. In low alloy steels the aggregate of ferrite and carbide is referred to as bainite which can occur in either two (35, 36) or three (37) forms. For a given carbon and alloying

element content the ultimate structure produced depends basically upon the cooling rate from the austenitizing temperature. The transformation of austenite to martensite and the forms of bainite will now be discussed.

2.2.2.1. Austenite - Martensite Reaction

Austenite transforms to martensite by a diffusionless transformation and is generally athermal, the degree of transformation depending upon the temperature. The temperature at which transformation begins on cooling is referred to as the M_s or 'martensite start' temperature and is strongly influenced by composition. Virtually every alloying element added to steel tends to lower this temperature. Several empirical formulae have been derived to enable prediction of the M_s temperature for steels containing given elements. One of the most successful has been that due to STEVEN and HAYNES (38):

$$M_s, ^\circ\text{C} = 561 - 474C - 33\text{Mn} - 17\text{Ni} - 17\text{Cr} - 21\text{Mo} \quad 2.2.1.$$

where C, Mn etc. refer to the amounts of element in weight per cent. Other formulae which have been used are available in the literature (39).

The austenite-martensite reaction is accompanied by an expansion and involves a shear transformation in the austenite lattice. BAIN (40) suggested a mechanism for such a lattice change which involved the formation of a body centred tetragonal martensitic lattice from the original f.c.c. austenite lattice. However, this model was found to be inconsistent with the observation that the austenite-martensite habit plane is essentially undistorted. BOWLES and MACKENZIE (41-44) and WECHSLER et al., (45) independently proposed a double shear theory which involves the Bain distortion occurring simultaneously with an inhomogeneous shear. Such a shear can occur either by twinning or by slip and the prevailing mode is dependent upon the composition and rate of cooling. KELLY and NUTTING (46) showed that carbon content affected the mode of inhomogeneous shear and BONISZEWSKI (47) indicated the effect of cooling rate. In plain carbon steels the carbon content separating the two morphologies is approximately 0.5% (21).

In the lower carbon steels martensite is formed by a shear occurring parallel to $\langle 110 \rangle$ on $\{111\}_\gamma$ (48-51) producing shear plates, termed laths by GRENINGER and TROIANO (50), which have a tendency to lie parallel to each other to form 'packets'. BOWLES and BARRETT (52) described the $\{111\}_\gamma$ plates as needles formed from degenerate $\{225\}_\gamma$ plates. Since the amount of retained austenite in these steels is small this tends to make X-ray analysis difficult and so the austenite-ferrite orientation relationships are not known. SPEICH and SWANN (53) have shown

that lath martensite contains a high density of dislocations arranged in plate-like cells. It is suggested that these cells may form co-operatively in adjacent regions to minimise the overall strain energy. EMBURY (54) claims that the dislocation cell structure may represent an early stage in the recovery of the transformation deformation.

Due to the presence of an appreciable amount of retained austenite in high carbon steels it is possible in their case to measure the orientation directly. The orientation is KURDJUMOV-SACHS (55) with an irrational habit varying from $(225)_\gamma$ to $(259)_\gamma$ with increasing carbon content. The second shear is substantially by twinning and each martensite plate is composed of parallel twin plates of thickness between 100 and 500 Å. Alloying elements do not affect the transition from dislocated to twinned martensite, since this does not appear to be a function of the transformation temperature. KELLY and NUTTING (46) have observed internally twinned plates in alloys containing only 0.2% carbon and the volume fraction occupied by twin plates appears to increase with carbon content. Thus, it seems that, within the range of carbon contents of 0.2 to 0.5% a change occurs from dislocated martensite arranged in laths to a twinned acicular martensite arranged in plates. SPEICH and WARLIMONT (56) have made similar observations in commercial low carbon steels. The internally twinned martensite plates ranged from 1 - 10 μm . in thickness and their internal defects consisted of a mixture of twins in lamellar array and uniformly distributed dislocations which exhibited a predominantly screw character i.e. have $\langle 111 \rangle$ line directions.

ABORN (57) found that low carbon martensite is very sensitive to tempering and a process known as auto-tempering can occur during cooling below the relatively high M_s temperature in low carbon, low alloy steels. The extent to which auto-tempering occurs depends both upon the temperature at which martensite forms and the subsequent cooling rate. IRVINE and PICKERING (58) found that auto-tempering resulted in the formation of cementite as a Widmanstätten precipitate within each martensite lath. The cementite plates were found to have an axial: width ratio of approximately 20:1 and this was in agreement with results obtained by Aborn.

It appears that a high stacking fault energy of the austenite ($\approx 80 \text{ erg. cm}^{-2}$) together with a high transformation temperature favours lath martensite whereas intermediate stacking fault energies ($80\text{--}40 \text{ erg. cm}^{-2}$) and lower transformation temperatures favour twinned martensite. In highly alloyed materials with very low stacking fault energies ($< 20 \text{ erg. cm}^{-2}$) and low transformation temperatures, lath martensite again becomes predominant (59).

2.2.2.2. Austenite-bainite reactions

Bainite is currently defined (60,61) as a constituent in steels which is formed by austenite decomposition within a temperature range located between the region of martensite transformation and that of pro-eutectoid ferrite and pearlite formation. This constituent is said to consist of an aggregate of acicular ferrite and carbide particles. As the transformation temperature decreases, within the bainite region, structural morphology changes also to give increased ferrite acicularity and a decrease in carbide particle size. The bainitic transformation region is often divided into two regions which yield, at higher and lower transformation temperatures respectively, the structural modifications termed upper and lower bainite (36). It has recently been shown by HABRAKEN and ECONOMOPOULOS (37) that, particularly during the continuous cooling transformation of low carbon, low alloy steels, a third type termed intermediate or granular bainite may form. The temperatures at which the bainite transformation starts and finishes, B_s and B_f respectively, have been related to steel composition by several workers. The formula due to STEVEN and HAYNES (38) for the B_s temperature has met with some degree of success:

$$B_s, ^\circ\text{C} = 830 - 270C - 90Mn - 37Ni - 70Cr - 83Mo \quad 2.2.2.$$

As noted in section 2.2.2.1., other empirical formulae are to be found in the literature (39).

2.2.2.2.1. Upper bainite

IRVINE and PICKERING (58, 62), as a result of studies of low carbon bainitic steels, suggest the following mechanism for the formation of upper bainitic structures from undercooled or metastable austenite. At high temperatures in the upper bainite range, the first stage in transformation is the nucleation of ferrite laths from austenite by a shear transformation having the Kurdjumov-Sachs (55) or Nishiyama-Wasserman (63, 64) orientation relationship between the bainitic ferrite and the austenite. At the temperatures involved carbon has a sufficiently high diffusion rate in austenite to diffuse away, ahead of the moving bainitic ferrite - austenite interface. Side-by-side, sympathetic or cooperative nucleation (65) of many bainitic laths is caused by the strains accompanying the shear transformation and as these laths grow together the solubility of carbon in the austenite trapped between them is exceeded. As a result, cementite is precipitated from the austenite with the Pitsch (66) relationship and is related in orientation to the bainitic ferrite through the Kurdjumov-Sachs (55) or Nishiyama-Wasserman (63, 64) relationships. Both the carbon content of the steel and the transformation temperature determine the extent to which carbon enrichment occurs before cementite forms directly from the austenite. The

higher the carbon content of the steel or the lower the transformation temperature the lower will be the degree of carbon enrichment necessary for this to occur and the larger will be the regions of austenite enriched carbon. This accounts for regions of higher carbon upper bainite found in low carbon, low alloy steels, where virtually continuous carbide films between the bainitic ferrite grains often completely obliterate any evidence of the boundary between these. In some instances carbon enrichment can decrease the M_s temperature to such an extent that some untransformed austenite can be retained at room temperature in apparently fully transformed upper bainitic structures. This type of behaviour has been established by WHEATLEY and BAKER (67).

As the transformation temperature decreases the rate at which carbon can diffuse in front of the ferrite-austenite interface decreases so that the sideways growth rate of the ferrite laths is also reduced, resulting in narrower laths. It is possible that the increased strain associated with the reaction tends to restrict edgewise growth of the ferrite laths in the initial stages of the shear reaction, although the axial length of the laths is little affected. Both effects lead to a finer bainitic ferrite grain size. Lath length appears to be a function of the austenite grain size or the space between prior formed bainitic volumes in which bainitic ferrite laths are nucleating and growing. The slower rate of diffusion of carbon away from the ferrite-austenite interface also infers that the ferrite tends to become more supersaturated with carbon. VASUDEVAN et al., (68) have shown that bainitic ferrite becomes more supersaturated in carbon as the transformation temperature decreases. PICKERING (69) has shown, by internal friction experiments, that a carbon-dislocation interaction does occur in bainitic structures which increases in intensity with lowering transformation temperature. The degree of interaction is very much less than in martensitic structures (70) and this is attributed to tempering of the bainite during the reaction period.

2.2.2.2.2. Lower bainite

A temperature is eventually reached at which the carbon diffuses so slowly that it can no longer diffuse away sufficiently in front of the bainitic ferrite-austenite interface. When this occurs upper bainite can no longer form and is replaced by lower bainite. According to IRVINE and PICKERING (62) the sequence of events on transformation to lower bainite is independent of carbon content. The initial bainitic ferrite laths are formed by shear, with the normal orientation relationship with the austenite, but, due to the relatively slow rate of carbon diffusion at the temperatures involved, such laths are very thin. The bainitic ferrite laths are supersaturated with carbon (68) and before lateral growth can occur, such supersaturation must be decreased by carbide precipitation. Even after such precipitation, the

ferrite still appears to retain a degree of carbon supersaturation. Certain workers (71, 72) have found that the initial carbide precipitate is ϵ -iron carbide but that cementite is generally substituted for this at a later stage. WILSON (73) has shown that the orientation relationship between the carbides and ferrite is very similar to that found in tempered, internally twinned martensite. Once precipitation has lowered the carbon content of the bainitic ferrite sufficiently, this phase will continue to grow laterally. The new growth again produces supersaturated ferrite and further carbide precipitation is required within the ferrite for growth to proceed. Many bainitic ferrite laths are randomly and not co-operatively nucleated within an austenite grain and the process of carbide precipitation and lath growth is continuous until a growing plate impinges upon an adjacent plate.

Decrease in transformation temperature within the lower bainite range causes the bainitic ferrite laths to become thinner and an increase in carbon content merely serves to increase the relative amount of carbides present. The carbides tend to precipitate, in the form of small rods or plates, at a relatively constant angle at about 55° to the axial direction of the lath (72, 74, 75). It has been suggested by SHACKLETON and KELLY (36) that the initial bainitic ferrite plates are formed martensitically and are of high enough carbon content to be internally twinned. Carbides are said to be nucleated at, and grow along, twin boundaries to relieve carbon supersaturation. This mechanism gives the observed morphology but is open to conjecture since, to date, no twins have been observed in bainitic ferrite. It has been proposed (36), however, that carbide precipitation on the twins would cause twin removal.

The dislocation density and degree of carbon-dislocation interaction in the bainitic ferrite are greater in lower than in upper bainite and, as in the case of upper bainite, increase with decreasing transformation temperature in the lower bainite range (69).

2.2.2.2.3. Intermediate Structures

Certain workers (76, 77) have shown the occurrence of so called intermediate structures in low alloy steels containing less than 0.3% carbon, which did not correspond to the accepted definitions of upper and lower bainite. These structures consist of bainitic ferrite, having a high dislocation density, and carbon rich zones containing austenitic-martensitic nodules. It was found that such structures, referred to as granular or massive bainites, were more easily reproduced by relatively slow continuous cooling than by isothermal treatment and were influenced by composition, cooling rate and thermomechanical treatment.

It has been shown (78) that quenched solid solutions such as austenite in the metastable region of a continuous cooling transformation (CCT) diagram may actually dehomogenise since the solid solution is within a temperature range at which it should normally transform. Despite the fact that the temperature is still relatively high, precipitation or concentration of carbon atoms on dislocation tangles in the austenite can occur. HABRAKEN (79) and ENTIN (80) both suggest that in low alloy steels, at relatively slow cooling rates, dehomogenisation of the austenite occurs in this manner above B_s , leading to the formation of regions enriched in carbon and probably substitutional elements. It is doubtful, however, if complete homogenisation of the austenite occurs even under equilibrium conditions.

HABRAKEN and ECONOMOPOULOS (37) claim that transformation to intermediate structures occurs initially by carbon depleted substitutional ferrite being nucleated by a process similar to that found by MASSALSKI (81) and OWEN et al., (82) in the formation of massive martensite in carbon free ferrous alloys. The ferrite interfaces then move along suitably oriented planes and are halted in their growth when the carbon enriched regions are encountered. Subsequent transformation of the enriched regions occurs to produce completely martensitic or austenitic-martensitic nodules.

It is found that at low cooling rates the carbon enriched clusters are relatively large and intermediate structures are formed. With increasing cooling rate the clusters decrease in size and transformation to conventional upper and lower bainite can occur.

2.2.2.3. The Effect of Plastic Deformation on Austenite Transformation

In a welded joint local plastic deformation occurs as a result of local expansion and contraction during the thermal cycle and this may affect the transformation of austenite during cooling.

Several workers (83-87) have studied the effect of an externally applied tensile stress and the consequent plastic deformation on the isothermal transformation of austenite. It was found that transformation start and finish times were decreased and that the transformation rate was increased. MANKARA (88) claims that stresses may distort the austenite formed in the heat affected zone and delay its transformation on cooling so encouraging some austenite to be retained at room temperature.

SMALLMAN (89) suggests that even elastic stresses, when applied above the M_s temperature and maintained during cooling, can effect the transformation. The M_s temperature is lowered by

hydrostatic stresses and raised by uniaxial compressive or tensile stresses.

It is therefore possible that the dynamic thermal stresses produced in the heat affected zone during welding could have a pronounced effect on the resultant microstructure.

2.2.3. Decomposition of Austenite Transformation Products

In earlier studies the sequence of changes which occurs on tempering was often classified into numerical stages. More detailed work has indicated that such a classification is not completely satisfactory and it is no longer generally adhered to.

2.2.3.1. Martensite

As the tempering temperature is increased in the case of lath martensites, precipitation of ϵ - carbide occurs in a Widmanstätten array within the laths followed by cementite nucleation and growth at the lath boundaries. The carbon content of the matrix is reduced, during this period, to the low level of saturated ferrite. Between about 400° and 500°C the original martensite laths which had an inherently high dislocation density due to transformation, recover and then recrystallize. Cementite is re-nucleated at the equi-axed ferrite grain boundaries and HYAM and NUTTING (90) have shown that further reaction involves Ostwald ripening of the carbides with simultaneous ferrite grain growth. ABORN (57) has shown that the subsequent tempering of auto-tempered lath martensite involves, initially, the growth of carbides formed during cooling from the M_s temperature followed by further cementite precipitation. These are the only modifications to the outlined scheme.

According to TEKIN and KELLY (91) the first change in structure on tempering twinned martensite is the formation of ϵ - carbide in a ladder-like array between the twins. Further precipitation of ϵ - carbide occurs along twin boundaries but eventually separate nucleation of cementite occurs at these sites. At this stage of tempering the structure is indistinguishable from that of lower bainite with the exception that there are no visible twins in the bainite. With increasing temperature the twins are annealed out of the martensite and carbide growth continues. However, carbide nucleation now occurs at the plate boundaries and the prior austenite grain boundaries. The next sequence of changes is identical to that found in lath martensites with recrystallisation, grain and carbide particle growth becoming the dominant phenomena until all traces of the original martensitic structure are removed.

Any retained austenite in an otherwise basically lath or twinned martensite structure generally transforms, isothermally, at about 230° to 280°C to form lower bainite.

IRVINE (92) has shown that the non carbide forming elements nickel, silicon and manganese tend to slow down the process rate at the various stages so that in order to reach equivalent structures in low alloy steels higher temperatures have to be used than with plain carbon steels. Strong carbide forming elements such as chromium, molybdenum, vanadium, tungsten and niobium tend to stabilize ϵ - carbide and retard recrystallisation. Above about 500°C cementite is replaced by the alloy carbide, the exact mechanism depending upon the element present, Matrix recrystallisation occurs, as already noted, but ferrite grain growth is controlled by the slowly growing alloy carbides. It is possible for separate nucleation of an alloy carbide to occur resulting in the phenomenon of secondary hardening (93). Trace elements can markedly influence secondary hardening behaviour particularly in the early stage of alloy carbide formation (94).

2.2.3.2. Bainites

Tempering generally has little or no effect on upper bainitic structures unless temperatures in excess of 500°C are used since the low energy bainitic ferrite boundaries are very slow to migrate. At such temperatures the cementite originally present at the bainitic ferrite lath boundaries tends to spheroidize. Only after prolonged treatment at 600°C do tempered upper bainite and tempered lath martensite become structurally indistinguishable.

Lower bainites, as indicated in section 2.2.3.1., are virtually indistinguishable from tempered martensite at the higher tempering temperatures. The beneficial structure, as regards mechanical properties, of numerous dispersed carbides in a ferrite matrix is retained.

Granular or massive bainite has, inherently, neither a fine bainitic ferrite grain size nor a uniform carbide dispersion. However, certain subcritical thermomechanical treatments can lead to recrystallisation within the ferrite, with the formation of a large number of sub-boundaries, and to the formation of a carbide dispersion in a ferrite matrix in prior austenitic-martensitic regions.

2.2.4. Strengthening Mechanisms

The degree of strengthening produced in low alloy steels by thermal treatment is a direct function of the structure obtained by such treatment.

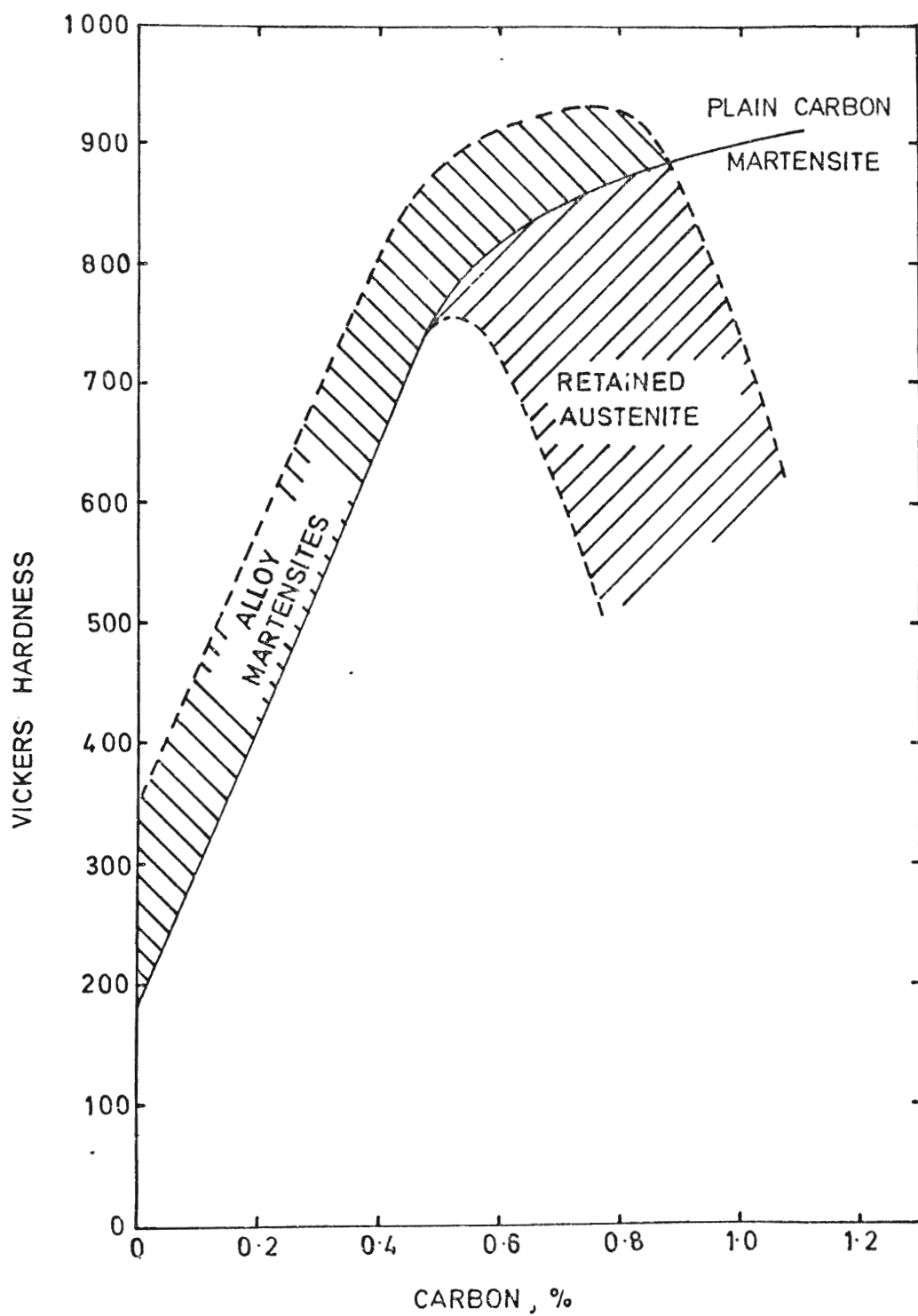


FIGURE 2. EFFECT OF ALLOYING ELEMENTS ON THE HARDNESS OF MARTENSITE (IRVINE et al. 1968)

2.2.4.1. Martensite

The hardness and the strength of martensite, as shown in figure 2, rise rapidly with increase in carbon content. KELLY and NUTTING (46) suggested that the internal twinning of high carbon martensite was a source of such strength but this has been discounted by other workers (95, 96). Grain size has some effect in strengthening (97) but it is now generally accepted that there are three major contributions to strength.

Theories on precipitation strengthening arise from the fact that martensite produced by a very rapid quench down to a low temperature is much softer than that formed at slower cooling rates when a certain amount of interstitial diffusion can occur both during and after transformation. This effect occurs in the twinned rather than in the lath form of martensite. It has been calculated (98) that approximately half the strengthening effect in a water quenched eutectoid carbon steel is due to some form of carbon segregation or precipitation at either twin interfaces or dislocations.

The relationship (99, 100) between the shear stress, τ , of an alloy and the solute concentration, C , is given by

$$\tau = \tau_0 + \frac{G C^{\frac{1}{2}}}{k} \quad 2.2.3.$$

where τ_0 is the shear stress at zero solute content, G the shear modulus and k a constant. Values for k of approximately 6.5 and 7.5 have been obtained for twinned and lath martensite respectively. CHILTON and KELLY (100) have shown that τ_0 is virtually independent of structural morphology. Substitutional solutes tend to increase the magnitude of k in equation 2.2.3. but, to date, no actual values have been determined.

The substructure within the martensitic plates, such as dislocations or internal twins, is extremely important as regards strengthening.

2.2.4.2. Bainites

The tensile strengths of bainites are closely related to their temperatures of formation, as shown in figure 3. The strength of bainite depends upon a number of different effects which have been reviewed by PICKERING (69). As the transformation temperature is reduced:

- (a) the bainitic ferrite grain size is decreased which gives strengthening according to the PETCH (101) relationship,

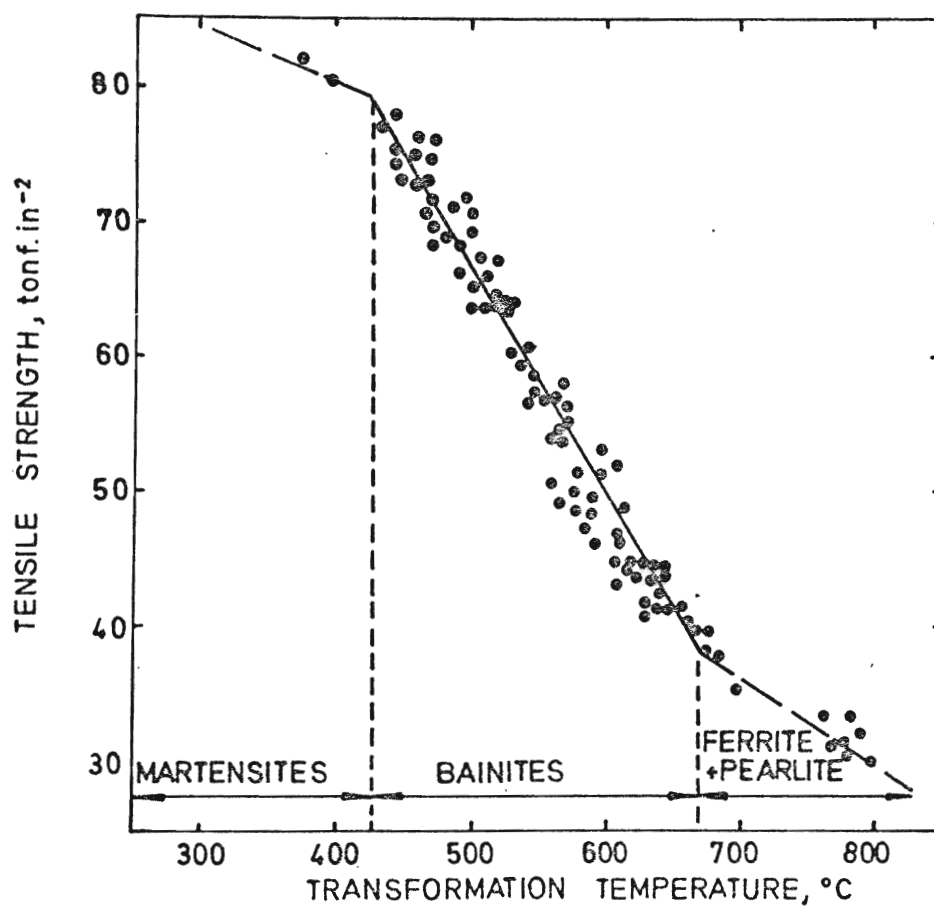


FIGURE 3 THE EFFECT OF TRANSFORMATION TEMPERATURE ON THE TENSILE STRENGTH OF BAINITE (IRVINE AND PICKERING 62)

$$\sigma_y = \sigma_i + k_y d^{-1/2} \quad 2.2.4.$$

where σ_y is the lower yield stress, σ_i the frictional stress for pure iron, k_y a constant and d the grain diameter. The low angle boundaries which form the majority of the bainitic ferrite boundaries are apparently capable of exerting a strengthening effect. In upper bainite these boundaries are, of course, the sites of the carbide particles.

- (b) for a given carbon content there is a reduction in carbide particle size and an increase in the number of such particles. This change causes an increase in the dispersion strengthening brought about by the carbide. According to OROWAN (102) the yield stress of a carbide dispersion in a ferrite matrix is determined by the shear stress, τ , required to force a dislocation line between two carbide particles separated by a distance L . If G is the shear modulus and b the Burgers vector of the dislocation line,

$$\tau = \frac{Gb}{L} \quad 2.2.5.$$

A similar relationship has been developed by FISHER et al., (103). The amount of carbon held in interstitial solution also increases as the transformation temperature decreases and this effect will give strengthening according to equation 2.2.3.

- (c) the greater the expansion during bainite formation the higher is the level of internal stress within the structure. The lower the temperature of transformation the less will be the thermal relief of such stresses. The presence of internal stress is thought to be of secondary importance compared with other strengthening mechanisms. However, other strengthening mechanisms may partly exert their influence through the medium of internal stress.
- (d) the dislocation density increases presumably due to the greater strains accompanying transformation and because there is less chance for the dislocations to be annealed out. Thus strength is increased on lowering transformation temperature due to an increase in dislocation - dislocation and dislocation-interstitial interaction. Lower bainite,

in which the carbides are formed by direct precipitation from the ferrite, contains a higher dislocation density than upper bainite. This is thought to be due to the dislocation generating effect of the stresses produced by carbide precipitation and the bainitic ferrite-carbide interface acting as a source for such dislocations. This effect is assisted by strains due to differential contraction of the bainitic ferrite and carbide during cooling from the transformation temperature and by the strains imposed on existing bainitic regions during the course of transformation.

It may therefore be concluded that the properties of lower bainites produced at low temperatures are controlled chiefly by the carbide dispersions and the carbon still held in solid solution, while in upper bainites the predominant strengthening mechanism is likely to be ferrite grain size hardening.

2.2.4.3. Tempered Structures

When tempered, low alloy steels, initially cooled to produce a martensitic structure, generally exhibit a decrease in tensile strength and an increase in tensile reduction of area and elongation as the tempering temperature is increased. Commercial tempering usually takes place in the region of 200°C or at 500°C and above to avoid an embrittlement trough which can occur in the intermediate range and which is often referred to as 350°C embrittlement. IRANI et al., (104) suggest that a combination of two mechanisms involving carbide precipitation and high dislocation density, resulting from dislocation pinning, is responsible for such embrittlement. McMAHON (105) has reviewed this form of brittleness and also temper embrittlement which can occur in some alloy steels on heating in or cooling slowly through the range 550° to 350°C. Martensite tempered in the range 550° to 700°C shows a uniform carbide dispersion in a ferritic matrix. HYAM and NUTTING (90) have shown that the major contribution to the strength of tempered martensite originates from the extremely small ferritic grain size, inherited from the fineness of the martensitic needles, and from the uniform dispersion of small carbide particles. The effects of ferritic grain size and degree of carbide dispersion as strengthening mechanisms have been discussed in Sections 2.2.4.2.(a) and (b) in regard to bainitic structures but are equally applicable to tempered martensite.

In upper bainite tempering produces fairly rapid bainitic ferrite grain growth and this, combined with a none too regular carbide dispersion, leads to inferior strength. This structure is also detrimental to the impact properties so that upper bainitic structures are seldom, if ever, intentionally tempered.

The uniform carbide dispersion found in lower bainitic structures associated with relatively high strength levels, permits tempering to be used to lower the strength whilst retaining many dispersed carbides. The lowering of strength also results in improved impact properties. Tempered granular bainite has inferior mechanical strength after tempering as a result of bainitic-ferrite grain growth and the widely dispersed carbide colonies resulting from the decomposition of austenitic-martensitic regions.

2.2.5. Structural Heterogeneity

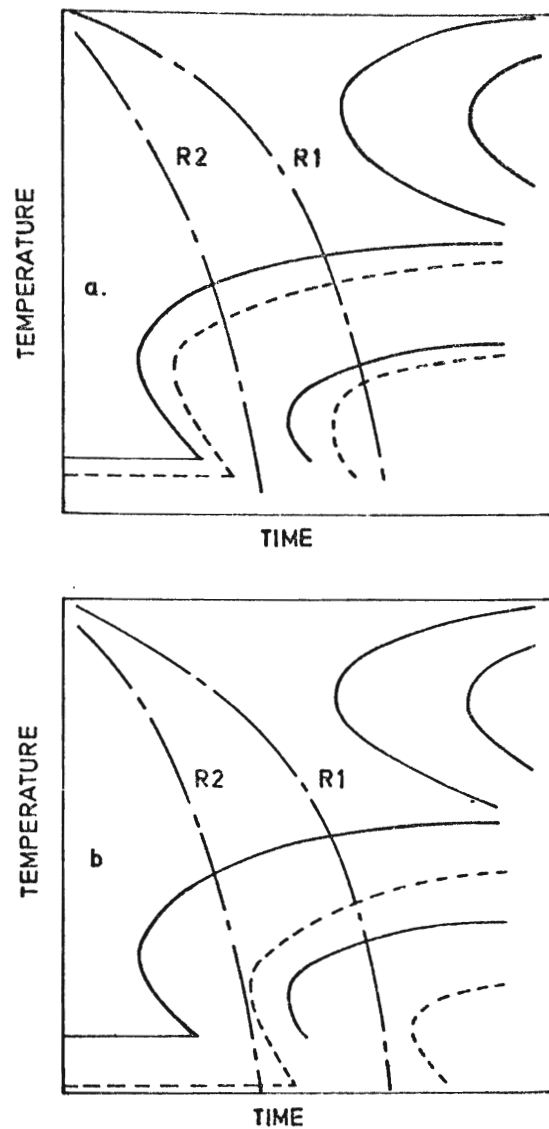
In any low alloy steel transformation of essentially homogeneous austenite can lead to mixed structures whose form depends upon cooling rate from the austenite field and steel composition. Such structures cannot really be termed heterogeneous since the prior austenite from which they were formed was basically of a uniform nature as regards composition. Both macroscopic and microscopic heterogeneity can be said to occur in low alloy steels. Macroscopic heterogeneity owes its existence to the interdendritic segregation of impurities and alloying elements during ingot solidification which, on cooling to room temperature after hot working, produces a banded or layered structure in the final product. Microscopic heterogeneity is due to the variation of alloying element content, particularly carbon, within individual austenite grains produced by rapid thermal cycling within the austenite region, such as is found in the heat affected zone during welding. It is obvious that microscopic heterogeneity may be present in steels which exhibit banding so that structural interpretation can, at times, be quite complex.

2.2.5.1. Banding

The term banding is used to indicate periodic or intermittent variations of alloy composition which occur in laminated or fibrous form. Unidirectional working such as rolling tends to produce a fibred structure whereas operations such as cross rolling produce a layered structure (106). Typical centre to centre band or layer separations are 0.1 to 0.25 mm.

KIRCALDY (107) has suggested that segregation of alloying elements and impurities in a rolled product changes the thermodynamic activity of carbon in the austenite, during holding within the austenitic range such that the carbon activity as opposed to the carbon concentration, is equalized. If, as in the case of phosphorus, the segregating element raises the activity then at soaking temperatures the phosphorus rich regions will have a lower carbon content than the rest of the material. Where the alloying or impurity element lowers the carbon activity the converse is true. This situation can lead to both pre-

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— BANDS DEFICIENT IN ALLOYING ELEMENT
 - - - BANDS RICH IN ALLOYING ELEMENTS
 — COOLING CURVES.

- a COMPOSITE TTT CURVES FOR SITUATION WHERE INTENSITY OF MICROSEGREGATION IS LOW.
 b COMPOSITE TTT CURVES FOR SITUATION WHERE INTENSITY OF MICROSEGREGATION IS HIGH

FIGURE 4. EFFECT OF MICRO SEGROGATION ON THE TRANSFORMATION CHARACTERISTICS OF LOW ALLOY STEEL.

transformation and transformation segregation although, as the cooling rate from the austenite region is increased, the latter tends to be decreased in intensity.

BASTIEN (108) claims that banding may be primarily attributed to the influence of segregates on the A_{C3} temperature but that the cooling rate from the austenite field is also important. He suggests that in the case of phosphorus the normal carbon displacements away from carbon rich regions, may be inverted due to its influence in raising the activity of carbon in austenite. This is a function of the fact that phosphorus increases the hardenability of the steel and the time-temperature-transformation (TTT) curves for high and low phosphorus content can intersect in the case of plain carbon steels.

SMITH et al., (109) have shown that in addition to phosphorus segregation, chromium, nickel, molybdenum and manganese also segregate in low alloy steels. All of these elements tend to retard the bainite transformation and molybdenum and chromium lower the temperature range over which it takes place.

Composite TTT curves for a low alloy steel may be drawn as shown in figure 4. The actual separation of the curves for alloy rich and alloy deficient regions depends upon the degree of microsegregation, being wider the more intense is the segregation. If the situation in figure 4a, corresponds to slight segregation then severe segregation could be represented as in figure 4b. By means of the superposition of identical cooling rates on each diagram it is possible to envisage how the severity of the appearance of banding can be directly related to the severity of microsegregation. In the case of cooling rate R1 in figure 4a, both the alloy rich and alloy impoverished regions will transform during cooling to upper bainitic structures. With severe segregation, figure 4b, during cooling from above A_3 , the alloy deficient regions start to transform first to upper bainite. As cooling proceeds, these bands complete their transformation in the upper bainite range but the more highly alloyed regions finish transformation in the lower bainite range. This difference in microstructure accentuates the banded appearance. If a faster cooling rate, R2, is considered then for slight segregation, figure 4a, the bands will contain predominantly upper bainite and lower bainite for the alloy lean and alloy rich bands respectively. In the case of severe microsegregation, figure 4b, it is possible that cooling rate R2 would cause the alloy rich bands to transform directly to martensite while the lean bands transform to a mixed, but predominantly lower, bainitic structure.

Periodic banding can only be eliminated, so that it will not recur on slow cooling, by soaking for sufficient time to homogenise all segregates present. However, as such a treatment leads to an increase in austenite grain size, with detrimental

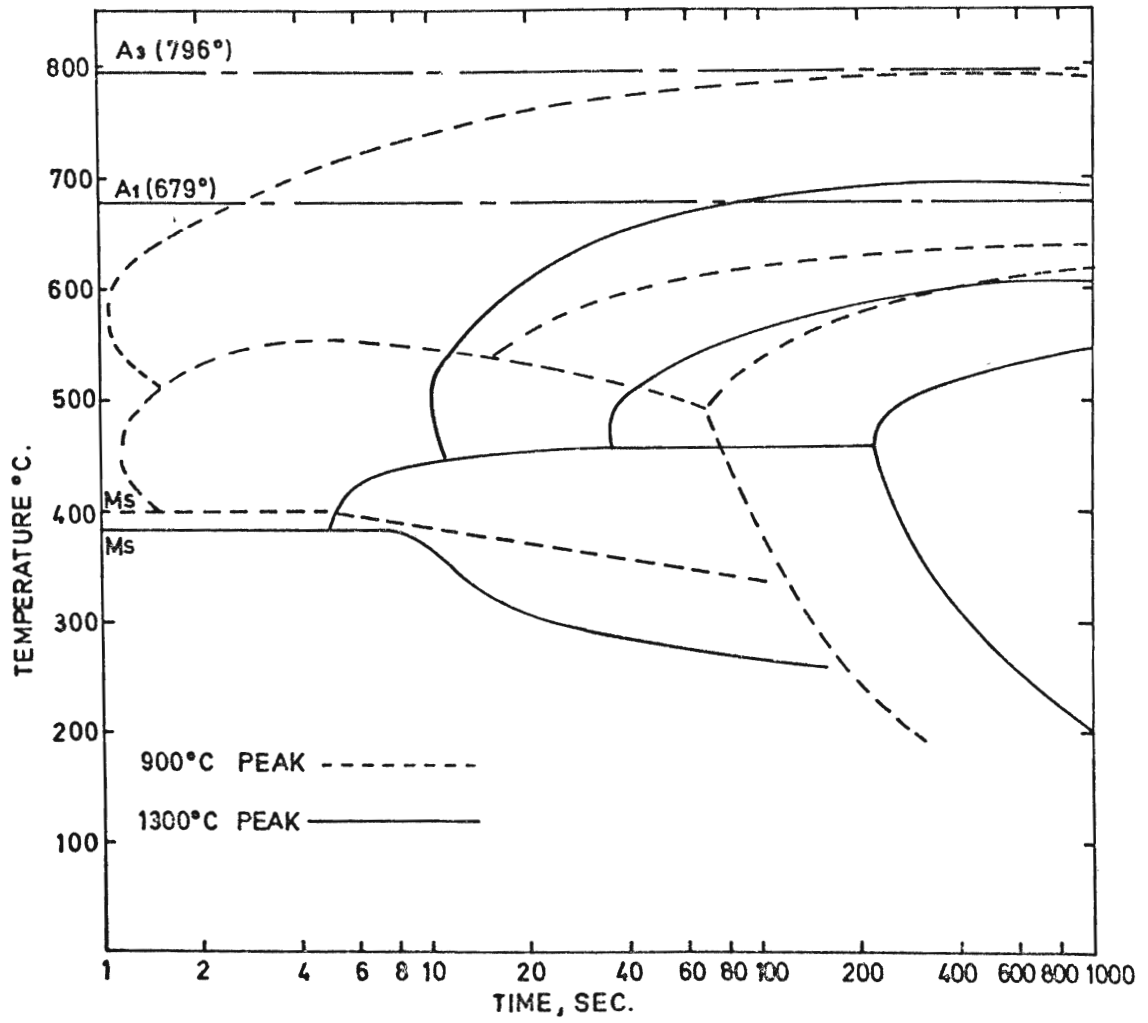


FIGURE 5: EFFECT OF RAPID HEATING MAXIMUM TEMPERATURE ON A CONTINUOUS COOLING TRANSFORMATION DIAGRAM (INAGAKI AND SEKIGUCHI¹¹³).

effect on mechanical properties, it is not often used in an effort to eliminate banding.

2.5.2. Austenite Inhomogeneity

As previously shown in section 2.2.3., intermediate bainitic structures can be produced in low alloy steels as a result of inhomogeneity developing in austenite during cooling to the transformation temperature (78). However, austenite heterogeneity is more commonly produced by rapid thermal cycling in the austenite field leading to partition of carbon by virtue of somewhat anomalous carbide solution.

In a simulation study of weld heat affected zone structures in a low carbon, low alloy steel SAVAGE and OWCZARSKI (110) found that heterogeneous austenite, with respect to carbon content, could be produced during the weld thermal cycle from a structure consisting of temper carbides in a ferrite matrix. It was suggested, as indicated in section 2.2.1., that, in the presence of difficulty soluble carbides, it is possible for partition of carbon and other alloying elements to occur in the austenite leading to a structure of predominantly low carbon martensite but containing regions of high carbon martensite if the cooling rate is sufficiently fast and all carbides are completely dissolved. BONISZEWSKI et al., (111) and DOLBY (112) have found such structures in the actual weld heat affected zones of similar steels. The existence of ferrite and regions of high carbon martensite with some partly dissolved temper carbides was found (110) for varying degrees of austenitisation in the temperature region between Ac_1 and Ac_3 for a range of thermal cycles corresponding to specific weld energy inputs.

INAGAKI and SEKIGUCHI (113) have shown, figure 5, for a low carbon (0.17%) - manganese steel that as the peak temperature of a thermal cycle is increased the temperature of transformation of austenite on continuous cooling is decreased and shifted to longer times. This seems to indicate a higher alloy carbide solubility and the production of a more homogeneous austenite as the peak temperature rises. If the peak temperature of a thermal cycle is sufficiently high then complete austenitization and homogenisation of the austenite can take place before transformation occurs on cooling. D'ANDREA and ADAMS (114) claim that for a given cooling rate, austenite homogeneity and the subsequent structure produced on transformation are not governed so much by the peak temperature attained but by the overall time spent above Ac_3 . However, although these workers found apparent structural similarity for a particular cooling rate, associated with different peak temperatures and time spent in the austenitic condition, they did not take into account the inherent variation in austenite grain size which must accompany such a combination of conditions. Prior austenite grain size does depend essentially

upon the peak temperature attained and is of fundamental importance in determining the brittle fracture characteristics of a low alloy steel.

2.3. BRITTLE FRACTURE

The fracture process in a material is initiated by the appearance of a crack and does not finish until the crack propagates through the material. All types of fracture have a common characteristic viz., the nucleation of fracture is a consequence of localised plastic flow. Crack nucleation requires that plastic flow can occur in local regions while the remainder of the material is either elastic or contains varying amounts of plastic strain. High stress concentrations are produced in these regions and when the concentrated stress becomes equal to the cohesive strength of the material cracking can occur. The propagation mode of a crack determines the nature of the final fracture. In ductile fracture continued plastic deformation is required for propagation. In brittle fracture plastic deformation is not a necessary adjunct but may occur during the spread of a crack.

Many metals with the exception of the f.c.c. type, although some anomalous behaviour has been found in this type (115, 116, 117), show a transition from ductile to brittle behaviour when the temperature is lowered. The temperature range in which the transition occurs is not fundamental since other properties do not exhibit discontinuous change in this range and alteration of strain rate, notching, composition and microstructural change can shift the transition. Brittle fractures are generally trans-crystalline in nature and travel along so called cleavage planes which calculations (118) show to be the planes, in a given structure, across which the atomic bonding is weakest. Brittle failure may also occur by means of intergranular fracture either due to the actual separation of a brittle phase at grain boundaries (119) or by grain boundary embrittlement by impurity segregation without the appearance of a separate phase (115, 120).

2.3.1. Crack Nucleation

Certain experimental observations suggest that plastic flow precedes crack nucleation in brittle fracture. It has been established that at low temperatures, the brittle fracture stress of mild steel, for a wide range of grain sizes, is equal to the yield stress. The close correlation obtained between fracture and yielding is indicative of the fact that a small amount of plastic flow is sufficient to cause failure at low average stress levels. GILMAN (121) during experiments on zinc single crystals found that at -196°C fracture normally occurred at low

stresses but that if plastic flow in them was suppressed, very high stresses were required for fracture.

In a modification of the Griffith theory, section 2.3.2., ZENER (122) proposed that cracks are not pre-existent but form during plastic deformation. He suggested that a crack could be nucleated in a slip band by the coalescence of glide edge dislocations at some obstacle in the slip band as a precipitate particle or a grain boundary. Two requirements must be met before such a crack can form viz., plastic deformation must occur and the obstacle blocking dislocation movement must be an efficient barrier.

Following an analysis of the stress at a dislocation pile-up, STROH (123) modified Zener's theory by showing that fracture could only be initiated by the pile up of glide dislocations if such pile-ups were not relieved by the operation of dislocation sources near the head of the pile-up. The assumption was made that the dislocation sources in the grain adjacent to the pile-up were subject to Cottrell locking (124) and were so prevented from relieving the stress concentration by slip. The inherent implication of this mechanism is that it will only operate during yielding when the density of operative sources is low and the few sources which do operate can be expected to form large pile-ups in a short time. A cleavage crack was considered to form when n dislocations, under the action of a resolved shear stress τ , satisfied the relation:

$$n b \tau = 12 \gamma \quad 2.3.1.$$

where γ is the surface energy and b the Burgers' vector. If the length of slip plane, L , occupied by the pile-up is given by:

$$L = \frac{n b G}{\pi (1 - \nu) \tau} \quad 2.3.2.$$

where ν is Poisson's ratio, then it may be shown that the fracture stress σ_f , is related to the grain diameter, d , by:

$$\sigma_f = k d^{-1/2} \quad 2.3.3.$$

where

$$k = 4 \left[\frac{6 G \gamma}{\pi (1 - \nu)} \right]^{1/2} \quad 2.3.4.$$

PETCH (125) showed that, on the basis of experimental data, better agreement was obtained with the relation:

$$\sigma_f = \sigma_i + k d^{-1/2} \quad 2.3.5.$$

where σ_i is the friction stress or stress resisting the movement of free dislocations along slip planes and k is a constant. In earlier work PETCH (101) and HALL (126) purposed a similar equation, previously quoted in section 2.2.4.2(a), which relates the lower yield stress, σ_y , and the grain diameter, d :

$$\sigma_y = \sigma_i + k_y d^{-1/2} \quad 2.3.6.$$

σ_i has the same significance as in equation 2.3.5. and k_y is a constant.

According to equation 2.3.6., if crack nucleation occurs by a dislocation pile-up model then any factors which increase σ_y will, tend to prevent the onset of yielding in the matrix and permit larger stress concentrations to be produced. The larger the pile-up and the larger the associated stress concentration at the tip of the pile-up, the easier it is for fracture initiation to occur.

HESLOP and PETCH (127) have shown that the friction stress term may be resolved into two terms,

$$\sigma_i = \sigma_i' + \sigma_i'' \quad 2.3.7.$$

σ_i' is the stress required to drive a free dislocation through the matrix against the resistance of dispersed impurity atoms, precipitate particles, other dislocations and developed sub-structures and is independent of temperature. CRACKNELL and PETCH (128) have found linear variation between σ_i' and the concentration of carbon and nitrogen in solution in ferrite. A typical value is 2 tonf. in⁻². for annealed mild steel. Irradiation, alloying, strain ageing, strain hardening, quenching and quench ageing are all treatments which produce additional resistance to dislocation movement through the iron lattice and cause an increase in σ_i' .

σ_i'' is extremely temperature sensitive and is considered to be due to a strong Peierls-Nabarro force. Its room temperature value in iron is, approximately 2 tonf. in⁻², but exhibits considerable increase both at low temperatures and high strain

rates.

k_y is a measure of the localised stress required to initiate yielding in an unyielded grain. Employing a model due to ESHELBY et al., (129) in which r is the average distance between the end of a slip plane and the nearest dislocation source in the next grain and that τ_d is the stress required to operate this source, it is possible to rewrite equation 2.3.6. to obtain:

$$\sigma_y = \sigma_i + 2r^{1/2} \tau_d d^{-1/2} \quad 2.3.8.$$

which gives,

$$k_y = 2r^{1/2} \tau_d \quad 2.3.9.$$

According to COTTRELL (130) two mechanisms for yielding are possible.

- (a) interstitial elements in solid solution 'pin' or 'lock' dislocations as the result of a strong interaction force (carbon and nitrogen in the case of ferrite). When the stress exceeds a critical value, un-pinning occurs and an avalanche of dislocations is produced in the slip plane which move with a high velocity and then pile-up against other dislocations, grain boundaries or twins.
- (b) if a small sharp stress concentration due to pile-up is present at the grain boundary, the concentrated stress field may be too localised to extend to the nearest dislocation in the adjacent grain. Creation of a new dislocation may occur in the highly stressed region before un-pinning occurs outside it.

For creation, the requirements are a large τ_d and small r with the converse being true for unpinning. The preferred process is that with the smaller value of k_y . If pinning is strong, as in annealed steels, the yield process occurs by grain boundary creation. Such boundary sources have been found by LIU (131) and HORNBOGEN (132). The operative process governs the temperature dependence of k_y since unpinning from a line of adsorbed atoms is highly sensitive to temperature, whereas grain boundary creation is not.

Under the influence of an applied stress slip initially occurs in the most suitably oriented grains. As noted previously stress concentrations due to dislocation pile-ups occur at the

grain boundaries. For a fine grain size, the grain boundary obstacle to slip is greater with the result that such steels may be expected to exhibit greater ductility than their coarse grained counterparts.

COTTRELL (133) has proposed a model for crack nucleation in b.c.c. metals which involves the coalescence between a dislocation gliding in the (101) plane with one gliding in the $(10\bar{1})$ plane to form a new pure edge dislocation which lies in the (001) plane according to the reaction;

$$\frac{a}{2} [\bar{1}\bar{1}1] + \frac{a}{2} [111] \rightarrow a [001] \quad 2.3.10.$$

where a is the lattice parameter.

The dislocation produced is sensile since (001) is not a common slip plane in the b.c.c. structure. Cottrell suggested that the dislocation produced by the reaction is a potential crack and could grow by additional dislocations gliding over $\{110\}$ planes since the $\{100\}$ planes constitute the cleavage planes.

According to GILMAN (118) there are three possible and essentially different mechanisms by means of which plastic flow can nucleate cracks:

- (a) the formation of shear discs.
- (b) the occurrence of plastic strain interfaces.
- (c) the intersection or crossing of glide bands.

A shear disc is a thin lamellar region in which slip has occurred, the shear direction being parallel to the plane of the disc and the largest dimension of the disc being large in comparison with its thickness. It may consist of a set of planes on which slip has occurred or it may be a twin. The concentrated shear stress, τ_c , at the tip of such a shear disc is given by,

$$\tau_c = \tau \left(\frac{c}{p} \right)^{1/2} \quad 2.3.11.$$

where τ is the shear stress acting in the plane of the disc and p , the tip radius. τ_c is accompanied by high tensile stresses which tend to form cracks.

The occurrence of high tensile stresses is also possible at the interface in a body where the plastic strain suddenly changes

to a much smaller value. Cracks produced by virtue of this situation have been observed at kink boundaries in zinc at low temperatures.

HULL (134) claims that twinning intersection is an important mechanism for crack nucleation in b.c.c. metals. The ease with which strain can be dispersed by slip determines the effectiveness of a twin in concentrating the strain ahead of itself. High local stresses can be obtained when a high velocity twin growing under a large overall stress meets a pre-existing twin. The probability of crack formation due to such an intersection is, however, geometrically dependent. BELL and CAHN (137) have shown that, in zinc, shear strain discontinuities can exist in the case of a twin intersection and lead to cracking. HONDA (136) and KNOTT and COTTRELL (137) have identified cracks associated with twins but it is difficult to conclude whether the cracks were produced by twins or that the twins were produced by the local stress concentrations associated with propagation of the crack tip. This mode of cracking has only been reported at low temperatures in the region below - 150°C.

The formation of cracks at glide band intersections has, as yet, only been reported in the literature for the case of ceramic materials (138).

Microcracks present in the carbide phase in a ferrite matrix are a possible source of crack initiation during plastic deformation. McMAHON (139) claims that many such cracks can be ignored since the ductility of the ferrite matrix is sufficient to accommodate such stress raising discontinuities and prevent cracks propagating into the ferrite. WARREN and BEEVERS (140) have found similar behaviour in zirconium alloys containing hydride precipitate particles where crack nucleation occurs by twin or dislocation interactions at the precipitate-matrix interface.

2.3.2. Crack Propagation

It may be shown that the ideal tensile strength, σ_t , of a material is given by:

$$\sigma_t = \left(\frac{2\gamma_s E}{b} \right)^{1/2} \approx \left(\frac{\gamma_s E}{b} \right)^{1/2} \quad 2.3.12.$$

where E is the elastic modulus for uniaxial tension, γ_s the surface energy per unit area of fracture surface formed, and b the equilibrium separation of the atoms in the material.

If a notch or slit is present in a thin plate the applied tensile stress is increased locally at the tip of the notch by a factor,

$$\left(\frac{a}{p}\right)^{\frac{1}{2}}$$

where a is the depth of the notch or half the length of the slit and p is the radius of the tip. For the case of a slit in a plate INGLIS (141) showed that the concentrated stress, σ_c , is

$$\sigma_c = 2\sigma \left(\frac{a}{p}\right)^{\frac{1}{2}} \quad 2.3.13.$$

When σ_c equals the ideal tensile strength, according to equation 2.3.12, fracture of the material occurs. The nominal breaking strength, is then given by:

$$\sigma_{nom} \simeq \left(\frac{\gamma_s E p}{4 b a}\right)^{\frac{1}{2}} \simeq \frac{E}{10} \left(\frac{p}{a}\right)^{\frac{1}{2}} \quad 2.3.14.$$

The ideal tensile strength, σ_t , for a given material is many times greater than the observed fracture strength. GRIFFITH(142) first gave an explanation of this discrepancy based upon the proposition that a brittle material contains many fine cracks. Griffith's criterion for crack propagation states how much force must be applied to a body such that a crack will propagate and derives from an energy balance. A crack will propagate when the decrease in elastic strain energy is at least equal to the energy necessary to create new crack surfaces. This criterion is dependent upon the shape of the body and the position of the crack.

For an elliptical, elastic crack, of major and minor axes $2a$ and $2b$ respectively, in a thin plate Griffith derived the following expression for the growth of a sharp crack:

$$\sigma_G = \left(\frac{2E\gamma_s}{\pi a}\right)^{\frac{1}{2}} \quad 2.3.15.$$

where σ_G is the so called Griffith fracture stress. Comparison with equation 2.3.14. shows that the surface energy sets a lower value, $p = 8b/\pi$ to the effective sharpness of the crack. Cracks with p equal to or less than this limiting value will spread at the Griffith stress, σ_G , whereas those with larger values of p require larger stresses to initiate propagation. However, once these latter cracks start to spread they tend to sharpen and the stress falls to that given by equation 2.3.15. Fracture is dependent upon both the material properties and the dimensions of a body. The important length, so far noted, is ' a ' but in practice, particularly if plastic deformation precedes cracking, the length of the plastic zone ahead of the crack tip may be the most important dimension.

In considering that the material through which a crack propagates in practice, is plastic as well as elastic in nature, the most significant effect is that the crack becomes blunted. By such means the stress concentration factor at the crack tip is reduced and a greater applied stress is required so as to make the concentrated stress greater than the ideal tensile strength of the material.

OROWAN (143) modified Griffith's criterion to include the plastic work required to extend the crack surfaces. This plastic work was considered to act as a surface energy, γ_p . Thus:

$$\sigma_c = \left[\frac{2E(\gamma_s + \gamma_p)}{\pi a} \right]^{\frac{1}{2}} = \left(\frac{2E\gamma_e}{\pi a} \right)^{\frac{1}{2}} \quad 2.3.16.$$

where γ_e is the effective surface energy.

In ductile materials $\gamma_p \gg \gamma_s$ so that equation 2.3.16. may be approximated to:

$$\sigma_c \simeq \left(\frac{2E\gamma_p}{\pi a} \right)^{\frac{1}{2}} \quad 2.3.17.$$

FRIEDEL (144) and GILMAN (118) have estimated γ_p for an ideal plastic material by taking into account plastic relaxation of the concentrated stresses at the crack tip. They obtained:

$$\gamma_p \simeq \frac{1}{6} \left(\frac{\sigma_t}{\sigma_y} \right) \gamma_s = \frac{E}{10} \cdot \frac{\gamma_s}{\sigma_y} \quad 2.3.18.$$

If the ratio σ_t/σ_y is large, as in steels, γ_p is also large and the material is strong and crack propagation difficult. The converse is true of glass. The temperature dependence of the yield stress in b.c.c. metals is, however, important and with increased σ_y the fracture toughness decreases and the probability of brittle fracture is increased.

COTTRELL (133) has linked yielding behaviour to a general dislocation model for fracture. In a modification of Cottrell's model TEGART (145) suggested that a crack opens at a tensile stress σ_f , given by equation 2.3.5., at the end of a length, l , of slip band or twin where l is approximately equal to the grain diameter, d . This gives:

$$\sigma_f \simeq \left(\frac{G\gamma_e}{a'} \right)^{\frac{1}{2}} \simeq \frac{4G\gamma_e}{a(\sigma_f - \sigma_i)l} \quad 2.3.19.$$

where a' is the depth of wedge of crack produced and a fraction of the dislocation energy, α^2 , is converted into surface energy of the crack.

The assumption is made that the same effective surface energy is relevant to both crack formation and propagation. From equation 2.3.5., brittle fracture will occur provided that the fracture stress is not less than:

$$\sigma_f = \frac{4 G \gamma_e}{\alpha k d^{1/2}} \quad 2.3.20.$$

2.3.3. Factors which favour brittle fracture

The general features of factors which favour brittle behaviour can be discussed in terms of equation 2.3.20.

A rise in fracture stress σ_f , is normally associated with an increase in the lattice friction stress, σ_i . As noted in section 2.3.2., σ_i is strongly temperature dependent in b.c.c. metals and leads to a transition temperature as regards fracture behaviour. On lowering the temperature the change to brittle fracture corresponds to an increase in σ_i such that equation 2.3.20, is satisfied. The effect of metallurgical condition on σ_i has already been described in section 2.3.2. The introduction of a notch changes the stress system in a specimen from uniaxial to triaxial tension. Plastic yielding at the notch causes an excess stress to be concentrated at the notch root which is mainly hydrostatic in nature. OROWAN (143) has shown that the ratio of tensile stress to shear stress in the plastically constrained zone is approximately three times greater than for uniform uniaxial tension. This effect raises the yield stress by a factor of 3 and as a result increases the transition temperature for a given strain rate. σ_i and consequently the ductile-brittle transition shows strain rate dependence. As a function of the small volume of material deforming at the notch root, the strain rate is increased locally above the nominal value applied to the specimen. Marked increases in transition temperature may be obtained by employing respectively unnotched and notched tensile and notched impact specimens.

Lowering temperature causes a reduction in effective surface energy γ_e , and an increase in σ_f in such a way that the condition for brittle fracture can be realised. In metals, γ_e comprises mainly plastic work e.g. γ_p for iron at -196°C is approximately ten times the true surface energy, γ_s . The number of available slip systems and the number of mobile dislocations in these systems appears to be important. f.c.c. metals

with a large number of slip systems are commonly ductile at the lowest temperatures in contrast to h.c.p. zinc in which the small number of slip systems leads to a low value of γ_p and consequent brittleness. In b.c.c. metals, even although sufficient slip systems are available, dislocation locking by interstitial impurities tends to prevent the relief of stress concentration by plastic deformation. If the energy requirement to overcome plastic deformation is too great the crack may decelerate and stop in its propagation. The implication is that a certain minimum velocity must be attained in order to permit the free movement of a crack.

According to equation 2.3.20., a decrease in grain size increases the stress required for brittle fracture. KOEHLER (146) has shown that the stress concentration, $\hat{\sigma}$, at the tip of a pile-up of n dislocations is related to the applied shear stress τ on the slip system by:

$$\hat{\sigma} = \alpha n \tau \quad 2.3.21.$$

where α is a constant. Thus in a fine grained steel not only is the fracture stress increased but the stress concentration at the head of a dislocation pile-up is decreased due to a decrease in 'n'. A larger applied stress is therefore required to increase the stress concentration to the fracture stress level than in a coarse grained steel in which a higher value of 'n' leads to a greater stress concentration. As the grain size decreases, therefore, the tendency is towards more ductile behaviour. For a given strain rate and temperature there is a critical grain size for transition from ductile to brittle behaviour (12). In addition, for a given strain rate an increase in grain size increases the transition temperature.

The effective slip band length, l , may be reduced by the introduction of precipitate particles in steels by means of quenching and tempering treatments with a consequent reduction in transition temperature. However, some treatments such as quench ageing can raise the transition temperature by virtue of the fact that incomplete barriers in slip planes tend to increase σ_i . The type of structure present in heat treated low alloy steels determines the effective length and hence the fracture behaviour.

2.3.4. Influence of microstructure

2.3.4.1. Bainites

IRVINE and PICKERING (62) have shown, figure 6, that the impact transition temperature of low carbon steels exhibiting bainitic structures varies with tensile strength. The discon-

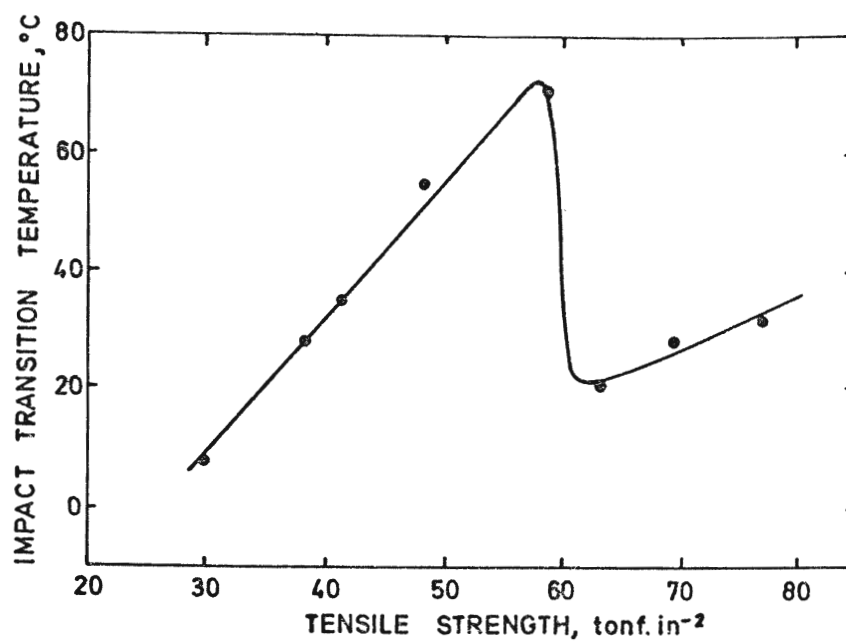


FIGURE 6. EFFECT OF TENSILE STRENGTH ON THE IMPACT TRANSITION TEMPERATURE IN BAINITIC STEELS (PICKERING ⁶⁰)

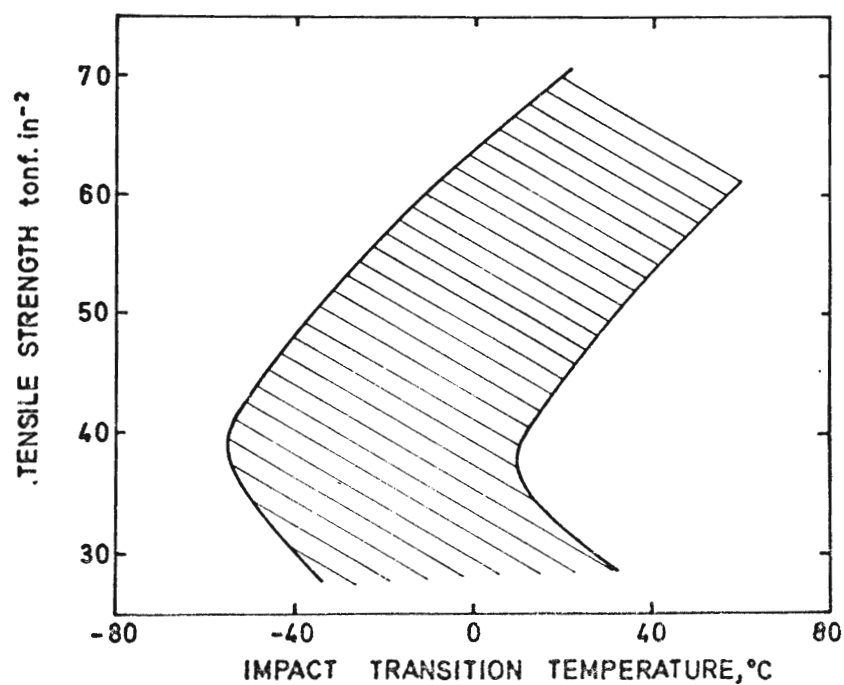


FIGURE 7. RELATIONSHIP BETWEEN TENSILE STRENGTH AND TRANSITION TEMPERATURE IN TEMPERED HIGH STRENGTH BAINITIC STEELS. (IRVINE AND PICKERING ⁶²)

tinuity in the relationship occurs at a tensile strength which is equivalent to the temperature for transformation from upper to lower bainite. It appears that despite its lower strength upper bainite generally has a higher impact transition temperature than lower bainite (62, 69). Simultaneously the transition temperature curve changes from an upright form for upper bainite to a very wide transition temperature range in lower bainite. This change is attributed to the difference in carbide dispersion in upper and lower bainite.

The carbides at the bainitic ferrite boundaries in upper bainite do not provide much obstruction to the propagation of a cleavage crack. This is particularly true in low carbon steels in which the carbides are relatively widely spaced. In addition, the low angle boundaries between neighbouring ferrite grains do not cause any major obstruction to the propagation of a cleavage crack. Only at the boundaries between different sheaves of bainitic ferrite laths, or, at prior austenite grain boundaries, is there a major change in the direction of the cleavage crack when some tearing and energy absorption can occur. Once initiated, a brittle cleavage crack will spread rapidly over a large number of bainitic ferrite grains in any prior austenite grain. Thus the transition temperature range is narrow and the impact transition temperature curve has an upright form.

The increased dislocation density and, more particularly, the increase in the number of dispersed carbides in lower bainite causes the crack to intersect and propagate through or around or actually fracture many more carbides. In this way the fracture path of a cleavage crack is interrupted so that the crack deviates from one plane to another, the different planes being joined by local plastic tearing which, as in the case of upper bainitic structures, increases the energy absorbed in fracture. Thus while the higher strength lower bainitic structures may more readily initiate a brittle crack they do not allow easy propagation of such cracks. When a crack is arrested some deformation and ductile fracture occur until a new cleavage crack is initiated since the work hardening involved raises the flow stress. The impact transition temperature range is considerably extended and while brittle cleavage fracture may initiate at higher testing temperatures than in upper bainitic structures the measured impact transition temperature appears to decrease due to the wider temperature range over which the ductile-brittle transition occurs.

Tempering of upper bainites only has an effect at very high tempering temperatures but, due to the rapid ferrite grain growth which occurs, is detrimental to the impact properties. Lower bainites may be tempered to improve the impact properties since they still retain a structure containing many dispersed carbides in a ferrite matrix. Figure 7 shows the relationship, for a range of compositions, between tensile strength and impact

transition temperature after tempering.

In granular or massive bainite, the large bainitic ferrite areas favour the formation of microcracks and the austenitic-martensitic colonies are neither sufficiently fine or dispersed to constitute efficient obstacles to cleavage crack propagation. Additionally, the presence of an appreciable amount of martensite contributes to brittleness.

2.3.4.2. Martensites

The fracture behaviour of martensitic structures can be essentially attributed to the differences in metallurgical structure and properties of martensites of different composition. However, as discussed in section 2.3.5.3., BONISZEWSKI and BAKER (147) have produced entirely auto-tempered and internally twinned martensitic structures, by virtue of cooling rate alteration, in a single low carbon, alloy steel.

In auto-tempered, low carbon lath martensite, due to the presence of the finely dispersed carbide aggregate within the laths, the fracture behaviour is very similar to that previously described for lower bainitic structures. Transformation from austenite does not lead to the production of severe short range stresses which could produce micro-cracks when long range stresses are applied. Thus plastic deformation is an essential prerequisite to crack nucleation.

In the case of internally twinned martensite a severe short range stress system can be formed during austenite transformation and on the application of long range stresses microcracks can be developed. It is possible that such cracks could be nucleated in the defective lattice structure at twin boundaries and to a lesser extent in the boundaries between adjacent martensite plates.

The tempering of internally twinned martensite leads to a dispersion of carbide particles in a ferrite matrix and this structure which is essentially similar to tempered lower bainite behaves in a like manner as regards crack nucleation and propagation.

2.3.4.3. Retained Austenite

AVERBACH (148) claims that the presence of retained austenite in a structure not only lowers its hardness but leads to rather poor impact properties. This is thought to be due to the transformation of retained austenite to high carbon martensite when plastic deformation occurs during mechanical testing. However, D'ANDREA and ADAMS (114) have shown that retained austenite was not necessarily detrimental to notch toughness at impact rates of

loading. It thus appears that retained austenite may not, of necessity, transform to high carbon martensite under rapid strain rate conditions.

2.3.4.4. Banding

The effect of periodic banding on the impact strength of steels has received little attention in the literature. CAIRNS and CHARLES (149) produced ferrite-martensite banded composites and claim an improvement in transverse and longitudinal tensile and impact properties when compared with those for the ferrite-pearlite banding exhibited by a plain carbon steel. Perhaps the most significant index of brittle fracture performance of welded, low carbon, banded, low alloy steels has stemmed from the investigation carried out by the BWRA (150) on a failed pressure vessel. Fracture was initiated within the weld heat affected zone in a band of high hardness which was predominantly martensitic in nature. Electron probe microanalysis showed this band to be alloy enriched. Bands on either side were of significantly lower hardness and exhibited an upper bainitic structure. Once initiated, the crack propagated through the heat affected zone and into the parent material, under the influence of the externally applied stress resulting from the pressure within the vessel.

2.3.5. Heat Affected Zone Cold Cracking

The most important single problem in welding low alloy high yield strength steels is that of avoiding heat affected zone cracking which occurs adjacent to the fusion boundary. A crack initiates in the region of any stress concentration and propagates in a manner that depends upon the residual stress pattern. In both fillet and butt welds, cracks often initiate at the root or toe of the weld and run parallel to the fusion boundary. Cracking does not always occur instantaneously on completion of a welding operation and often there is an incubation period.

WATKINSON and BAKER (151) attribute cracking to three main factors:

- (a) the residual stress level in the joint.
- (b) the presence of hydrogen.
- (c) the presence of a susceptible microstructure in the heat affected zone.

These workers suggest that together with a temperature range around normal ambient, all these factors must be appropriate for cracking to occur. The most important factor is the presence of hydrogen since crack free joints of high restraint can be made with low hydrogen welding processes e.g. carbon dioxide shielding, where martensitic structures exist in the heat affected zone.

2.3.5.1. Residual Stress

The stress level in a joint will be associated with the yield stress of the steel in question and the maximum stress attainable will be equal to this value. However, joints in a high yield strength steel do not necessarily have higher stresses associated with them than those found in joints of lower strength steels since some control with respect to welding procedure can be applied. GRAVILLE (152) has shown that where parts are not closely fitted or adequately aligned, prior to welding, this can lead to severe stress concentration resulting in cracking where it otherwise would not have occurred. The danger from high stresses, particularly if they are concentrated, is obvious.

2.3.5.2. Hydrogen

Hydrogen is introduced into welds, in a variety of ways, by absorption into the molten weld pool. Normally a potentially low hydrogen process is chosen which is further controlled by limiting the amount of hydrogen bearing compounds in the consumables. HOPKIN (153) has shown that where flux is used the water content of the flux can be reduced by choice of materials while BRADSTREET (154) suggests drying prior to welding. BLINK (155) claims that by reducing the partial pressure of hydrogen in the arc atmosphere by introducing carbon dioxide from carbonates included in the flux, the level of hydrogen in a weld deposit can be kept to a minimum. BRAIN and SALTER (156) have indicated that in gas shielded processes the main sources of contamination with hydrogen are water vapour in the shielding gas and hydrogen generating lubricants and/or moisture either on the wire surface or occluded in pits in the wire.

There has been a large amount of theoretical speculation regarding the method by which hydrogen lowers the fracture stress. Generally, only a small amount of hydrogen present in a steel at room temperature, H_i , is in interstitial solid solution in the matrix and the bulk, H_v , is held at so called traps. The atomic interaction theory of JOHNSON (157) attributes embrittlement to the direct effect of H_i . ZAPFFE (158) and HEWITT and BILBY (159) have proposed an internal pressure theory in which the traps are voids containing H_v in the form of molecular gas, H_2 , at a pressure in equilibrium with H_i ; the gas pressure exerts an internal stress so that fracture occurs at a lower applied stress. GRIFFITH (142) suggested that some or all of H_v is adsorbed at internal surfaces and lowers their surface energy so that the critical size of the crack nucleus is reduced.

SMITH and BAGNAL (160) have recently shown that where hydrogen is present in the heat affected zone of a weld, sulphur has a beneficial effect in reducing the susceptibility to delayed cracking. This is similar to the experience of HEWITT (161) who

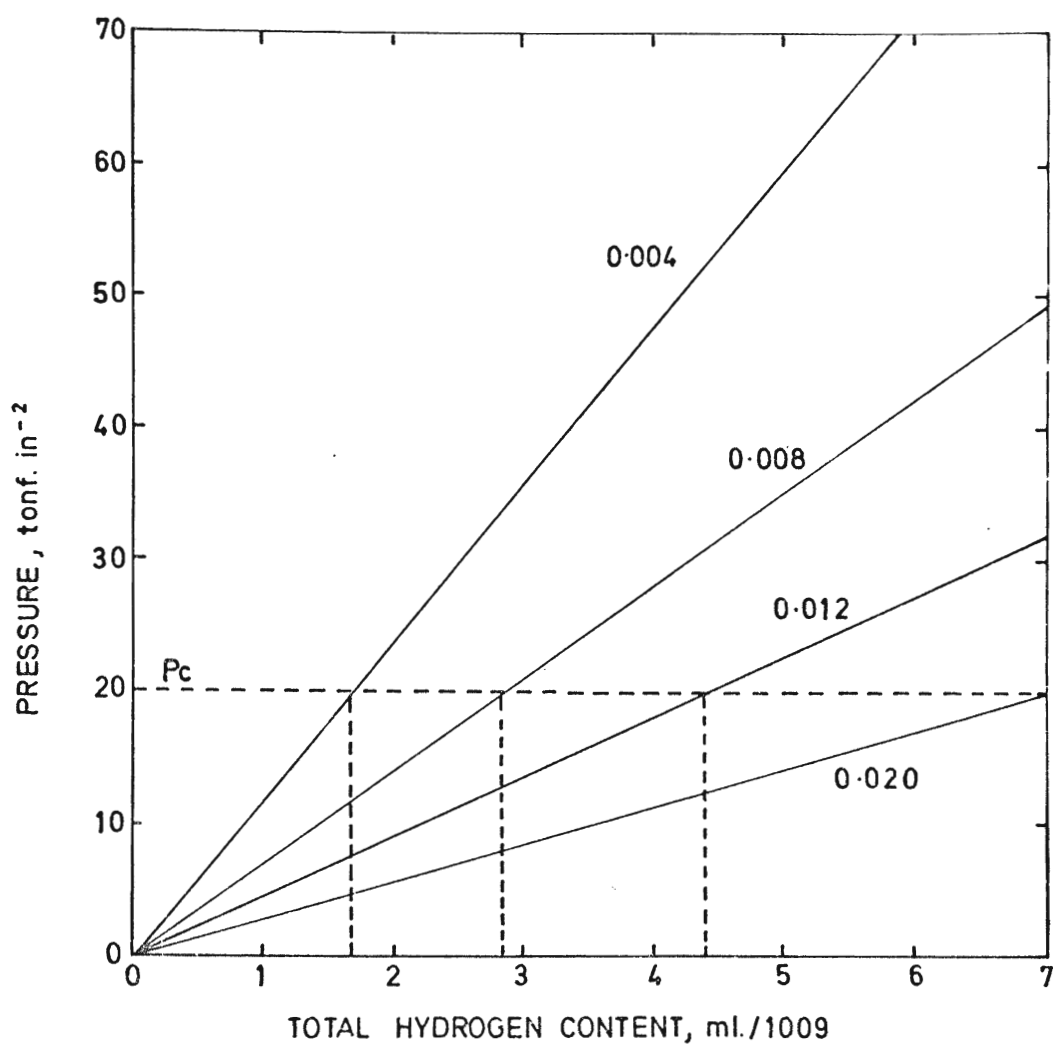


FIGURE 8 BEHAVIOUR OF FOUR STEELS OF DIFFERENT VOID VOLUME (HEWITT¹⁶⁸)

found that sulphur influences the degree of flaking in forgings. For stoichiometric reasons sulphur, when in combination with elements such as manganese, is the most potent void producer in low alloy steels. In addition, VAUGHAN and MORTON (162) found that MnS has a coefficient of expansion greater than the steel matrix so that, on cooling, shrinkage cavities can form voids at the MnS - matrix interfaces. BROOKSBANK and ANDREWS (163) also found that due to this difference in expansion coefficient stresses can be set up in the region of MnS inclusions by thermal cycling. The increased effectiveness of MnS inclusions as hydrogen 'sinks' has been demonstrated by BONISZEWSKI and MORETON (164) who showed that solution of hydrogen in the inclusions could occur. Vacuum degassing during steelmaking has also been claimed by LEECH and WATSON (165) to have a pronounced effect on the incidence of hydrogen cracking.

HEWITT (166) has suggested that above a particular temperature hydrogen is in solution in the matrix but that below this temperature a high proportion is precipitated into voids so reducing the activity of hydrogen in the matrix and retarding its escape to the atmosphere. For a given concentration of hydrogen the pressure developed in the voids and the activity of hydrogen in the matrix are related to a reciprocal function of the void volume. Thus the greater the void volume the lower the pressure in the voids and the lower the concentration of hydrogen in the matrix. HEWITT and MURRAY (167) assume that a critical pressure, P_c , is necessary within a void before crack initiation and propagation, from the void, can occur. As shown in figure 8, for a given value of P_c the hydrogen content increases as the void proportion increases.

The sensitivity of low sulphur steels to delayed cracking may be explained on the basis of inadequate void capacity to safely absorb the hydrogen diffusing from the weld metal into and through the heat affected zone. If cooling to the preheat temperature after welding, is relatively rapid the weld metal will retain a high degree of supersaturation with hydrogen so that the critical factor is the ability of the heat affected zone to accept hydrogen at a finite rate from the weld metal without the pressure in voids or activity in the matrix building up to levels which can cause cracking in a susceptible structure. In low sulphur steels if the heat input is small and a significant stress concentration is present at the heat affected zone - weld metal junction i.e. the weld toe, and the preheat temperature is not maintained to allow hydrogen diffusion to occur, then delayed cracking is a distinct possibility. With high energy inputs and prolonged application of preheat low sulphur steels may not crack since degassing of the weld metal can occur between individual weld runs and will be able to absorb some hydrogen from newly deposited weld metal, so reducing the hydrogen input to the heat affected zone.

Indiscriminate additions of sulphur to improve the resistance of a steel to hydrogen cracking are not recommended since a compromise is necessary between the maximum sulphur content to achieve this aim and the minimum to insure against hot tearing. HEWITT and MURRAY (167) suggest a safe minimum sulphur content of 0.02% in low alloy steels.

2.3.5.3. Microstructure

BONISZEWSKI et al., (111) have employed notched tensile and constant load rupture tests on specimens charged, to varying degrees, with hydrogen to obtain a correlation between microstructure and cracking susceptibility in a range of low alloy steels of differing carbon content.

When auto-tempered low carbon lath martensite is present cracking appears to initiate at inclusions. Initiation is only observed in zones adjacent to the notch where some local plastic deformation has taken place during loading. Crack initiation is also observed at prior austenite grain boundaries lying in a plane transverse to the major applied stress and it has been suggested that the cracks could originate at small clusters of inclusions located at such boundaries. The crack path is found to exhibit frequent changes of direction in passing through the martensite laths, a propagation mode which was described in section 2.3.4.2.

Where the structure is of a mixed nature containing auto-tempered lath martensite and upper bainitic structures crack initiation can occur as indicated above. However, another mode of initiation is possible in that cracks may initiate in the bainitic colonies either parallel or perpendicular to the bainitic ferrite laths. Where the crack passes through auto-tempered martensite laths, changes of direction are frequent but this is not so when bainitic ferrite laths are encountered and crack propagation shows little deviation from a direct path.

In the presence of internally twinned martensite crack initiation may occur by the cracking of large martensitic crystals followed by propagation along prior austenite grain boundaries. This mode of initiation, as indicated in section 2.3.4.2. is believed to depend upon the very high short range transformation stresses which are present in twinned martensitic structures. In the presence of hydrogen the applied stress necessary to nucleate cracks at the small twin defects may be considerably reduced and a low stress fracture can result. Thus, one condition to achieve freedom from cracking when hydrogen is present is to avoid transformation to twinned martensite if any appreciable stresses are going to be present in the material at room temperature. Hydrogen diffusion is much lower in twinned martensite than in other structures. As a result it could be expected that delayed cracking in the weld

heat affected zone could occur after longer times when twinned martensite was present instead of other acicular microstructures.

To avoid any ambiguity due to compositional differences BONISZEWSKI and BAKER (147) subjected a low carbon, 9% nickel steel to an extremely severe quench from the austenite field to produce an internally twinned martensitic structure; normal quenching rates give a structure of auto-tempered lath martensite. In the presence of twinned martensite this steel was found to be embrittled by hydrogen to the same extent as any medium carbon hardenable steel which transforms to the same structure.

2.4. ASSESSMENT OF FRACTURE TOUGHNESS

A small specimen approach to fracture toughness assessment can only be reliably made when a large amount of information exists from the examination of service failures. The Charpy V-notch impact test has been correlated with practice for a limited number of structural types and in such instances can be used, with regard to different materials, in a comparative manner. 'Type' tests may be used in which the conditions correspond more closely to conditions in an actual service structure than can the conditions in small specimen tests. Examples are the crack arrest, wide plate and explosion bulge tests which can be carried out on sizeable specimens using the full plate thickness to be employed in an engineering structure. Simulative type tests can be used to test models which give a close approximation to an actual structure. If a theoretical understanding of all the factors involved in a possible large scale fracture can be compounded then a method of small scale testing and interpretation can be based on this. This approach has inherent difficulties which, for many cases at present, requires approximation for solution but methods of analysis applicable to particular situations have been developed.

Both the Charpy V-notch impact and the interpretative crack opening displacement small specimen tests are suitable for the investigation of weld heat affected zone fracture behaviour. The former test was the only technique available for this present study and as such its characteristics are discussed, in this section, in somewhat more detail than the latter test which, nevertheless, is briefly reviewed together with other forms of test which lend themselves to the assessment of the fracture toughness of weldments.

2.4.1. The Charpy V-notch Impact Test

This test is subject to conventional limitations in that the results are strictly comparative and not fundamental. It

serves a definite purpose provided that it is only used in standardised form. If specimens of non standard dimensions are employed the results obtained can only be compared with those of other test series using similar specimens.

The standard specimen is 60 mm. long by 10 mm. square having at the centre of one side a 2 mm. deep notch forming a 45° angle, with a root radius of 0.25 mm. The specimen is fractured by a single impact blow at the side immediately opposite the notch by a weighted pendulum which, at the point of impact, has a kinetic energy of 240 ft. lb. The energy absorbed by fracture is recorded but does not account for the small amount of energy lost in friction, air resistance, vibration of the testing machine and setting the fractured specimen or specimen parts in motion. However, due to the comparative nature of the test this is not significant.

When a series of Charpy V-notch specimens made from a low alloy steel, is broken at progressively decreasing temperatures the energy required for fracture steadily decreases and finally levels off at a substantially constant value. The ductility also decreases and the fracture appearance alters from dull and fibrous to bright and faceted for high and low energy absorption respectively. It is convenient to describe a so called 'transition temperature' for the transition from ductile to brittle behaviour, based on such observations and a suitable criterion must be selected. The selection of a criterion is usually quite arbitrary and different criteria usually yield different values for the transition temperature. The criteria which can be used are:

(a) Energy Criteria

- (i) Energy at a particular temperature.
- (ii) Arithmetic mean of maximum, E_m , and minimum, E_l , energy values.
- (iii) Temperature at an arbitrary low value of energy such as 15 ft lb. or 20 ft lb.
- (iv) Maximum energy at 100% fibrous fracture appearance.
- (v) Temperature range for transition from low energy fracture to maximum energy.

(b) Fracture Appearance Criteria

- (i) Temperature for a particular level of fibrosity or crystallinity in the fracture, such as the temperature at which the area of fracture with crystalline appearance is 15, 50 or 90% of the original area. This temperature is called the fracture appearance transition temperature (FATT).

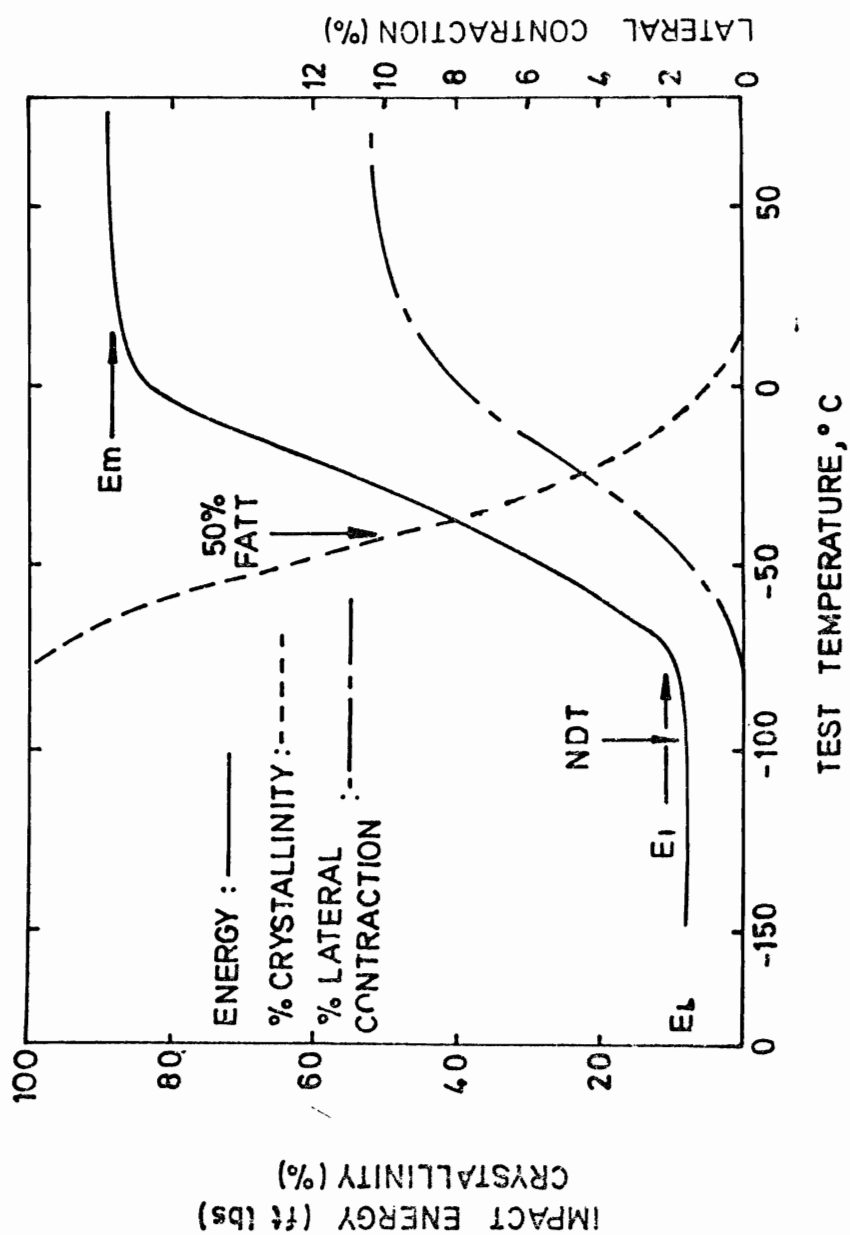


FIGURE 9. TYPICAL CHARPY V-NOTCH IMPACT TRANSITION CURVES.

(c) Ductility Criteria

- (i) Contraction in width at the notch root.
- (ii) Expansion in width at the side opposite to the notch.

In some instances special significance is given to the highest temperature of the lowest horizontal shelf of the transition curve since this value appears to be less dependent upon the testing procedure than other toughness criteria. This temperature is referred to as the nil ductility temperature (NDT) and correlates well with that obtained from the larger scale drop weight test (169). Typical Charpy V-notch impact transition curves, with certain transition criteria annotated, are given in figure 9.

Occasionally the fracture criteria are inter-related and often rate a series of steels in the same order of merit. However, as pointed out by STOUT and MCGEADY (170) a great deal of confusion can occur if the same criterion is not strictly adhered to when different test series, using standard specimens, are compared.

2.4.2. Type, Simulative and Interpretative Tests

The conditions under which a brittle crack will arrest are of importance in limiting catastrophic failure. Tests (171, 172) have been devised to measure the crack arrest temperature (CAT) and involve essentially the tensile loading of an edge-cracked specimen across which a temperature gradient is maintained. The crack propagates through the specimen until it reaches a point where the temperature is sufficiently high to enable yielding to halt its further extension. COWAN and VAUGHAN (173) and NICHOLS (174) prefer the use of an isothermal crack arrest temperature (ICAT) test but economic, amongst other factors, have tended to restrict its use.

The conditions which provide for the initiation of an unstable crack give an alternative basis for designing against failure. The WELLS (175) wide plate test employs fine saw cut notches in the edges of two plates which are then welded together and a tensile stress applied to the line of the weld. In spite of some inherent shortcomings this test provides a valuable procedure for assessing the results of different small specimen tests.

A number of attempts have been made to test to failure models which simulate an actual structure (176-179). These tests are, in the main, concerned with the determination of conditions for crack initiation although in some instances the problem of crack propagation may also be investigated.

PELLINI and PUZAK (176) produced a fracture analysis diagram to give an overall design scheme which took into account both the applied stress level and defect size. The method involved comparison between actual service failures and the results of laboratory tests, particularly the drop weight (169) and explosive bulge (180) tests. It was found, however, that agreement of failure behaviour was restricted, in practice, to mechanical or hydraulic loading conditions in which elastically controlled crack arrest occurred.

A sophisticated approach to the assessment of failure instability condition and brittle fracture is that employing linear elastic fracture mechanics (LEFM) (181). The technique involves stress analysis around a crack or defect in terms of elastic behaviour which is used in conjunction with either of the experimentally determined parameters, stress-intensity factor (K_{Ic}) or crack extension force (G_{Ic}) to establish the conditions at which crack propagation will occur. RADON and TURNER (182) have developed a method of obtaining K_{Ic} values from measurements of failure stress in instrumented Charpy tests. The treatment is based upon the determination and application of fracture toughness values relevant to fully plane strain conditions, which can be much more severe than those experienced in practice.

As discussed in section 2.3. even a so called brittle fracture in structural engineering materials is associated with considerable local plasticity. In such materials, with conventionally sized notched specimens tested at normal strain rates, the plastic zone of a stressed crack can spread across the specimen before fracture occurs so that stress analysis and calculation of K_{Ic} by the LEFM approach is made impossible. The general yielding fracture mechanics approach of WELLS (183) avoids such difficulty by considering that the unstable propagation of a defect occurs at a critical value of local displacement adjacent to the tip of the defect, the critical crack opening displacement (COD), and assuming that this critical value is the same in small specimens of similar thickness to that found in an actual structure. Under conditions of small plastic deformation:

$$G = \sigma_y \delta \quad 2.4.1.$$

where G is the crack extension force, σ_y the yield stress and δ the critical COD. δ exhibits a transition with temperature which suggests that a failure criterion may be selected such that unstable crack propagation in the plane strain mode will not occur. WELLS (184) has extended the stress analysis to include cracks in pressurised cylinders and cracks at nozzles. Currently work is continuing on the COD test in the U.K., to

provide additional evidence of its applicability, to standardise various aspects of testing techniques and to give correlation with more conventional tests such as the Charpy V-notch impact test.

2.5. THE INVESTIGATION OF WELD HEAT AFFECTED ZONE STRUCTURE AND PROPERTIES.

The major difficulty which must be faced when attempts are made to carry out mechanical testing on actual weld heat affected zones is that the complete zone width is of the order of only a few millimetres and may be considered as being composed of a virtually infinite series of planes of varying structure, perpendicular to a line extending from the fusion boundary to the unaffected parent material. However, several workers, notably CABELKA (185) and HATCH and HARTBLOWER (186), have attempted to directly determine the Charpy V-notch impact properties of the weld heat affected zone. In their tests a heavy weld deposit was made on the surface of a plate and Charpy specimens were machined out of the plate, normal to the weld centreline and parallel to the plate surface, with the notch accurately positioned at various distances from the fusion boundary, parallel to the plate surface. The results showed that welding deteriorated the notch toughness properties of the parent plate in the heat affected zone region.

Using this procedure the determination of notch toughness of the various regions in the heat affected zone of a weld is difficult for two fairly obvious reasons. In the first instance, the heat affected zone is so narrow that it is virtually impossible to position the notch accurately in any given microstructure and it is extremely difficult to ensure that the line of the notch root is parallel to the fusion boundary whose lengthwise contour, though essentially straight, does vary due to slight, inherent alteration of process variables during welding. Secondly, during propagation, a crack will pass through a whole series of structures so that the net impact value obtained will be a function of the relative proportion and distribution of the structures present.

It has been established by BURDEKIN (187) and SAUNDERS and DOLBY (188) that the existence of pre-existing defects in the heat affected zone can give rise to severe local embrittlement at the tip of the defect. The severity of the embrittlement was attributed to thermal strain concentration effects and it was suggested that these could only be investigated realistically by the deposition of welds rather than by simulation techniques. On this basis, the majority of fracture toughness studies carried out at the Welding Institute have employed small scale specimens

extracted from various regions of actual weldments. Pre-notched and unnotched plates have been welded and used for comparison purposes. The COD techniques of general yielding fracture mechanics were used as the test method to enable the critical conditions for fracture initiation to be established (189-190). DOLBY (191) has recently studied the nature of the heat affected zone in HY-80 low alloy steel plate using the submerged arc multi-run welding technique, specimens being extracted and notched after welding. COD testing was employed and it is claimed that a more realistic picture of fracture characteristics is obtained since, in practice, a crack initiates at some point in the heat affected zone then propagates through a series of structures before continuing through the parent material. However, the difficulties noted previously with regard to the positioning of notches in the heat affected zone for the purpose of Charpy testing still exists, but, due to the smaller test region concerned, this factor is not considered to be so critical in COD testing.

The overall method is limited to notch toughness testing, be it Charpy V-notch, COD or some other technique, since the extraction of say tensile test specimens from a heat affected zone leads finally to somewhat meaningless information. It may be argued that hardness testing of heat affected zones is an efficient substitute for tensile testing, but this point is open to question in regard to the greater, inherent scatter of results often found in this test.

When optical microscopy is used to examine the heat affected zone structure it is possible to estimate, reasonably accurately, distance from the fusion boundary and hence thermal cycle peak temperature. However, in dealing with low alloy steels in particular it is necessary to resort to carbon extraction replica or thin foil transmission electron microscopy in order to resolve the fine structural detail. In using either method it is almost impossible to obtain an accurate estimate of distance from the fusion boundary and the peak temperature to which a particular specimen has been subjected.

2.5.1. Heat Affected Zone Simulation

An alternative method of ascertaining the structure and assessing the mechanical properties of a heat affected zone is to simulate, in a sensibly sized piece of parent material, the thermal cycle experienced at a particular point in the zone. The simulated material can then be used for testing in the conventional manner.

Many attempts at heat affected zone simulation have been aimed at producing similar microstructures without involving the expense of using large dimension plates. Examples of these are the work of BRUCKNER (192) and APBLETTE et al., (193) who end-

quenched tapered plate sections after furnace and electrical resistance heating, respectively, in an effort to reproduce heat affected zone structures. Structural study was also carried out by CALVO et al., (194) who made a tungsten inert gas (TIG) spot weld on thin plate and used optical microscopy to examine the relatively wide heat affected zone rings on the reverse side of the plate.

Other workers (195-197) attempted to measure the mechanical properties of the heat affected zone by conventionally heat treating fairly large samples of the parent plate material in such a manner as to develop structures comparable to those found in the various regions of the actual heat affected zone.

The material for mechanical testing must be structurally similar to the plane of the heat affected zone which it represents and must also possess identical mechanical properties throughout a sufficiently large volume. To ensure that these requirements are met, the material should ideally undergo thermal and strain cycles of a similar pattern to the plane of the heat affected zone being studied. NIPPES and SAVAGE (198) were the first to develop equipment to simulate any thermal cycle experienced in the heat affected zone in a specimen of suitable size which could subsequently be mechanically tested. In their equipment a specimen of parent material was held at each end by water cooled copper jaws, separated by approximately one inch, which were free to move in the direction of the specimen axis when expansion or contraction of the specimen occurred as it was heated or cooled respectively. Resistance heating of the specimen was due to the flow of current from the secondary winding of a conventional resistance welding transformer. A thermocouple was welded to the mid-point of the specimen and provided a feed back signal indicative of the specimen temperature which was then compared with a reference signal obtained from a cam operated potentiometer. The unbalanced signal obtained was used to control the current in the primary winding of the transformer and hence the heat input to the specimen. A similar reference generator has been used by SUZUKI and TAMURA (199) while other workers have employed plug-in resistor networks (200) or potentiometer arrays (201, 202) giving a stepped cycle which approximates to a continuous cycle at sufficiently small time increments.

Many investigations have used the simulation technique and after thermal cycling the specimen is machined to form a standard Charpy V-notch specimen in which the notch is positioned at the point where the thermocouple has been removed. Materials tested in this manner have included mild and low alloy steels (110, 203-215), austenitic steels (15), maraging steel (216) and titanium alloys (217). The simulation studies of GROTKER (210) are interesting in that they are the only ones reported in the literature which used simulation techniques in an effort to

establish parent plate behaviour during multi-run welding. A range of low alloy steels was studied using thermal cycle relationships established by PLATTE (218). The investigation was somewhat limited in that only the effect of peak temperature temper cycles of approximately 540, 600 and 700°C were applied to specimens initially cycled to an approximate peak temperature of 1310°C.

The simulation technique was extended and used in conjunction with a high speed dilatometer to study the transformational behaviour of materials to be welded (110, 114, 209, 211, 219). In a typical study D' ANDREA and ADAMS (114) found, not unexpectedly, that the whole shape of the thermal cycle is required in order to define the nature of the transformation products, rather than simply the peak temperature attained and the cooling rate through a given temperature range. The equipment devised by INAGAKI et al., (220) utilised induction heating of the specimen with a high frequency current, regulated by a programme control system to reproduce the weld heat affected zone thermal cycles. Continuous cooling transformation diagrams were obtained for a wide range of steel compositions.

The use of weld thermal cycle simulators in investigating heat affected zone hot cracking during welding and post weld heat treatment has recently been reviewed by WIDGERY (221).

In order to duplicate, more accurately, a region of the heat affected zone, the concurrent deformation or strain cycle due to restricted expansion and contraction of the material should, ideally, be considered in addition to simulating the thermal cycle. The heating and cooling cycles accompanying welding result in expansion and contraction within the region affected by the change in temperature and to an extent depending upon the degree of temperature change experienced. When the weld deposit is being made the adjacent plate suffers transverse compression and while the deposit cools the plate undergoes transverse tension. The fact that the metal is unable to expand and contract by virtue of restraint during the thermal cycle leads to plastic deformation within a limited region of the weld deposit and heat affected zone and profound changes occur in the material at the root of a notch or other stress raiser (187, 188).

The greatest difficulty is in measuring the strain cycles associated with the weld heat affected zone and, possibly due to the practical complexity of the problem, these have yet to be obtained. Some workers (222, 223) have made detailed mathematical analyses of the dynamic stresses involved, but these are limited in their application and have yet to be verified experimentally. It has been claimed by KASATKIN and LOBANOV (222) that the thermal stresses which develop in the immediate vicinity of the heat source are much higher than the final residual stresses and that

these dynamic stresses are, therefore, bound to have a much greater effect on the properties of the heat affected zone than the terminal residual stresses. COWARD (224) imposed virtually rigid clamping on thermally cycled specimens which he estimated to be the equivalent of 2-4% strain. He found that this procedure had no measurable effect on the subsequent impact properties of Charpy V-notch specimens machined from the simulated blanks. Whether dynamic strains due to the rapid heating and cooling effect of weld thermal cycles produce changes in the mechanical properties regardless of external restraint and in the absence of a notch is a point still open to conjecture.

2.6 HY-80 LOW ALLOY STEEL

After the cessation of World War 2, world wide effort was made towards the development of high notch toughness steels, particularly for naval construction. Considerable experience had been obtained in the marine industry in using the quenched and tempered STS-steel. Test results on low carbon content heats of this material prompted modification to a low carbon, nickel, chromium, molybdenum steel which when quenched and tempered developed a yield strength of 80,000 lbf. in⁻². (≈ 35.7 tonf. in⁻²) in thicknesses up to 1.25 in., with notch toughness values, based on the Charpy V-notch impact test, of 50 ft.lb. at - 84°C. It was given the grade HY-80 under the general U.S. Military Specification, S-16216. In this form HY-80 steel was employed, with good success, for light armour applications to surface ships such as aircraft carrier flight decks. Further research on HY-80 steel showed that for better weldability two composition ranges, depending upon plate thickness, would be necessary.

In the application of nuclear power to submarine propulsion the hull diameter had to be increased, over that required for conventional power, in order to encase the necessary equipment. Due to its extremely good strength: weight ratio, notch toughness at low temperatures and anti-ballistic properties, when compared with other steels, HY-80 has been extensively used in the fabrication of pressure hulls and other components for the U.S. nuclear submarine fleet. In the U.K., it is now superceding the British QT 35 quenched and tempered low alloy steel in the production of the Dreadnought Class of nuclear powered submarines. McKEE (225) has discussed the position of HY-80 steel in relation to submarine design practice and HELLER et al., (226) in an excellent review paper, have evaluated its use as a structural material for nuclear powered submarines.

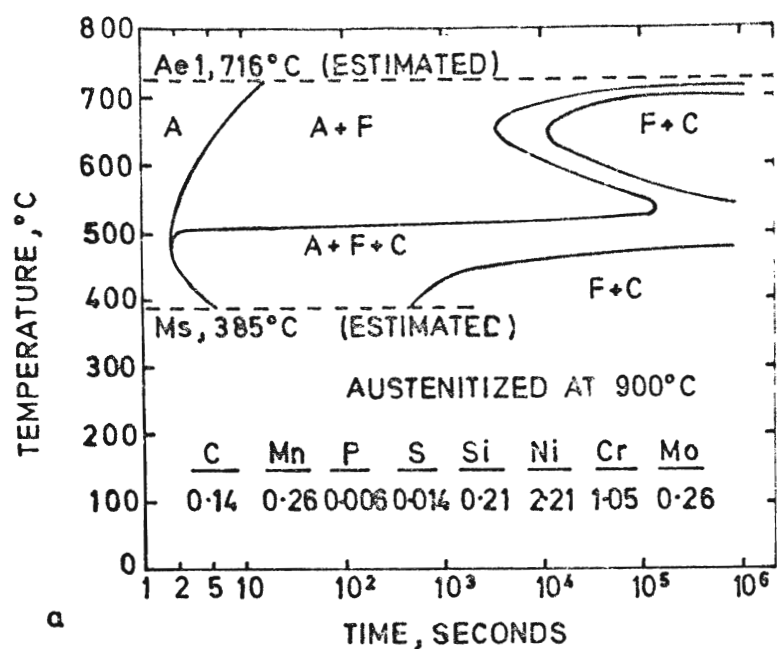
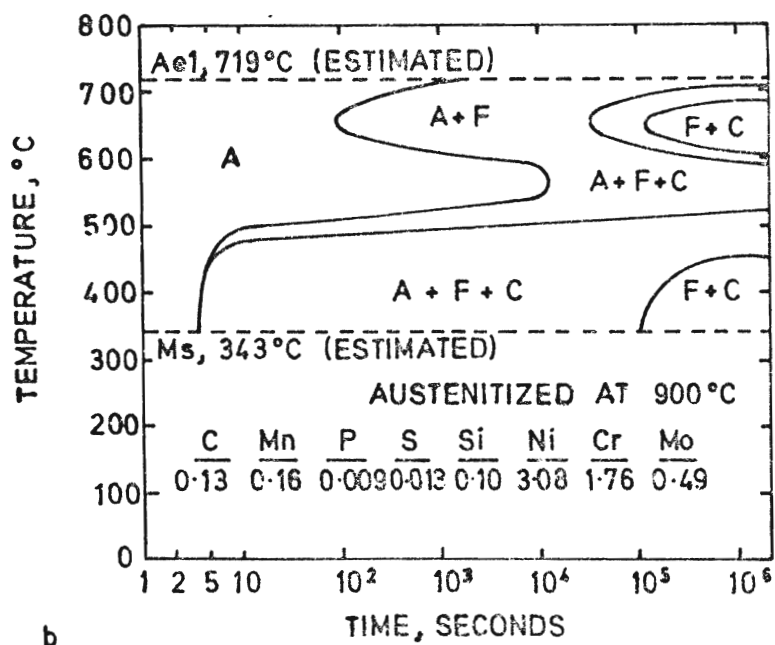


FIGURE 10. ISOTHERMAL TRANSFORMATION DIAGRAMS FOR LEAN (a) AND RICH (b) HY-80 STEEL COMPOSITIONS. (SIMPSON²²³). (ESTIMATED TEMPERATURES AFTER LAMBERT AND GRANGE²³⁴).

2.6.1. Characteristics of HY-80 steel

In the following sections the specification of HY-80 steel is presented and the very limited number of published papers on its characteristics, reviewed.

2.6.1.1. Chemical Composition

The composition ranges used in producing HY-80 steel have a pronounced effect on its transformation characteristics and the associated mechanical properties of the transformed structures. The present Ministry of Defence specification DG. SHIPS/PS/9027 (227) requires plates to the U.S. Navy Department specification, MIL-S-16216G (228). Previous specifications, denoted by the index letters, D (229), E (230) and F (231) indicated composition ranges for two different plate thickness ranges whereas the G modification quotes only one such range. The chemical compositions to these specifications are given in Table 2 and it should be noted that plate to MIL-S-16216D is still being used in submarine fabrication.

The reduction in sulphur content for steels produced to the G specification could improve weldability. EMMANUEL et al., (232) showed that a low sulphur content was necessary to reduce, to a minimum, the incidence of intergranular cracking in the vicinity of the fusion boundary in the heat affected zone of welds in HY-80 steel.

SIMPSON (233) determined TTT diagrams for lean and rich HY-80 compositions and these are shown in figure 10. When lean composition plates over 1 in. thick are heat treated, the TTT diagram indicates the formation of ferrite. Only by increasing the proportions of chromium and molybdenum in the steel can the form of the "nose" of the TTT curve be altered. Increase in carbon content merely extends, somewhat, the time for initiation of transformation.

2.6.1.2. Mechanical Properties

Under MIL-S-16216G mechanical property requirements still remain as for previous specifications in that the limits are referred to plate thickness. The specified mechanical properties are given in Tables 3 and 4.

2.6.1.3. Mechanical Working and Heat Treatment

For deformation below that corresponding to 12% tensile elongation i.e. a radius of 4.2 times the plate thickness, on bending, cold working below 260°C can be employed. For higher deformation hot working is necessary followed by quenching and tempering to recover the original properties. The hot working

range is 1000° - 1150°C, the finishing temperature being above 870°C.

For normal heat treatment the austenitizing temperature prior to quenching should be within the range 885°C - 915°C and a tempering temperature between 620°C and 700°C should be used. Quenching must be sufficiently rapid to satisfy the criterion of 80% martensite at mid-plate thickness. Soaking at the austenitizing temperature should be 1 hour per inch of section thickness and for tempering, a period of 5 hours at the requisite temperature is normally used. Accelerated cooling from the tempering temperature is recommended and double tempering, if necessary, is permitted in order to achieve the specified properties.

2.6.1.4. Microstructure

Apart from work by DOLBY (112), discussed in section 2.6.2.2., on weld heat affected zone structures produced by heat inputs lying outside a specified range (227), no electron microscope study of HY-80 steel has been reported in the literature. WILLNER and SALIVE (236-238) using optical microscopy techniques conducted a series of investigations into the relationship between the structure and properties of HY-80 steel subjected to relatively standard heat treatment procedures. These workers refer to martensite and bainite but, due to the limitations of optical microscopy, do not differentiate between internally twinned and lath martensites or upper and lower or mixed bainitic structures. Their results however infer that the presence of ferrite in an otherwise martensitic structure, produced by quenching from the intercritical range, decreased the transverse Charpy V-notch energy level used for predicting the NDT temperature and after tempering increases the possibility of the presence of inconsistent yield strengths. It was also claimed that a large prior austenite grain size had no effect on the maximum Charpy V-notch level for a given yield strength but that the Charpy V-notch crystalline fracture transition temperature was markedly sensitive to yield strength and microstructure.

An investigation is at present being undertaken by KILPATRICK et al., (238) to correlate structure, as examined by electron microscopy techniques, with TTT and continuous cooling (CCT) diagrams for the grade of HY-80 steel used for submarine construction in the U.K.

In view of the fact that HY-80 has been used in marine construction since 1952 it is more than somewhat surprising that such a small amount of information regarding its structure, as produced by standard heat treatment procedures let alone the complexities introduced by the weld thermal cycle, has been published.

2.6.2. Welding of HY-80 Steel

HY-80 is a relatively simple steel to weld and most welding processes can be successfully applied to produce as-welded joints with toughness and strength approaching that of the quenched and tempered parent material. This performance may be largely attributed to the inherently good toughness of its low carbon auto-tempered martensitic structure when cooled relatively quickly. However, welding variables can produce undesirable structures and three important factors are usually necessary to obtain adequate joint toughness:

- (a) the cooling rate of any austenitized region in the heat affected zone must be sufficiently rapid to form an auto-tempered martensitic structure and avoid the formation of higher temperature transformation products.
- (b) the use of multi-run techniques to derive the benefit from the tempering effect of subsequent runs.
- (c) low hydrogen conditions must be maintained to avoid the possible incidence of underbead cracking.

The cooling rate is a function of heat input, preheat temperature and plate size and the structure produced should not be susceptible to hydrogen embrittlement. Provided that auto-tempered lath martensite is the transformation product in the grain coarsened region of the heat affected zone, tempering can have a beneficial effect on toughness. Thus multi-run welding can be used advantageously where, due to the actual size or design of a fabrication, reheating for tempering is not considered suitable.

2.6.2.1. Heat Input and Preheat Temperature

The cooling rate in HY-80 heat affected zones is critical and is a function of both heat input and preheat temperature.

A low heat input with its associated rapid thermal excursion into the austenite region produces fast cooling rates in the heat affected zone. Due to the degree of austenite heterogeneity produced by the rapid thermal cycle the grain coarsened region of the relatively narrow heat affected zone can contain mixed auto-tempered lath martensite, lower bainite and internally twinned martensite. As a function of the presence of internally twinned martensite low heat inputs produce structures which can promote underbead cracking due to their susceptibility to hydrogen embrittlement. This, therefore, sets a lower limit on the heat input which can be employed.

Conversely, if too high a heat input is used the heat affected zone will be relatively wide and the associated cooling rates from the austenite field, slow. This condition tends to produce a structure, consisting of mixed auto-tempered lath martensite and upper-bainitic transformation products, whose susceptibility to heat affected zone cold cracking is also relatively high. Thus an upper limit on heat input must also be imposed.

DG. SHIPS/PS/9027 (227) specifies the permissible range of heat input to be 30 to 55 kJ/in., calculated according to the formula,

$$\text{Heat input (joule/in)} = \frac{\text{Arc voltage} \times \text{Amperage}}{\text{Welding Speed (in/sec)}}$$

The welding current, arc voltage and travel speed should be adjusted to remain within the heat input range. This is important in multi-run welding where the initial runs employ low voltage - high current conditions for maximum penetration and subsequent runs high voltage - low current in order to obtain maximum bead width.

It is claimed that good notch toughness can be obtained through the use of heat inputs within the specified range. Presumably this is due to the structure consisting predominantly of auto-tempered lath martensite. DOLBY (112) has examined welds produced by heat inputs of 26 kJ/in and 67.5 kJ/in and found adjacent to the fusion boundary in the latter small regions of internally twinned martensite and lower bainite and in the former areas of upper bainite, both being associated with a mainly auto-tempered martensitic structure. Fracture toughness assessment, by means of COD tests, indicated that these regions were inferior to the parent material, which had a structure of temper carbides in a ferrite matrix, but were, themselves, similar in nature. It is feasible that even within the specified heat input range regions of internally twinned martensite, upper and lower bainitic structures and mixtures of these, could result from austenite heterogeneity and also due to the presence of banding in the parent plate. Thus, it is possible that fracture toughness can be detrimentally affected even if the heat input is within the specified limits.

Due to its marked effect in reducing the cooling rate, preheating is not generally favoured in the welding of quenched and tempered low alloy steels. However, if a given amount of preheat gives a satisfactory cooling rate in the heat affected zone, as with HY-80, it is used since it can provide a high degree of insurance against cold cracking, particularly due to the presence of hydrogen. Where preheating is employed, the heat input must be lower the higher the preheat temperature for a given plate thickness, in order to achieve an acceptable cooling rate in the

heat affected zone. The specified preheat temperature for HY-80 is within the range 120° - 150°C. For small welds full preheat is required e.g. attachment and tack welds. High inter-run temperature are generally favoured in multi-run welds as they tend to prevent cold cracking and slag entrapment but they should be such as to give the required structure in the heat affected zone. In repair work preheat is important since the degree of restraint is generally greater than during the original welding. MARNER (239) recommends the following scheme for the welding of HY-80 steel,

Plate thickness, in.	Preheat temp. °C min.	Interpass temp. °C max.	Heat input joules/in. max.
< 0.5	25	150	45,000
0.5 to 1.125	50	150	55,000
> 1.125	95	150	55,000

2.6.2.2. Multi-run Technique

Multi-run welding is carried out using smaller gauge electrodes, at smaller heat inputs, than would be employed in a single pass in order that the process may provide:

- (a) the required cooling rate in the heat affected zones of different runs.

and (b) improved notch toughness of the weld deposit.

STOUT et al., (240) have recently demonstrated the beneficial effects to be gained from multi-run welding in improving the strength and toughness of low alloy steel weld metal.

As discussed in the previous section, undesirable structures with associated high hardness and relatively low fracture toughness may be developed in the heat affected zone of a weld in HY-80 steel. The tempering bead technique is used for all multi-run welds in HY-80 steel except where specific approval has been given for its omission. Its object is to reduce the high hardness and relatively low fracture toughness associated with undesirable structures in the heat affected zone, particularly in the region of the toe of the weld. The technique consists of arranging the deposition of weld passes in such a manner that the heat affected zone of each pass, fusing with the parent material, is tempered by a subsequent pass. Thus passes in contact with the parent material are never made last. The technique is of

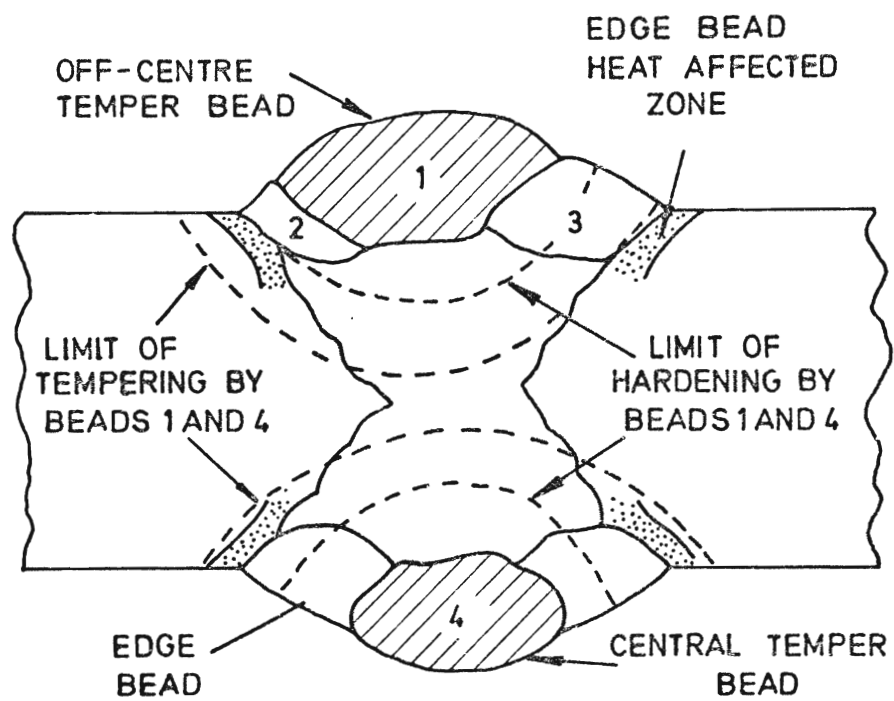


FIGURE 11. EFFECT OF TEMPER BEAD POSITION ON THE DEGREE OF TEMPERING IN EDGE BEAD HEAT AFFECTED ZONES

primary importance in respect of the final surface runs since sub-surface layers generally have their heat affected zones tempered by subsequent layers. The positioning of the tempering bead must be carefully controlled since if it is not centred properly with respect to the two heat affected zone boundaries a significant tempering effect cannot be obtained in part of the heat affected zone. These effects are illustrated in figure 11 where the lower tempering bead is correctly positioned but the upper one, bead 1, which is off centre has adequately tempered the heat affected zone due to bead 2 but not that due to bead 3. However, had tempering bead 1 been positioned more adjacent to the parent plate, over bead 2, then instead of having a tempering effect in the heat affected zone due to bead 2, further thermal cycling with the production of detrimental structures could occur in this region. Thus structures susceptible to cold cracking can exist in the region of the weld toe, on each side of a weld, even though the tempering bead technique has been employed. It is possible to envisage a multiplicity of combinations of two thermal cycles in the edge bead heat affected zone due to the application of a tempering bead. In an effort to minimise the effect, the welding specification for HY-80 requires that approximately 0.125 in. of the edge beads should be exposed after deposition of the temper bead. As shown by OLDRIDGE (241) there are immense practical difficulties involved in maintaining accurate positioning of the tempering bead during deposition such that its efficiency in providing a tempering effect in the heat affected zones of the last deposited layers, adjacent to the parent material, may be considered to be open to conjecture.

Typical bead sequences are shown in figure 12 but varying job and welding conditions will require departures from the indicated number of beads. In certain instances logical bead sequence can provide the desired effect. If a tee joint is made between HY-80 and mild steel then, if the fillet is made with more than one pass, the last pass can be made in contact with only the mild steel to avoid the formation of an untempered zone in the high strength HY-80 steel.

YUKAWA (242) claims that a so called anti-crack initiation (ACI) bead extra welded to the weld toe locates a possible undercut at a position away from the zone along the main weld fusion line. The stress in the fusion line is supposed to be mitigated by slanting it towards the major stress direction near the toe. Even so, a crack could be nucleated at such a position and it is open to doubt as to whether a crack would extend into the parent material as opposed to propagating along the heat affected zone fusion boundary.

2.6.2.3. Low Hydrogen Conditions

The deleterious affects of hydrogen in promoting heat affected

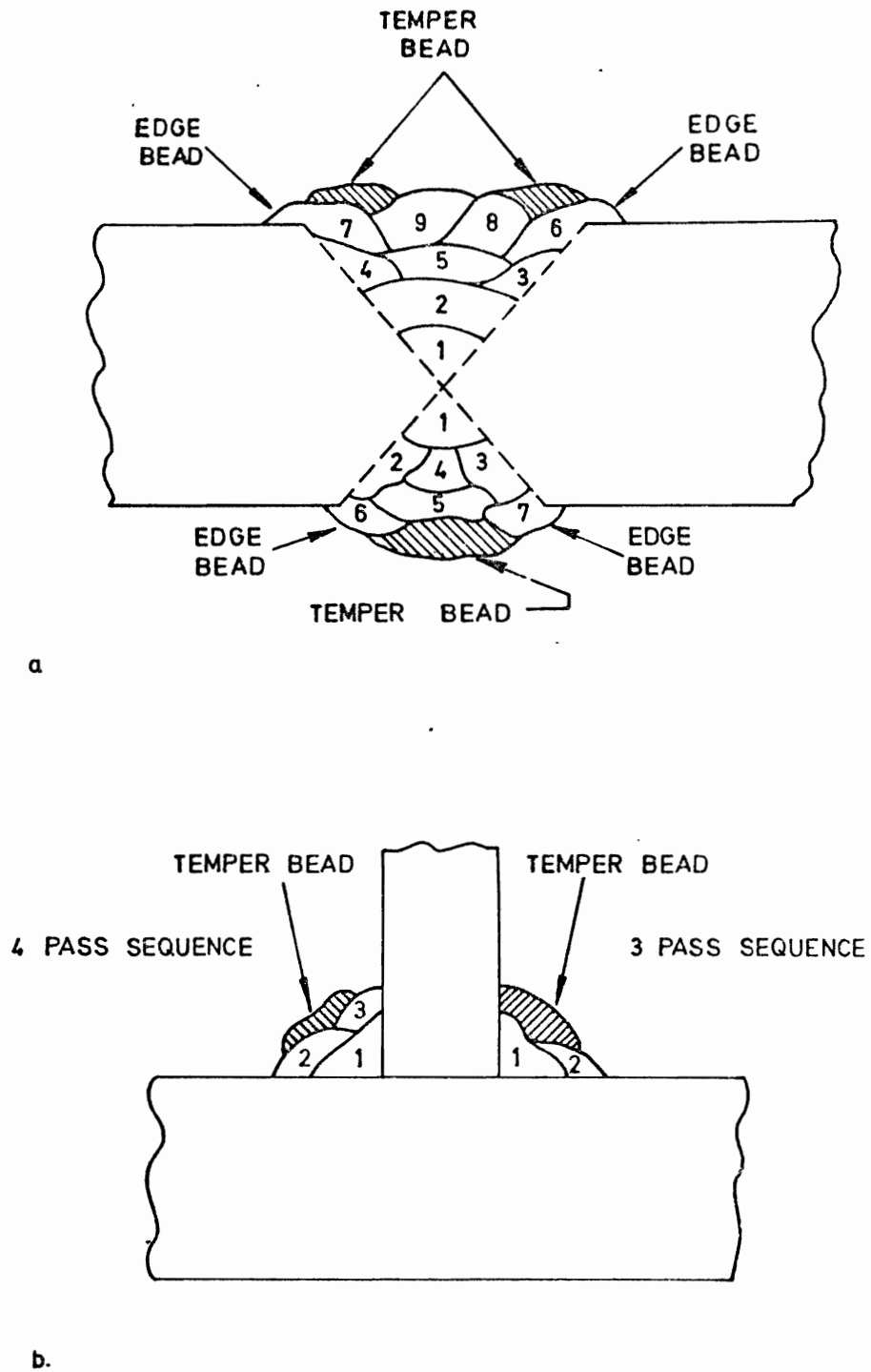


FIGURE 12. TYPICAL APPLICATIONS OF THE TEMPER BEAD TECHNIQUE.

zone cold cracking have been discussed in section 2.3.5, where it was also noted that the presence of hydrogen in a heat affected zone structure is primarily due to the decomposition, in the arc, of water and other compounds. Wet work pieces or bare metal electrode wire are sources of hydrogen and in such cases the remedy is obvious. In the welding of HY-80 hydrogen can be introduced via the electrode coating, the shielding gas or the flux in the shielded metal arc, gas metal arc and submerged arc processes respectively. In the case of coated electrodes and fluxes the moisture content can be reduced by baking at 430° - 450°C followed by preheating at 150° - 200°C , only preheating being necessary for fluxes. Complete details of electrodes and fluxes used in the welding of HY-80, and their treatment prior to welding are contained in D.G. SHIPS/PS/9027 (227).

2.6.2.4. Post Weld Heat Treatment

The M.O.D., specification (227) does not consider the stress relieving of HY-80 fabrications necessary nor desirable and is therefore not normally carried out. If for any reason stress relief is required it is undertaken at $550^{\circ} \pm 15^{\circ}\text{C}$ this temperature being held for one hour per inch of thickness of the thickest member in the weldment. The heating rate above 250°C should not exceed 200°C per hour or $200/t$ per hour, where t is the material thickness, whichever is the lower, in order to avoid setting up thermal stresses. Depending upon the fabrication configuration any method of cooling from the stress relieving temperature may be employed.

3. EXPERIMENTAL PROCEDURE

3.1. MATERIAL

The 1.5 in. thick HY-80 steel plate supplied by the Welding Institute, Abington, Cambs., had the following chemical analysis which conformed to specification MIL-S-16216G (Table 2).

C	Mn	Si	S	P	Ni	Cr	Mo	V	Al	N
0.16	0.32	0.30	0.017	0.006	2.54	1.31	0.28	0.01	0.015	0.01

No mechanical property data was supplied with the plate and tensile and Charpy V-notch impact tests were conducted on specimens obtained from the plate mid-thickness. The results indicated that the mechanical properties were in accordance with the MOD (Navy) schedule and conformed to specification MIL-S-16216G, (Table 3).

Specimen Orientation	Charpy Energy [†] at -84°C, ft-lb.	Yield strength tonf. in ⁻²	U.T.S. tonf. in ⁻²
Transverse	79	37.4	46.7
Longitudinal	93	38.6	47.8

[†] notched in the through thickness direction.

Details of the heat treatment received by the plate were not supplied but the results of mechanical testing and metallographic examination suggested that it had been carried out in accordance with the specification D.G. SHIPS/PS/9027 (227) viz., austenitized within the range 885 to 915°C followed by water quenching and then tempered for approximately 5 hours in the range 620 to 700°C.

3.2. THERMAL CYCLE MEASUREMENT

Thermal cycles were measured by SMITH (243) in the heat affected zone, at various distances from the position of maximum penetration below the weld centre line, for a bead-on-plate weld

in 1.5 in. thick material. A detailed description of the technique employed has been given by COWARD (224). The bead-on-plate weld was produced by a British Oxygen Company submerged arc welding unit using 0.1875 in. diameter filler wire. The welding conditions used were:

Current, amp.	Voltage	Travel Speed, in. min ⁻¹ .	Preheat Temper- ature, °C	Nominal Heat Input kJ. in ⁻¹ .
500	30	16.66	120	54

Due to the experimental difficulty involved in thermocouple positioning prior to welding only three complete thermal cycles were determined; these cycles had peak temperatures of 1275, 765 and 550°C. An additional bead-on-plate weld was produced for metallographic investigation and hardness traverse of the heat affected zone.

3.3. COMPUTED THERMAL CYCLES

The 1275 and 550°C peak temperature thermal cycles measured by SMITH (243) are typical of those experienced by points within the grain coarsened and sub-critical regions of the heat affected zone respectively. On the basis of experimentation, involving a large number of alloy steels, ANDREWS (244) developed the following empirical formulae, which are claimed to be correct to $\pm 10^\circ\text{C}$, for the A_{c1} and A_{c3} temperatures in alloy steels:

$$A_{c1} = 723 - 10.7\text{Mn} - 16.9\text{Ni} + 29.1\text{Si} + 16.9\text{Cr} + 290\text{As} + 6.38\text{W}$$

3.3.1.

$$A_{c3} = 910 - 203\text{C} - 15.2\text{Ni} + 44.7\text{Si} + 104\text{V} + 31.5\text{Nb} + 31.1\text{W} \\ - 30\text{Mn} - 11\text{Cr} - 20\text{Cu} + 700\text{P} + 400\text{Al} + 120\text{As} + 40\text{Ti}$$

3.3.2.

where C, Mn, Ni etc., are expressed as weight percentage. In terms of these equations the A_{c1} and A_{c3} temperatures of the HY-80 steel plate used in this study are:

$$A_{c1} = 708^\circ\text{C} ; \quad A_{c3} = 800^\circ\text{C}.$$

COWARD and APPS (17) found that rapid heating rates, similar to those experienced within a weld heat affected zone raised the Ac_1 and Ac_3 temperatures, calculated by means of equations 3.1 and 3.2, by approximately 35 and 75°C respectively. Modified values for the rapid heating of HY-80 are, therefore:

$$Ac_1 \cong 743^\circ\text{C} ; \quad Ac_3 \cong 875^\circ\text{C}.$$

Even accounting for the inherent error in estimation by using equations 3.1 and 3.2, the 765°C peak temperature cycle measured by SMITH (243) corresponds to a thermal excursion into the inter-critical region.

Many previous investigations have shown that the grain coarsened region in the heat affected zone of low alloy steels inevitably has a structure which results in poor impact test performance and a high susceptibility to hydrogen cracking. SAVAGE and OWCZARSKI (110) demonstrated that thermal cycling into the intercritical range could also produce structures which had a detrimental effect on impact properties. In view of this behaviour it was decided to employ the measured 1275 and 765°C peak temperature thermal cycles in the simulation study on HY-80 steel.

For comparison, it was desired that a thermal cycle representing an excursion into the grain refined region of the heat affected zone should also be employed. In addition, a relatively high subcritical peak temperature cycle was required, by virtue of which the effect of a subsequent tempering cycle on previously simulated material could be investigated. This procedure would indicate the effectiveness of virtually the maximum degree of tempering, in the heat affected zone of an edge bead, due to the deposition of a partly overlapping tempering bead. It was, therefore, decided to attempt to produce 930 and 650°C peak temperature thermal cycles by using a solution to the basic differential equation for 3-dimensional heat flow (equation 2.1.1., page 7). The 3-dimensional situation was chosen since this appeared to give the closest possible correlation to the conditions which exist during the submerged arc welding of a pre-heated 1.5 in. thick plate which is relatively long and wide.

In order to obtain as high a degree of flexibility as possible the solution due to GROSH et al., (12) (equation 2.1.16., page 11) was chosen since this included provision to effect variation of thermal properties with temperature. A slightly re-arranged version of equation 2.1.16. was used for computer programming, the symbols used, however, being in accordance with Table 1, page 165:

$$T = \frac{1}{m} \left\{ \left[\frac{Q m}{2 \pi k_o R} \exp(-\lambda v (vt+R)) + (1+mT_o)^2 \right]^{\frac{1}{2}} - 1 \right\} \quad 3.3.3.$$

A computer program, VARTHERM 1, based on this equation, was written in Algol 60 programming language (245) and run on an ICL 1905 Series computer to obtain a plot of peak temperature versus distance, y , from the point heat source. Values which were consistent with the welding conditions listed in section 3.2 were assigned to appropriate variables in equation 3.3.3. and presented to the computer as data.

A second program VARTHERM 2 gave plots of temperature versus time for values of y , obtained from VARTHERM 1, which corresponded to 1275, 930, 765 and 650° C peak temperatures respectively. In this program, as in VARTHERM 1, the thermal conductivity was related to the temperature by

$$k = k_0(1 + mT) \quad 3.3.4.$$

where k_0 is the intercept on the thermal conductivity axis at 0°C and k_0m the slope of the thermal conductivity - temperature relationship for an austenitic steel (25). By comparison with the measured 1275 and 765°C peak temperature cycles the corresponding cycles produced by VARTHERM 2 showed quite major discrepancies throughout the whole temperature range during the cooling portion of the cycle. The heating to the peak temperature was found to be consistent with practice.

APPS and MILNER (10) suggested that due to the existence of convection currents within the molten weld pool, the latent heat of fusion was effectively released at the tail of the pool, some distance from the electrode heat source. A model was therefore invoked which provided for a second point heat source to lag behind the primary source, by a delay time t_D , which corresponded to the evolution of latent heat by the molten weld pool. By comparing the measured cycle and the cycle computed by VARTHERM 2 for a peak temperature of 1275°C and incorporating the temperature-time differences observed, in another program, VARTHERM 3, the heat flow from the idealised secondary latent heat source was found to be essentially of a sigmoidal nature. On this basis the following equation was derived to account for heat flow from the secondary point source:

$$L = L_0 \left\{ 1 - \exp \left[-(t + t_D)^n/b \right] \right\} \quad 3.3.5.$$

where L is the latent heat flow at any time $(t + t_D)$ after latent heat release has begun, L_0 the maximum value of latent heat flow, b a positive constant and N are even integer. It should be noted that t_D is positive and that the trend of t is from positive values through zero to negative values. A further computer program VARTHERM 4 was written to include the latent heat release function and with the incorporation of suitable values for L_0 ,

N, t_p and b, in data form, a revised plot of peak temperature against distance, y, from the point source was obtained. Using y values corresponding to peak temperatures of 1275 and 765°C another program VARTHERM 4(2), which also included the latent heat release function, was run to obtain full thermal cycles. Good agreement was obtained with the measured 1275°C peak temperature cycle in the higher temperature region.

However, below about 550°C the computed values of temperature were relatively high and this discrepancy was attributed to the use of the thermal conductivity - temperature relationship for austenite in the lower temperature region.

An assumption was therefore made that in a 1275°C peak temperature cycle a fully austenitic structure was undercooled to about 550°C and below this temperature transformed to a predominantly martensitic structure containing some upper bainite. This fact was confirmed by metallographic examination. Below 550°C, during austenite transformation, the thermal conductivity was taken to vary as the polynomial function:

$$k = A + BT^n + CT^{n-1} + DT^{n-2} + \dots \quad 3.3.6.$$

where A, B, C etc., are constants and n is a positive integer. A final program, VARTHERM 5, was written to include both the low temperature thermal conductivity function (3.3.6.) and that for latent heat release (3.3.5.) and very good agreement was obtained when comparison was made with the measured 1275 and 765°C peak temperature cycles. Accordingly program VARTHERM 5 was run to obtain, for subsequent simulation studies, complete thermal cycles for peak temperatures of 930 and 650°C.

This description of the application of computer programming to obtaining close correlation between theoretical thermal cycles and the actual cycles experienced by points in a weld heat affected zone has, of necessity, been brief. Complete details regarding the development and use of the series of computer programs, VARTHERM 1-5, and the reasoning behind their format are contained in the Appendix to this thesis.

3.4. THERMAL CYCLE SIMULATION

Weld thermal cycle simulation was carried out on an apparatus designed and built at Cranfield in which a material blank of suitable dimensions is heated by virtue of its own resistance to the passage of an electric current. The Cranfield simulator has been described in detail by CLIFTON and GEORGE (202)

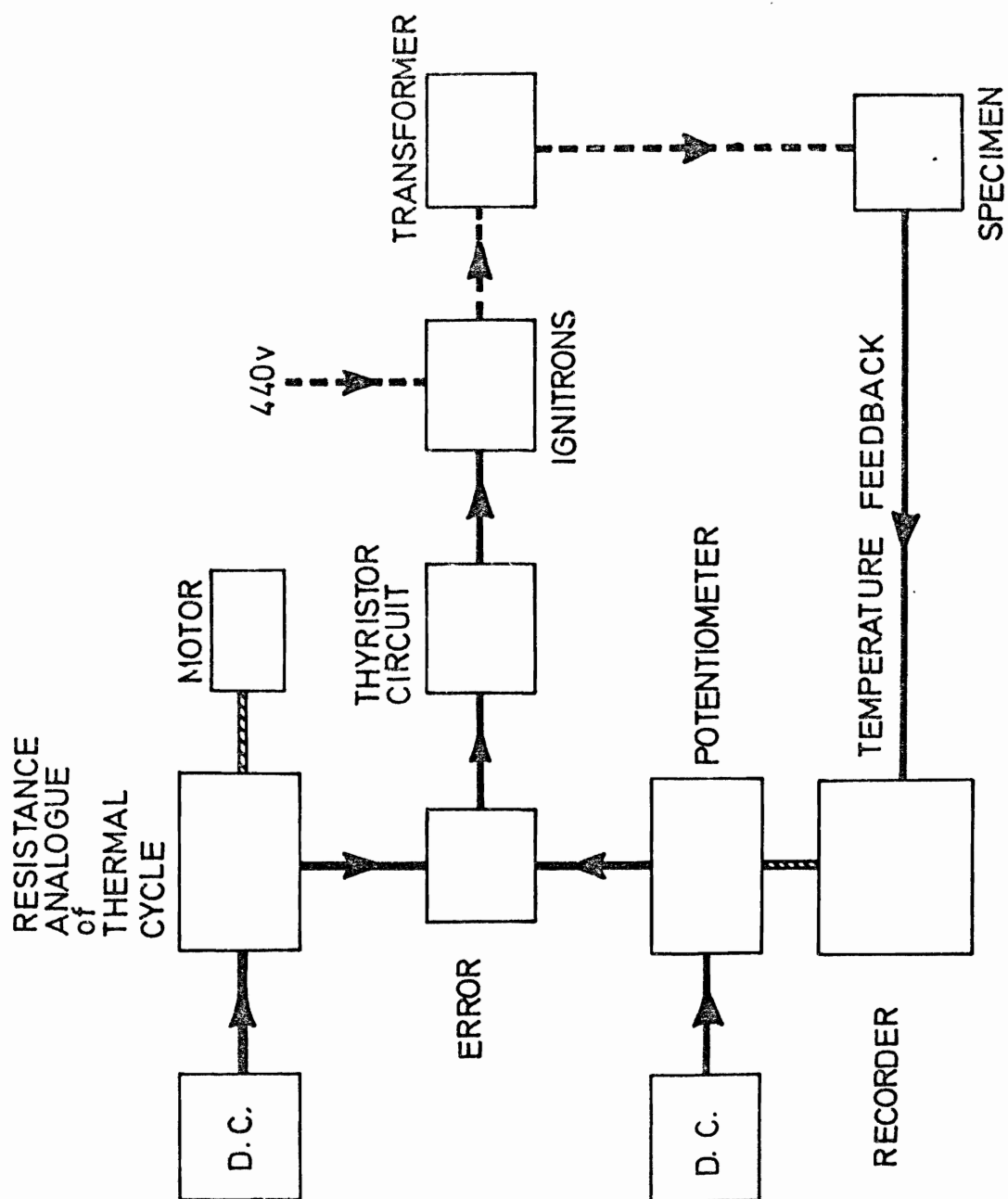


FIG.13. SCHEMATIC DIAGRAM OF SIMULATOR.

and only a brief description of its construction and operation is necessary in the context of this study. A block diagram of the simulator is given in figure 13. for illustration of the foregoing text.

The heating current to the simulator is supplied by a 50kVA resistance welding transformer and transmitted to the blank via hollow, water cooled brass clamp blocks. To standardise as far as possible the contact resistance between the blank and the clamps, aluminium packing pieces are placed on each face of the blank which would be in contact with a brass block and the clamping pressure maintained constant by tightening the clamp holding bolts to 50 lb-in. by means of a torque wrench. The thermal cycle experienced by the blank is measured by a thermocouple, capacitor discharge welded to the centre of one of the blank faces; this thermocouple also provides an error signal to control the thermal cycle. A voltage analogue of the required thermal cycle is generated by a single turn potentiometer with 34 taps, equally spaced over its resistance range, driven at constant speed over its 340° arc. A bank of multi-turn potentiometers is connected in parallel with the segments of this function generator potentiometer and the e.m.f., generated by the control thermocouple is fed into a Sefram 'Graphispot' rapid response single channel pen recorder. A 10 k Ω linear potentiometer is mechanically connected across the scale of the recorder with the wiper fitted to the pen carriage so that not only is a record of the thermal cycle sustained by the blank recorded but by the application of an external voltage across the potentiometer an output voltage analogous to the blank temperature is given. With selected voltages applied to the function generator and 'Graphispot' potentiometer a suitable error voltage is obtained which is fed directly into the heating control circuit. The power supply to the 50kVA transformer is controlled by 2 ignitrons connected back-to-back such that each fires on successive half cycles of the 440 volt AC mains supply. The error signal voltage is fed to a thyristor circuit which varies the proportions of each half cycle over which the ignitrons fire. The circuit is such that zero voltage is required for maximum ignitron firing time and + 10 volts for the no-firing condition. Thus the error signal has to be biased from a subsidiary DC supply. Modification to thermal cycle shape is obtained by altering the parallel resistor values while the peak temperature attained and its position with respect to time are controlled by variation of the input voltages to the 'Graphispot' and function generator respectively. An important feature of the equipment in that heating rates up to approximately 600°C/second can be obtained.

The blanks employed in this simulation study were 0.42 in. square with a length of 3.25 in. and were all machined from the mid-thickness of the as-supplied 1.5 in. thick plate, axially transverse to the rolling direction. The ends of the blanks were

stamped to identify the faces normal to the original plate surface and to one such face of each blank a chromel/alumel thermocouple, of 0.015 in. wire diameter, was capacitor discharge welded. After machining, the blank surfaces had a milled finish and the faces parallel to the original plate surface were finished to 00 grit size to provide as even a surface as possible for contact with the aluminium spacers in the simulator clamps. The additional expense and labour involved in fine machine grinding was found to be unnecessary since the quoted finished finish gave comparable thermal cycle reproducibility during simulation. Constancy in blank length and cross sectional area were, however, both extremely important for reproducible results.

COWARD (224), showed that the application of restraint by rigid clamping of the blanks did not affect impact performance in the Charpy V-notch test. In this present study all blanks were allowed axial freedom of movement during cycling to eliminate, as far as possible, the number of variables in the simulation technique.

Voltage analogues were set up on the basis of the measured and computed thermal cycles and batches of blanks simulated accordingly. A standard clamp gap width of 0.5 in. was used for each peak temperature thermal cycle for blanks from which Charpy V-notch impact test specimens would be produced. SMITH et al., (213) had previously found that by using modified No. 13 Hounsfield tensile test specimens made from mild steel with a 0.3 in. gauge length, there was no variation in yield strength, U.T.S., or reduction of area when compared with results for a standard No. 13 specimen. This provided the criterion of a minimum axial length of 0.15 in. of uniformly cycled material on either side of the control thermocouple in simulated blanks from which modified No. 13 Hounsfield tensile specimens would be produced. The clamp gap width was increased, therefore, as far as possible, when such blanks were simulated. An increase in clamp gap width produced a specimen with an increased volume of uniformly cycled material but such increase was limited by the rate of thermal conduction along the specimen to the water cooled copper blocks. The clamp gap widths employed for blanks which were to be subsequently used for tensile test specimens were:

Cycle Peak Temperature, °C	Clamp Gap Width, in.
1275	0.6
930	0.8
765	0.8
650	1.0

Particularly during the 1275°C peak temperature cycle trouble was experienced due to thermocouple detachment during cycling and resulted in loss of the blank for further investigation. In specimens which were used to simulate the multi-run welding technique, where a second cycle was imposed on an already cycled blank, it was decided, in the interest of material economy, to remove the original thermocouple and capacitor discharge weld a new one on to the surface prior to the second cycle. This procedure proved satisfactory although isolated instances of thermocouple detachment occurred.

The following table lists the double cycles used in this simulation study to investigate multi-run effects. Also listed are the single cycles which formed a basis for comparison.

Peak Temperature, °C		Peak Temperature, °C	
Initial Cycle	Second Cycle	Initial Cycle	Second Cycle
1275	-	1275	765
930	-	1275	650
765	-	930	765
1275	1275	930	650
1275	930	765	650

28 blanks were simulated for each condition, half of which were subjected to a post cycle furnace heat treatment at 650°C prior to mechanical testing and metallographic examination. Thus, 20 conditions were investigated each of which was represented by 14 blanks, 10 to be used for the preparation of Charpy V-notch impact test specimens, 3 for modified No. 13 Hounsfield tensile test specimens and 1 for metallographic examination and transverse hardness survey. Additional specimens were cycled to 1275, 930 and 765°C peak temperatures using gap widths for Charpy and tensile specimen blanks in order that the volume of uniformly cycled material could be estimated by means of longitudinal hardness traverses.

3.5. POST CYCLE HEAT TREATMENT

As noted in section 3.4, half the total number of simulated blanks were subjected to a post cycle heat treatment (PCHT) in an effort to determine the effects on structure and properties of post weld heat treatment in practice. The PCHT temperature of 650°C was 150°C higher than that recommended by D.G. SHIPS/PS/

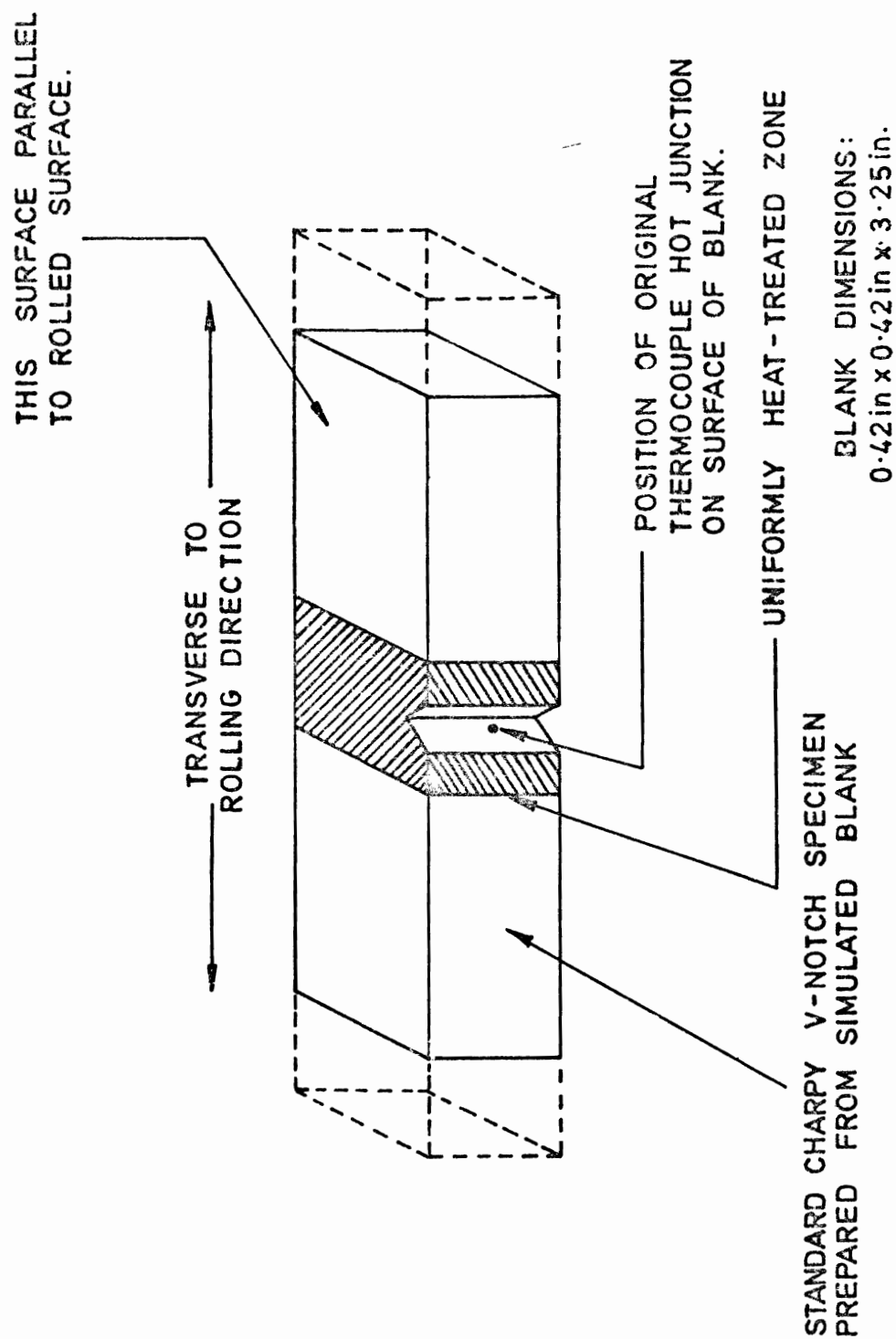


FIGURE 14. DETAILS OF SIMULATED SPECIMEN PREPARATION.

9027 (227) for post weld stress relief but was employed to investigate the possible innovation of another practical treatment. The recommendations of D.G. SHIPS/PS/9027 regarding rate of heating to the PCHT temperature were adhered to in that the simulated blanks were placed in an electric hump furnace at 250°C and heated to 650°C in 50 minutes. The blanks were soaked at 650°C for 1 hour then removed from the furnace and allowed to cool in still air to room temperature. In practice, furnace cooling is employed to eliminate the possibility of distortion occurring in welded assemblies. However, in the present investigation the problem of distortion was absent and as it was solely of a metallurgical nature air cooling from the PCHT temperature was warranted.

3.6. MECHANICAL TESTING

The simulator blanks were ground and notched in the through thickness direction, along a line which included the control thermocouple, to produce standard Charpy V-notch impact test specimens. The notch orientation with respect to the simulated blank and thermocouple position are shown in figure 14. Except for the determination of the primary rolling direction in the parent plate, specimens oriented transverse to this direction were used exclusively in this study to obtain a more conservative estimate of notch toughness. All specimens were tested on a Losenhausen impact testing machine. The energy absorbed in specimen fracture, at a known temperature, was recorded automatically and the percentage crystallinity in the fracture surface was estimated by measuring the extent of the crystalline area with a finely graduated steel rule. This latter technique is rapid in application and is considered to be accurate to $\pm 10\%$.

The procedure suggested by DRISCOLL (247) was used to minimise errors in the test. Specimens were tested between -196 and + 60°C. For temperatures below ambient the specimens were placed in iso-pentane ($(\text{CH}_3)_3 \text{CH} \cdot \text{C}_2\text{H}_5$) solution in a thermos flask and cooled to the required temperature by the addition of liquid nitrogen. This technique was limited to a temperature of -168°, the freezing point of iso-pentane. Only one temperature lower than -168°C was employed viz., -196°C the boiling point of liquid nitrogen. Temperature measurement was made using a pentane-in-glass thermometer reading to -200°C and accurate to 0.2°C. The Charpy V-notch specimens and the ends of the tongs used for transferring them to the impact testing machine were immersed for 15 minutes in the solution and held within $\pm 1^\circ\text{C}$ of the required temperature for the 5 minutes prior to transfer and testing. For temperatures above ambient the specimens and tong ends were heated in a controlled temperature water bath and held within $\pm 1^\circ\text{C}$ of the required temperature for the last 5 minutes

of a 15 minute soak period. When ready for testing the specimen was transferred to the impact testing machine and fractured immediately. The temperature error due to transfer was obtained by attaching a thermocouple to representative specimens and recording the variation of the specimen surface temperature with time from removal from the liquid bath at both $+40^{\circ}\text{C}$ and -120°C . In both instances the transfer time was of the order of 4 seconds and the maximum temperature variation, before fracture, was found to be 2°C .

Modified No. 13 Hounsfield tensile test specimens with a 0.3 in. gauge length were prepared from the simulated blanks with the axis of specimen and blank coinciding and the centre of the gauge length accurately positioned in the transverse plane of the pre-existing control thermocouple. The specimens were tested on a standard Instron tensile testing machine using a 5000 lb. load range and a strain rate of $0.05 \text{ in. min}^{-1}$. The criterion of transverse reduction of area at the point of fracture was used to estimate ductility since the only guide to elongation using a standard Hounsfield measuring gauge was related to the standard 0.5 in. gauge length of a No. 13 specimen. Such measurement produced somewhat erroneous values for tensile elongation.

All hardness tests were carried out on a Zwick hardness testing machine, model 23.2A, using a diamond pyramid indenter. The length of the simulated zone on either side of the control thermocouple was determined by means of metallographic examination and hardness traverse after the blank was ground to 3 mm. below the original surface. A transverse hardness survey and metallographic examination were carried out on sections taken through the control thermocouple position on the blank surface. In both cases a 5 kg. load was used. 12 random hardness determinations were made on each transverse specimen in order to obtain a mean value and standard deviation. In the case of the bead-on-plate weld, hardness traverses were made across the heat affected zone from the fusion boundary to the unaffected parent plate using both 5 kg. and 0.5 kg. loads. Banding present in the plate was investigated, by means of band width measurement and 0.5 kg. load transverse hardness surveys, in specimens which had been simulated using the 930°C peak temperature cycle. Specimens in this simulated condition were chosen since, after etching, they gave the optimum inter-band optical contrast. Band width was measured using the microscope and vernier attachment on the hardness testing machine.

3.7. METALLOGRAPHIC EXAMINATION

Optical microscopy was carried out using a Reichert 'MeF' projection microscope the illumination being provided by a high

pressure mercury vapour lamp. A Siemens Elmiskop 1A electron microscope was used for electron metallography.

A section of the bead-on-plate weld, taken normal to the weld centre line, was used for optical metallographic examination of the heat affected zone. Sections were obtained for each single and duplex simulated thermal cycle condition, both with and without PCHT, by transverse sectioning, of appropriate simulated blanks, through the control thermocouple position. In each instance specimens were prepared in the normal manner for ferrous metallographic examination i.e., by mounting in bakelite, wet grinding on graded silicon carbide papers and finally polishing on rotating 'Metron' cloth pads using 6, 1 and 0.25 micron diamond dust pastes.

All sections were successfully etched using 2% nital solution for 10 - 40 seconds depending upon the structures involved. Prior to the preparation of carbon extraction replicas, for electron metallography, it was found necessary in some instances to prolong the etching period to about 1 minute 10 seconds to ensure the production of a satisfactory replica. A 'modified' super picral etchant containing 3 gm. picric acid and 3.5 ml. of hydrochloric acid in 300 ml. of ethyl alcohol was found to be useful in revealing the nature of carbide dispersion in some PCHT specimens.

Optical microscopical examination was found to be satisfactory in defining distinct structural regions within the heat affected zone of the bead-on-plate weld section but could not resolve the fine structural detail present, even at the highest magnifications attainable. The same problem existed when the simulated blank sections were examined and except for some isolated instances and for grain size determination structural examination was carried out exclusively using the much higher resolution carbon extraction replica technique in the electron microscope.

Carbon extraction replicas (248) were prepared by initially etching the section, as detailed above, then evaporating carbon onto the specimen surface in vacuo. In the majority of cases direct evaporation was used but where difficulty was encountered in structural identification the technique of self-shadowing was employed. This latter technique was also useful in differentiating between 'debris' produced during extraction and 'in-situ' carbide particles in the replica. Upon removal from the evaporator the carboned surface was scribed into small sections of about 3-4 mm² in surface area. The specimen was then immersed in 10% nital solution for approximately 8 minutes by which time the replica sections had either floated off the specimen or begun to lift and blister at the edges on the specimen surface. The specimen was transferred to a bath of 5 parts ethyl alcohol and 85 parts distilled water and the surface tension of this bath

was sufficient to pull off any replica sections remaining on the specimen surface. Replica sections which were freed in the 10% nital solution were transferred directly to another bath containing the same proportions of ethyl alcohol and water. All the replica sections were then transferred for 'washing', using small copper grids to accomplish the transfer, to another bath containing 3 parts ethyl alcohol and 87 parts distilled water, the difference in surface tension between this and the first bath being sufficient to straighten any tightly curled replicas. After 15 minutes in this bath further 'washing' was carried out for 20 minutes, in another bath of the same composition, to ensure as far as possible maximum cleanliness and freedom from 'debris' in the replica sections. Finally, the replica sections were collected on circular copper grids and dried by touching a grid edge onto a piece of filter paper. In some instances this procedure assisted in straightening curled replicas.

Electron probe microanalysis, using a Cambridge Microscan model 2A, was carried out at the Welding Institute to determine the nature of microsegregation within the HY-80 steel plate.

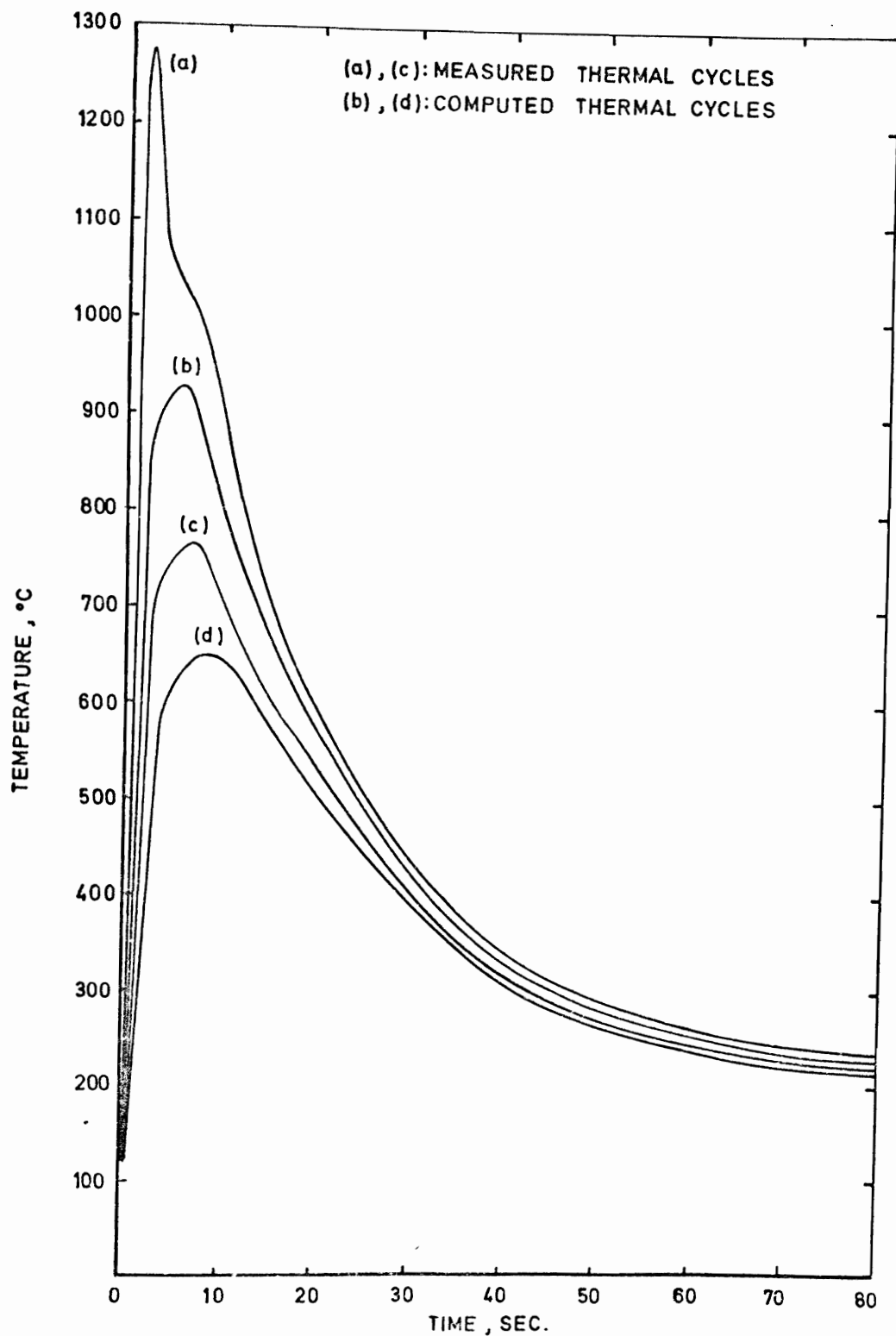


FIGURE 15. THERMAL CYCLES PRODUCED IN THE PARENT PLATE ADJACENT TO THE WELD, ((a) AND (c)), AND THOSE PRODUCED BY COMPUTATIONAL TECHNIQUES, ((b) AND (d)).

4. RESULTS

4.1 COMPUTED AND MEASURED THERMAL CYCLES

Computer program VARTHERM 5 which was used to obtain representative weld heat affected zone thermal cycles is given on pages 195 and 196 . 'Print outs' for the 650 and 930°C peak temperature cycles are contained in pages 208 and 210 respectively.

The computed 650 and 930°C, and the measured, 1275 and 765°C, peak temperature thermal cycles which formed the basis for the simulation phase of this study are shown in figure 15.

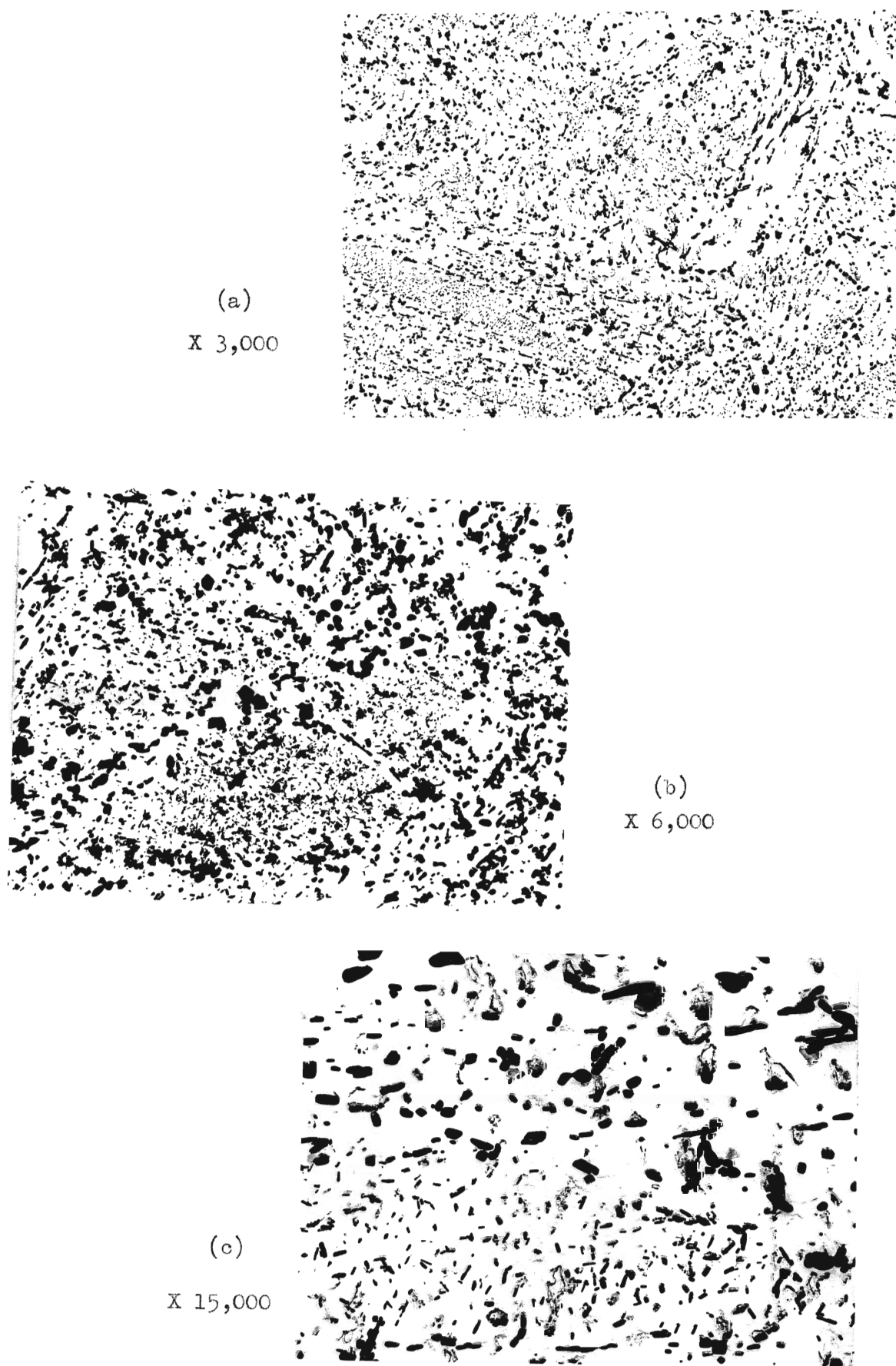


Figure 16. Parent plate microstructure. Carbon extraction replicas.

4.2. MICROSTRUCTURE AND HARDNESS VARIATION IN THE WELD HEAT AFFECTED ZONE

The parent plate microstructure, figure 16, was typical of a quenched and tempered low alloy steel in that it consisted of a carbide dispersion in a ferrite matrix. Two distinct carbide types were present viz., a Widmanstätten array of rod shaped particles sited within the ferrite sub-boundaries and spheroidised carbides situated mainly at high angle ferrite boundaries. Carbide identification was not carried out in this present work but DOLBY (112) has stated that the carbides in tempered HY-80 steel are cementite. The prior austenite grain size in the parent material was not clearly defined due to the difficulty in differentiating between prior austenite and high angle ferrite boundaries.

The alteration in parent plate microstructure due to the bead-on-plate weld, with distance from the fusion boundary, is shown in panoramic form in figure 17. As indicated in this figure the heat affected zone may be considered to consist of 3 virtually distinct regions viz., grain coarsened, grain refined and intercritical, extending from 0 - 0.6 mm., 0.6 - 2.6 mm. and 2.6 - 2.8 mm., from the fusion boundary respectively. Photomicrographs of structures typical of these regions are also presented in figure 17.

Due to the high temperatures experienced within the grain coarsened region rapid austenite grain growth occurred during thermal cycling but the austenite grain size decreased with decreasing peak temperature as distance from the fusion boundary increased. Within the prior austenite grains the structure was acicular but its actual character could not be determined by optical microscopy due to the inherent resolution limitations of this technique. The boundary separating the grain coarsened and grain refined regions is rather diffuse. However, with increasing distance from the fusion line, away from this boundary, the structure consisted of a fine irresolvable ferrite-carbide aggregate and showed a fine decreasing prior austenite grain size. The intercritical region contained an increasing proportion of pre-transformation ferrite as the distance increased from the fusion boundary towards the unaffected structure of the parent plate. The transformed regions appeared to consist of a very fine carbide aggregate but once again these regions could not be resolved using the optical microscope.

There was slight evidence of banding in the tempered structure of the parent plate and sections taken perpendicular to each other and normal to the supposed rolling direction indicated, after prolonged etching in Fry's reagent, the layered structure typical of a cross-rolled product. The primary rolling direction, normal to the plane of the paper in figure 17, had, therefore, to be

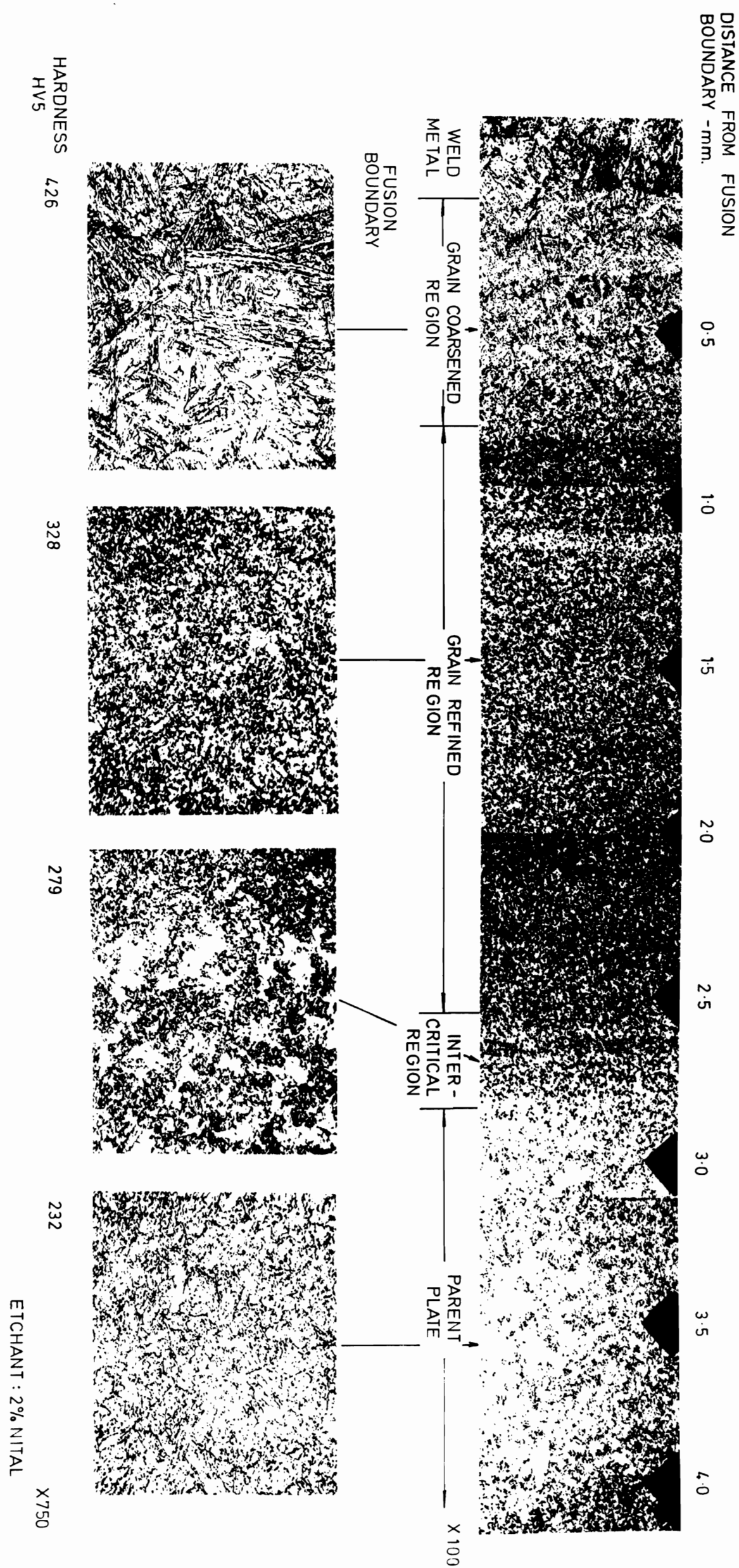


FIGURE 17. THE HEAT AFFECTED ZONE STRUCTURES ASSOCIATED WITH A BEAD-ON-PLATE WELD IN HY-80 STEEL.

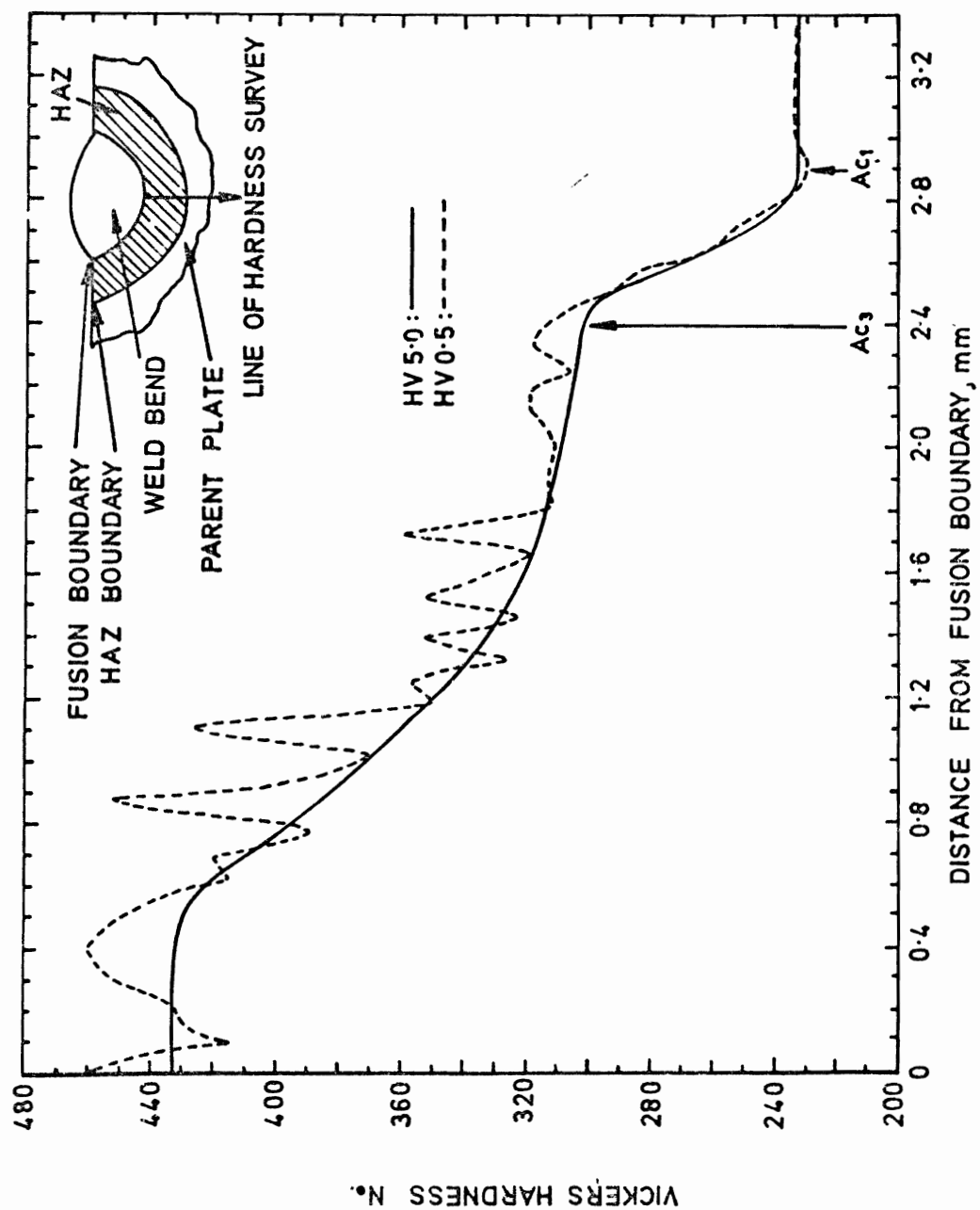


FIGURE 18. HEAT AFFECTED ZONE HARDNESS SURVEY.

determined by means of energy transition curves produced by testing standard, through thickness notched, Charpy impact specimens taken from the transverse and longitudinal plate orientations. Within the weld heat affected zone banding was marked and revealed at short etching times in 2% nital, within the grain refined region.

5kg. and 0.5kg. load hardness traverses were made, figure 18, to indicate, respectively, the general hardness trend across the heat affected zone and to show the hardness variation, due to alloying element segregation, within the banded regions. The presence of banding in the intercritical region and the parent material was not revealed by 0.5kg. load hardness testing although its presence was detected microscopically due to the slight differential etching characteristics of adjacent bands.

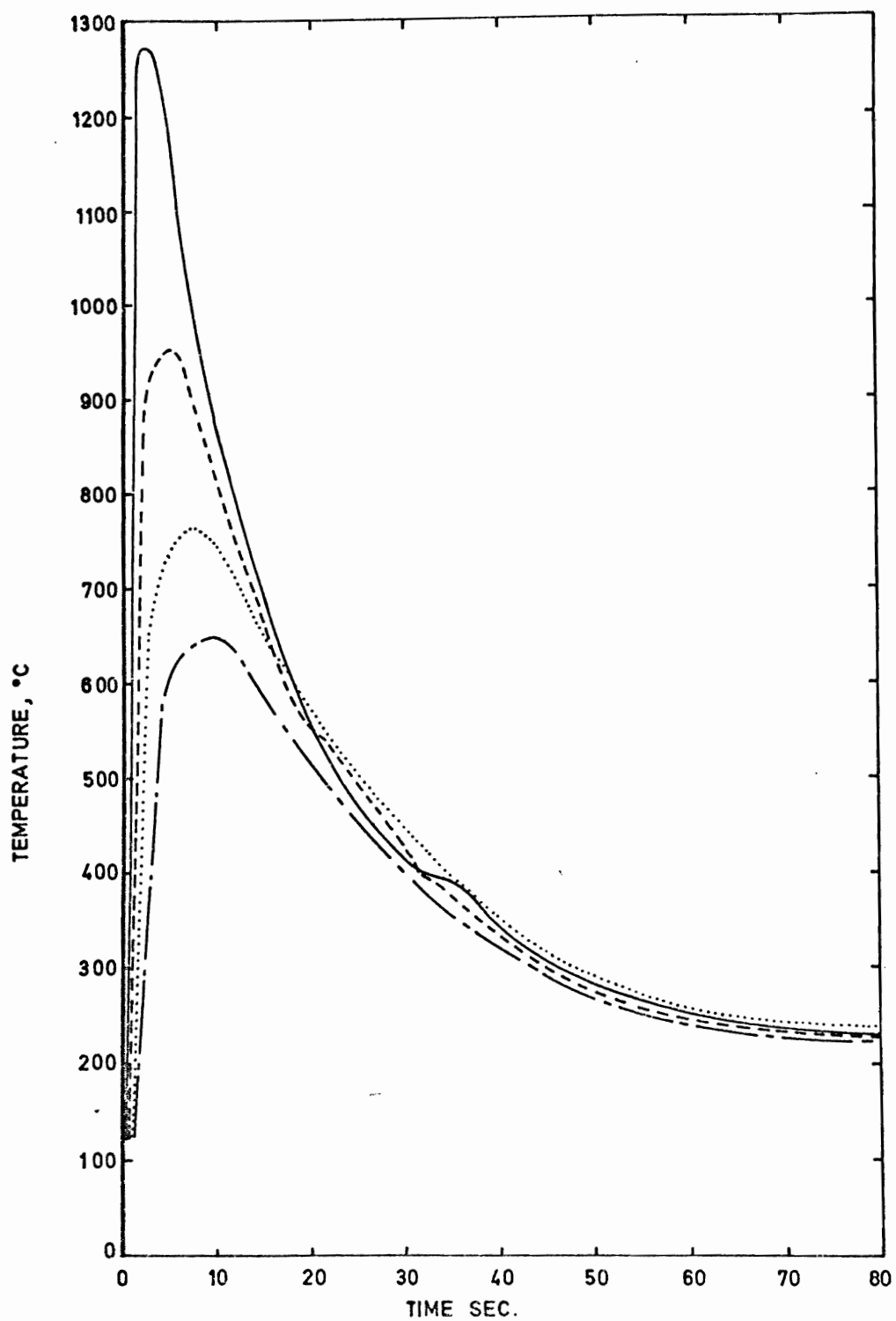


FIGURE 19. SIMULATED THERMAL CYCLES.

4.3. THERMAL CYCLE SIMULATION

Typical thermal cycles produced during simulation are shown in figure 19. The characteristics of the simulator equipment gave some degree of variation for given settings on the peak temperature obtained. For the 1275°C peak temperature thermal cycle a limit of $\pm 25^\circ\text{C}$ was used and specimens which did not have a peak temperature within the range $1275 \pm 25^\circ\text{C}$ were rejected. For 930, 765 and 650°C peak temperatures the limit was $\pm 10^\circ\text{C}$. Reliability of cycle reproduction was very much better for the lower peak temperatures than for the 1275°C peak temperature.

Longitudinal hardness traverses were made on specimens produced with different clamp gap widths to gauge the extent of the uniformly cycled region on either side of the control thermocouple. The results obtained were:

Clamp gap width, in.	Total length of uniformly cycled region, in.
0.5	0.31
0.6	0.38
0.8	0.42
1.0	0.46

4.4 MECHANICAL PROPERTIES OF THE SIMULATED STRUCTURES

The results of Charpy V-notch impact tests on single and double cycled simulated specimens, in both the as-cycled and post cycle heat treated conditions are given in figures 20 to 25. Figures 20 to 23 and 24 and 25 show energy transition and fracture appearance transition curves respectively. Since close correlation was found between lateral expansion measurement and energy absorbed in fracture, as shown by the scatter band in figure 26, ductility transition curves have not been included in this presentation.

Figure 20 includes energy transition curves for both the transverse and longitudinal parent plate directions. These curves were initially employed to establish the primary rolling direction in the plate, and, together with tensile data, to determine if the mechanical properties were in accordance with specification MIL-S-16216G.

The results obtained from tensile and hardness testing together with various Charpy V-notch impact transition criteria are given in Tables 5 and 6, pages 169 and 170 for as-cycled and post cycle heat treated conditions respectively. Data for the parent plate is included in Table 5.

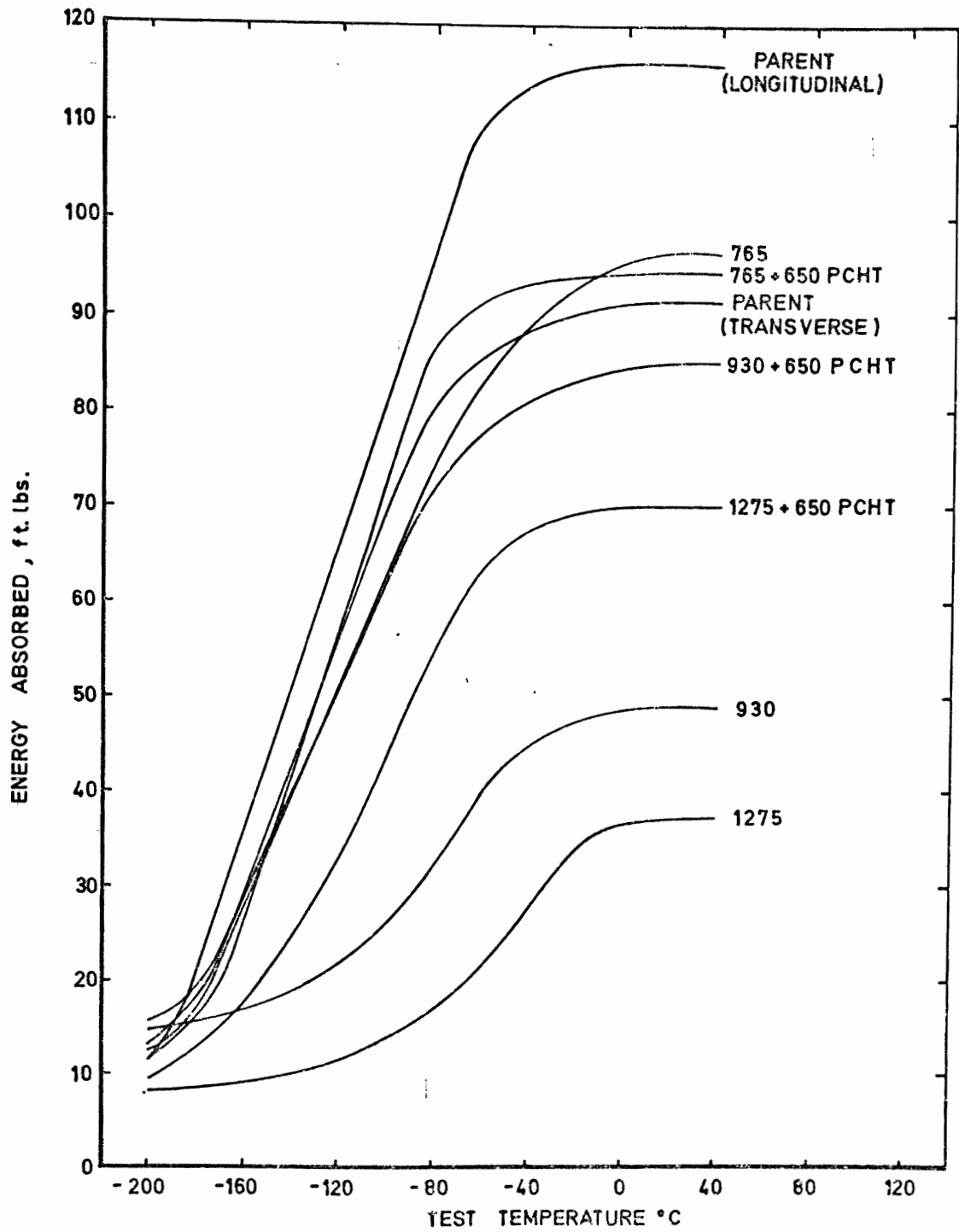


FIGURE 20. EFFECT OF THERMAL CYCLING AND SUBSEQUENT POST-CYCLE HEAT TREATMENT (PCHT) ON CHARPY V-NOTCH IMPACT-TEMPERATURE DATA FOR HY60 SIMULATED SPECIMENS.

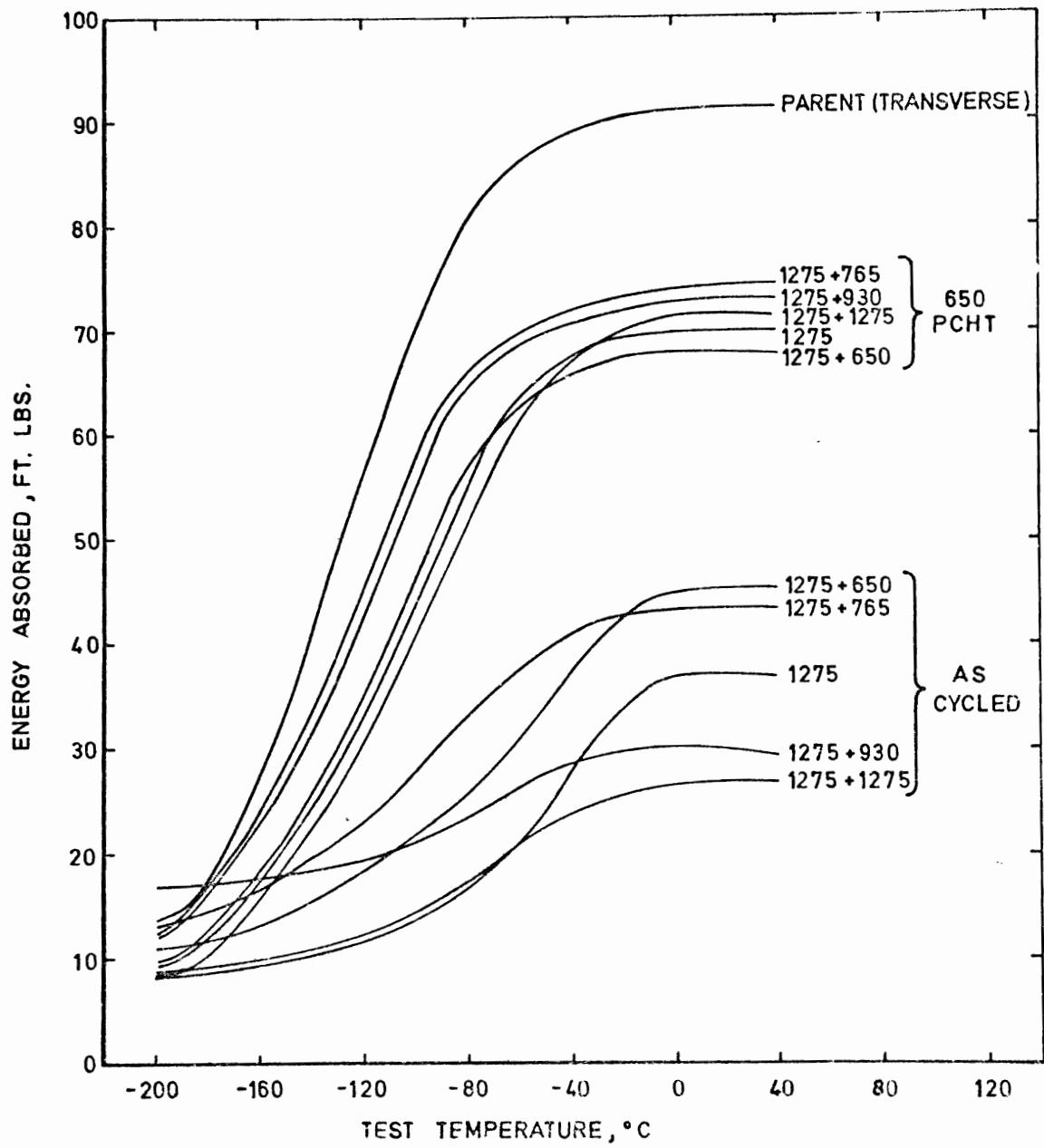


FIGURE 21. EFFECT OF SUBSEQUENT THERMAL CYCLING AND POST - CYCLE HEAT TREATMENT (PCHT) ON CHARPY V-NOTCH IMPACT - TEMPERATURE DATA FOR HY80 SIMULATED SPECIMENS, INITIALLY CYCLED TO A PEAK TEMPERATURE OF 1275°C.

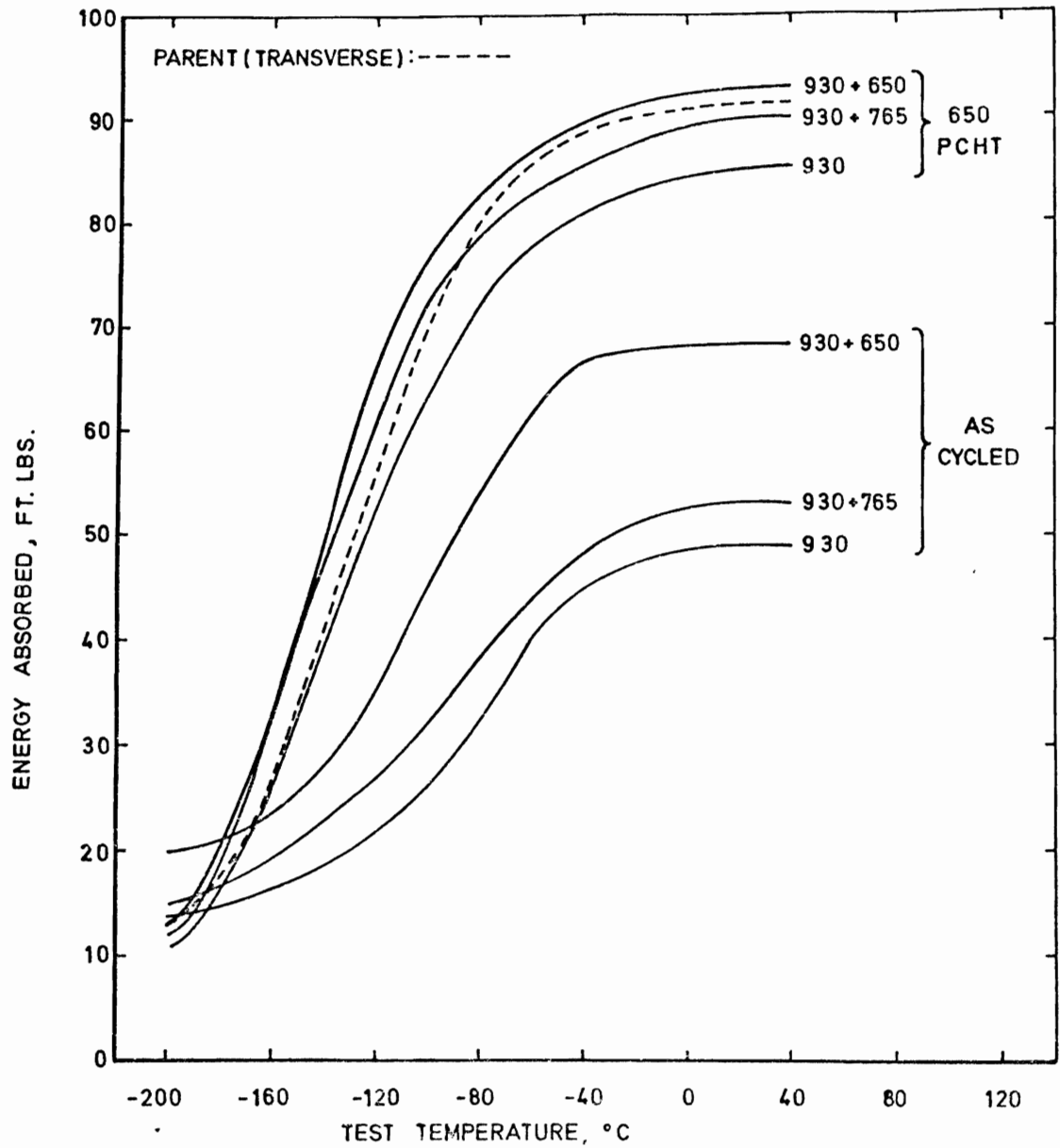


FIGURE 22. EFFECT OF SUBSEQUENT THERMAL CYCLING AND POST-CYCLE HEAT TREATMENT (PCHT) ON CHARPY V-NOTCH IMPACT-TEMPERATURE DATA FOR HY80 SIMULATED SPECIMENS, INITIALLY CYCLED TO A PEAK TEMPERATURE OF 930°C.

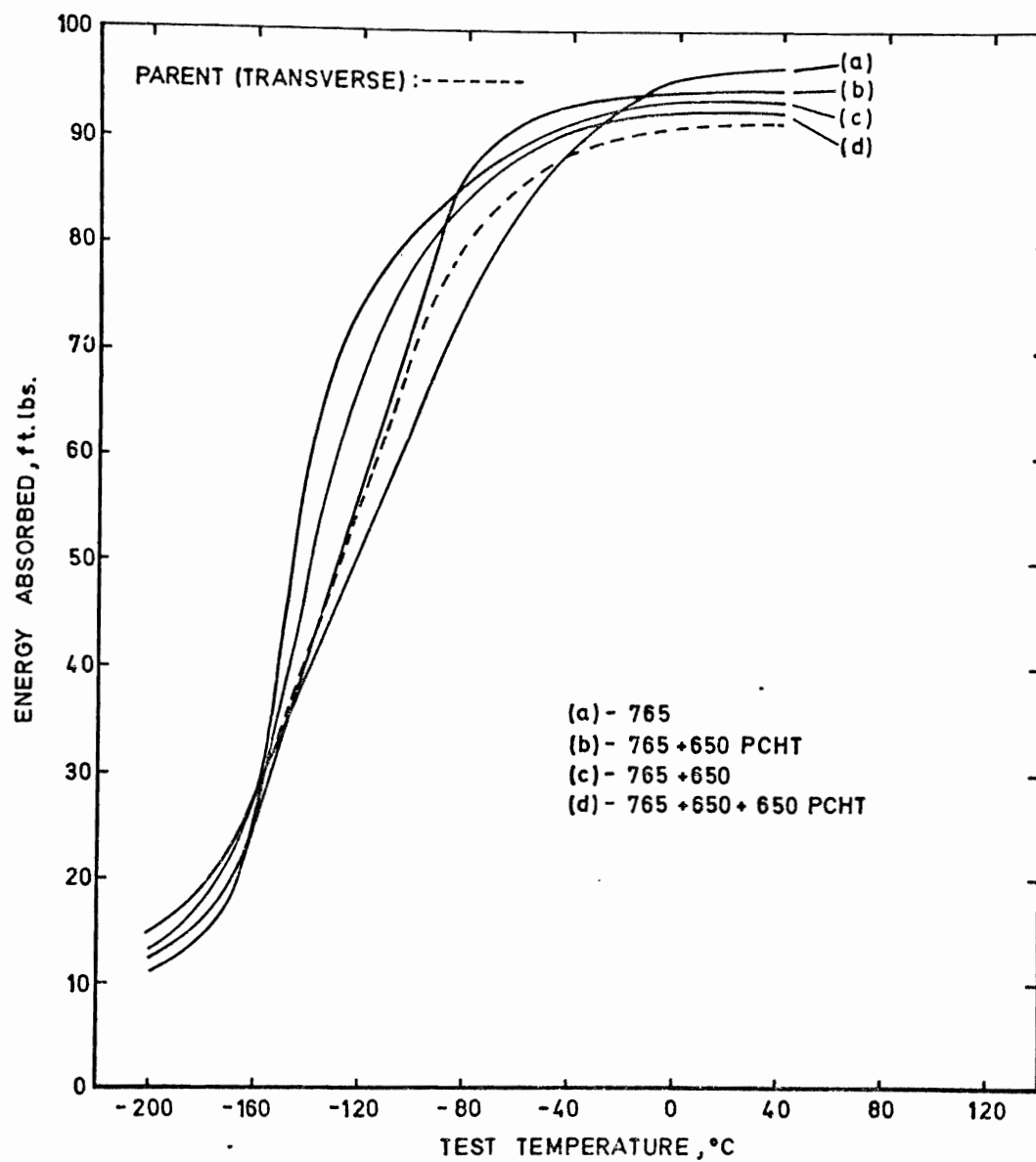


FIGURE 23. EFFECT OF SUBSEQUENT THERMAL CYCLING AND POST-CYCLE HEAT TREATMENT (PCHT) ON CHARPY V-NOTCH IMPACT - TEMPERATURE DATA FOR HY80 SIMULATED SPECIMENS, INITIALLY CYCLED TO A PEAK TEMPERATURE OF 765°C.

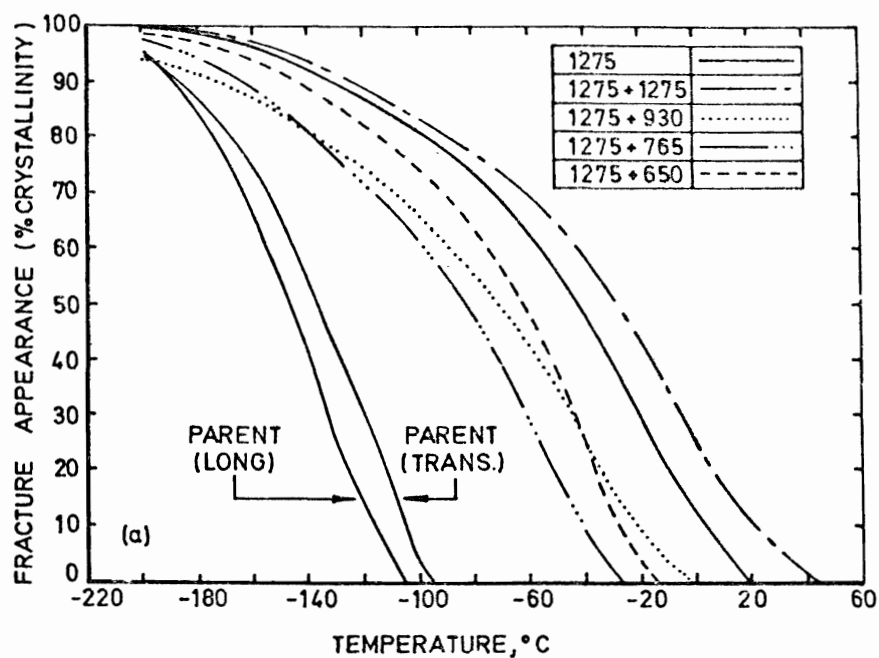
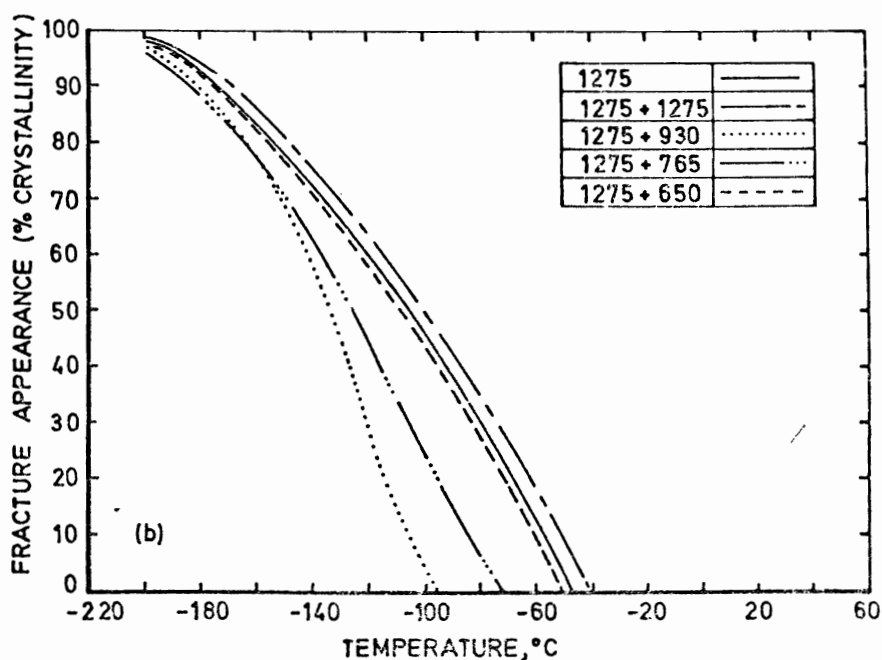


FIGURE 24. CHARPY V-NOTCH IMPACT (PERCENTAGE CRYSTALLINITY) - TEMPERATURE CURVES FOR HY80 SIMULATED SPECIMENS. INITIAL PEAK TEMPERATURE 1275°C (a) AS CYCLED. (b) AS (a) BUT POST CYCLE HEAT TREATED (1HR. AT 650°C).

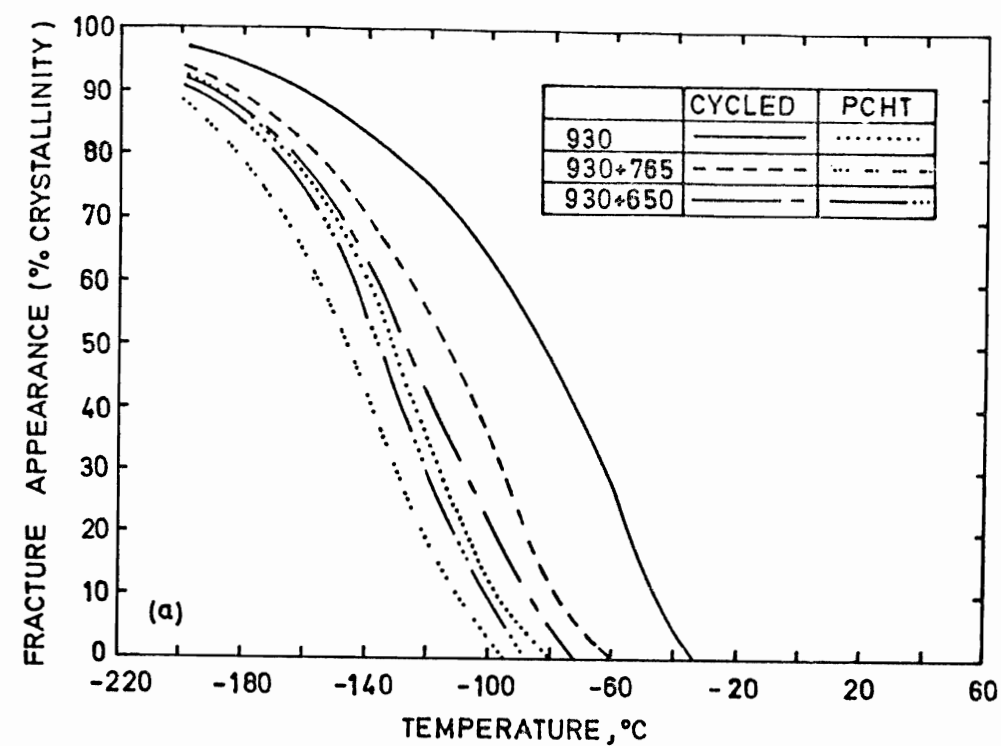
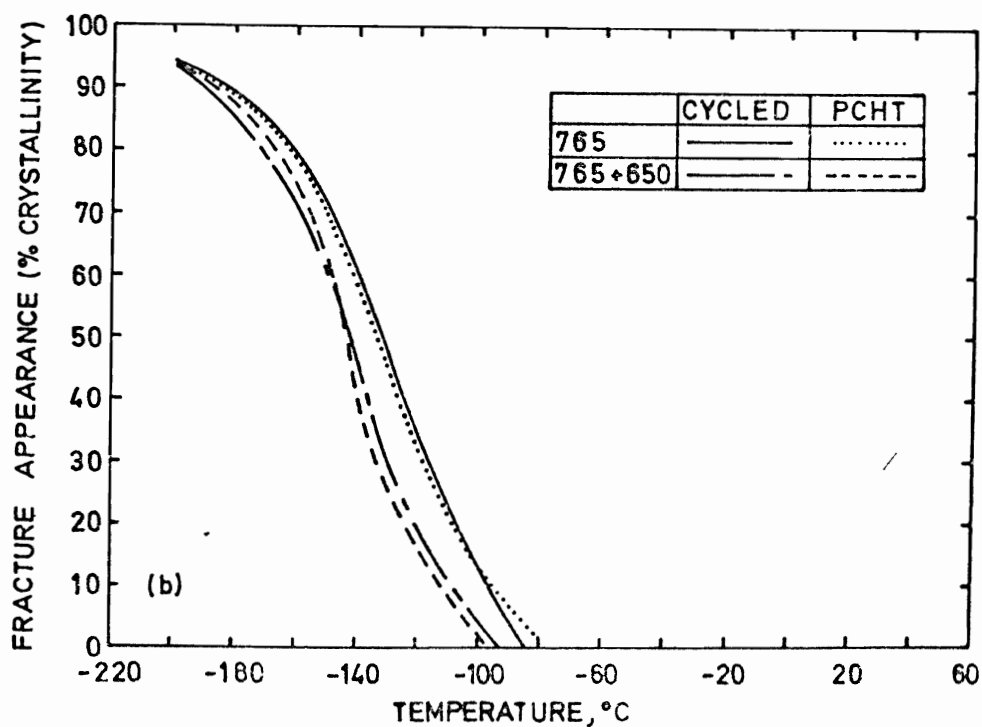


FIGURE 25. CHARPY V-NOTCH IMPACT (PERCENTAGE CRYSTALLINITY) - TEMPERATURE CURVES FOR HY80 SIMULATED SPECIMENS WITH INITIAL PEAK TEMPERATURES OF (a) 930°C AND (b) 765°C. POST CYCLE HEAT TREATMENT, PCHT, (1HR. AT 650°C) AS INDICATED.

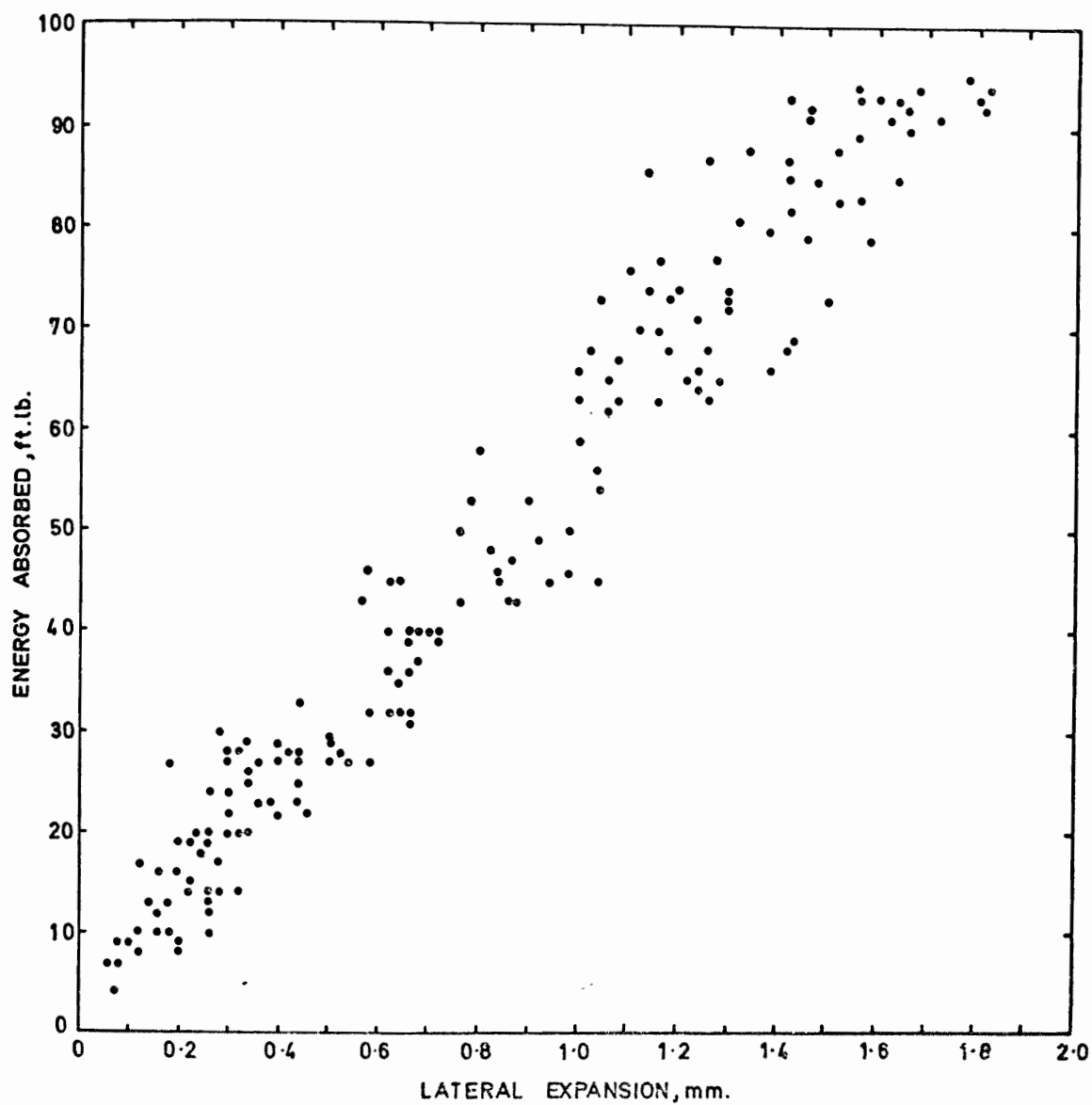


FIGURE 26, CHARPY V-NOTCH IMPACT TEST. ENERGY-EXPANSION RELATIONSHIP FOR A RANGE OF CYCLED AND POST CYCLE HEAT TREATED HY80 SIMULATED SPECIMENS.

4.5. METALLOGRAPHY OF THE SIMULATED STRUCTURES

In the two sections which follow, simulated thermal cycles are designated solely by the peak temperature attained e.g. 930°C refers to a single cycle having a peak temperature of 930°C whereas 930/765°C indicates a double cycle with initial and second cycle peak temperatures of 930°C and 765°C respectively. The structures are considered as two groups viz., as-cycled and post cycle heat treated.

4.5.1. As-cycled structures

1. 1275°C (Figure 27)

This structure consisted mainly of auto-tempered lath martensite (a) in which a very fine carbide dispersion was present within each lath. Some regions of upper and mixed bainites (b) were also present.

2. 1275/1275°C (Figure 28)

The structure produced by this treatment was similar to that described in 1 above in that it consisted predominantly of auto-tempered lath martensite, (b). However, some regions containing well defined upper bainitic colonies were present, (a), which were composed of long ferrite laths separated by stringer-like carbides aligned in one direction and extending throughout a large proportion of the prior austenite grain in which they were sited. The prior austenite grain size was somewhat larger than in 1.

3. 1275/930°C (Figure 29)

This structure is rather difficult to define. It was composed mainly of auto-tempered lath martensite (b) in which the laths were very much finer and less regular in shape than in either structures 1 or 2. As is evident in (a), the general character of the structure produced by the initial 1275°C cycle has been retained and in some regions transformation occurred between pre-existing bainitic ferrite laths leaving them narrower in width and separated by auto-tempered martensite. Some regions of predominantly upper bainite but containing small areas of lower bainite were evident, (c). As regards prior austenite grain size a duplex character was present in that there was evidence of a large boundary system due to the initial cycle and a much finer distribution within this due to the second cycle.

4. 1275/765°C (Figure 30)

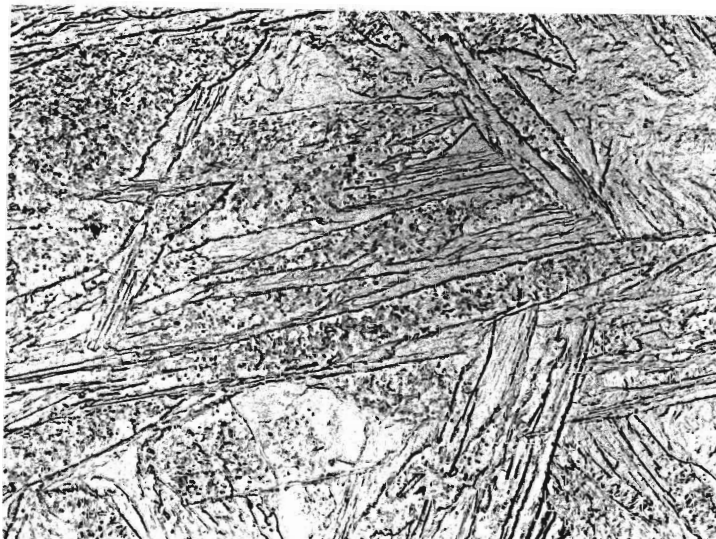
Optical examination showed a virtually continuous constituent delineating the prior austenite grain boundaries due to the initial

cycle, (a). Electron microscopy showed this constituent to be composed of small auto-tempered martensite units, (c). The areas surrounded by the auto-tempered martensite network consisted of internally veined ferrite and auto-tempered martensite which had a lath-type morphology in many regions, (b).

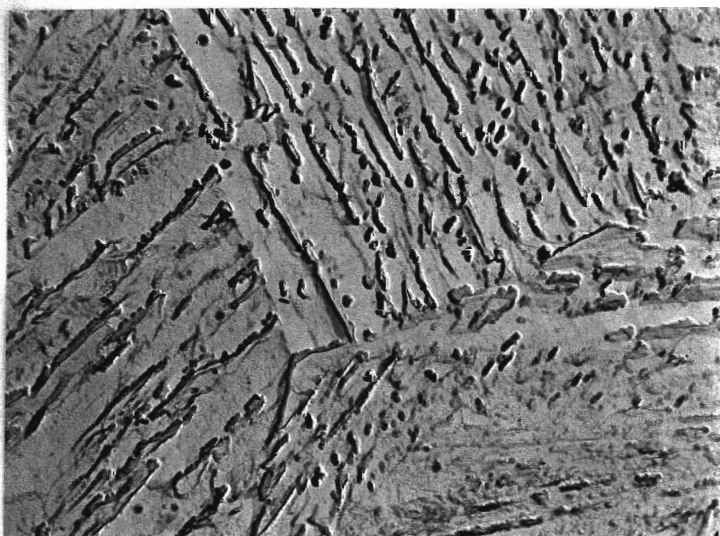
5. 1275/650°C (Figure 31)

The second cycle had a tempering effect on the structure produced by the initial cycle and caused carbide precipitation at the prior austenite grain boundaries, (c), and in their interior in pre-existing auto-tempered martensite regions (b). Precipitation was not very advanced and in many regions was directional in nature with discontinuous carbide stringers separating adjacent elongated ferrite grains, (a).

(a)
X 3,000



(b)
X 6,000



(c)
X 6,000

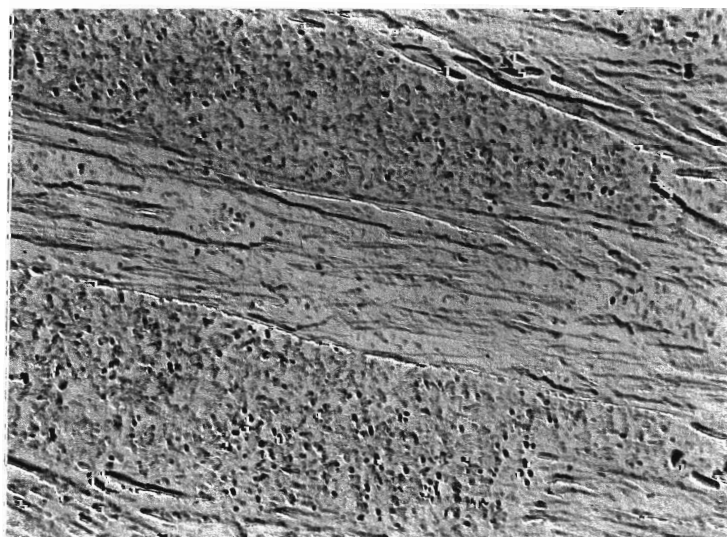
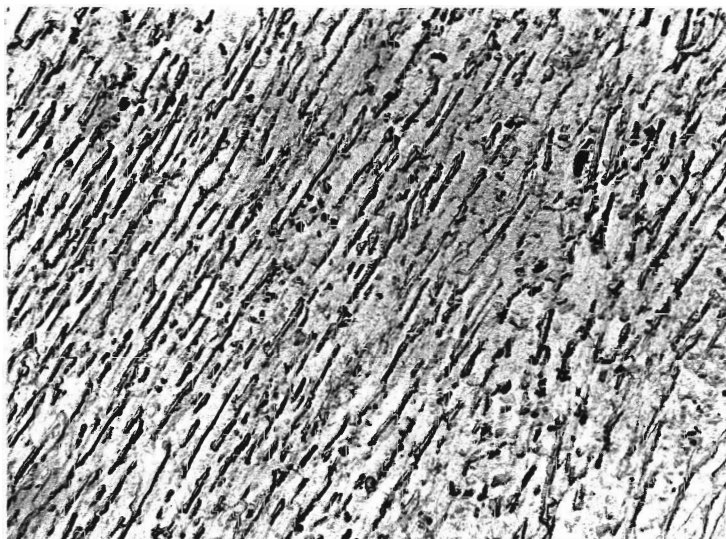
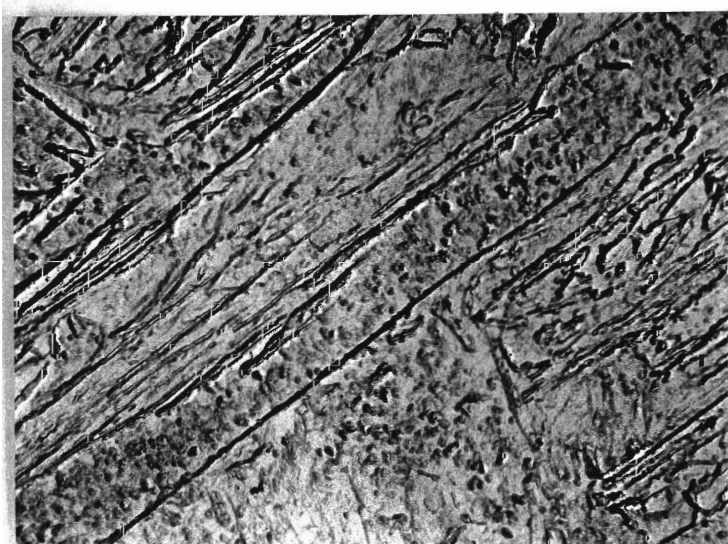


Figure 27. Microstructure produced by thermal cycling to a peak temperature of 1275°C. Carbon extraction replicas.

(a)
X 6,000



(b)
X 6,000

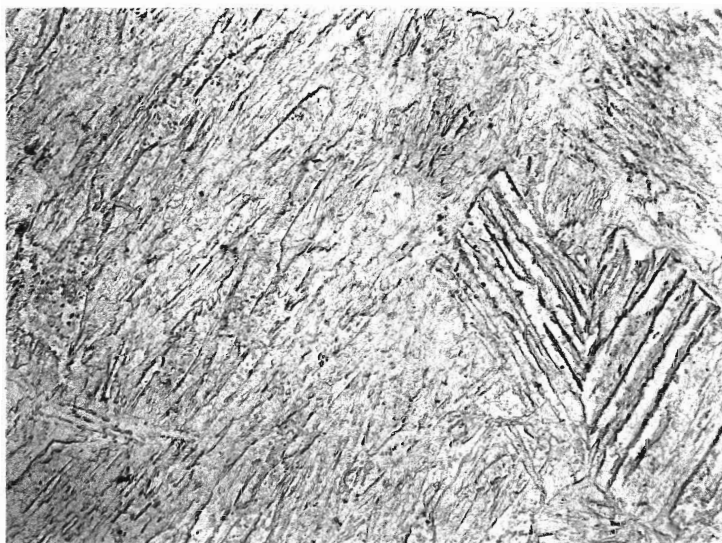


(c)
X 6,000

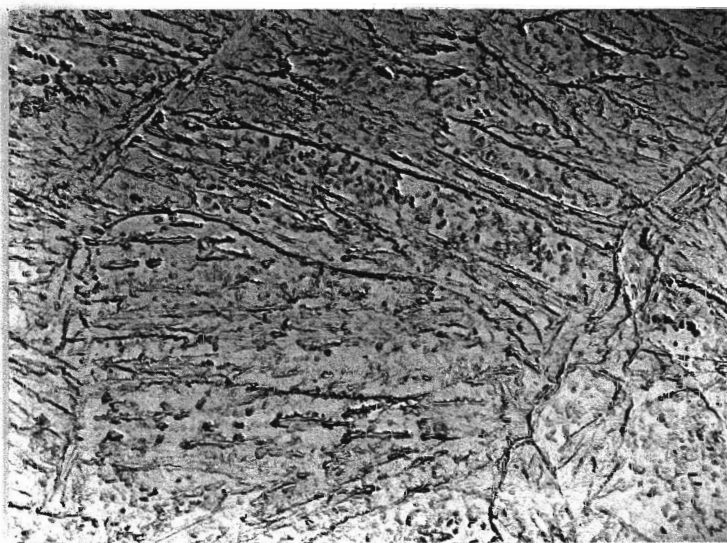


Figure 28. Microstructure produced by double thermal cycling to a peak temperature of 1275°C. Carbon extraction replicas.

(a)
X 3,000



(b)
X 6,000



(c)
X 6,000

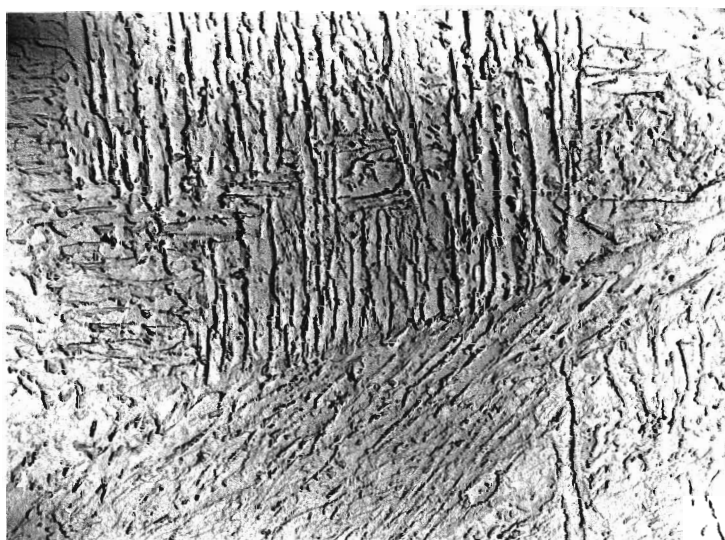
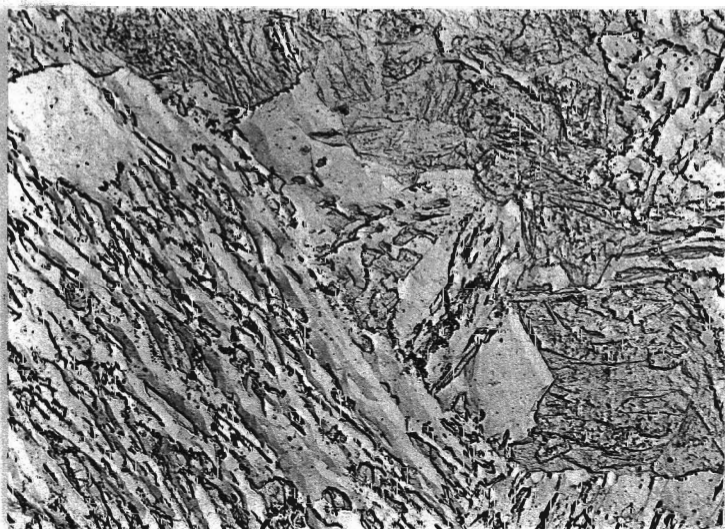


Figure 29. Microstructure produced by double thermal cycling: initial and second peak temperatures of 1275°C and 930°C respectively. Carbon extraction replicas.

(a)
X 800



(b)
X 6,000



(c)
X 6,000

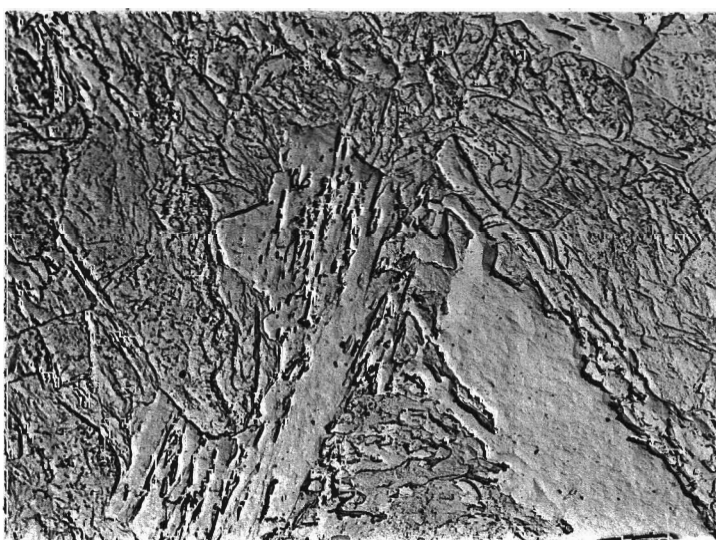
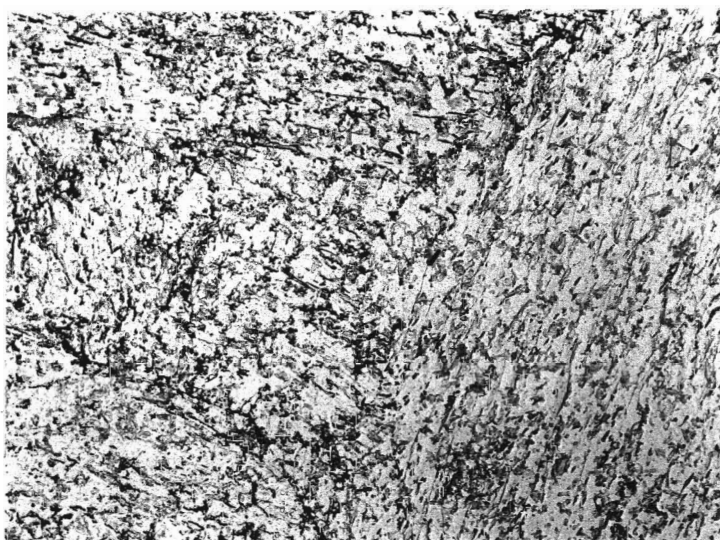
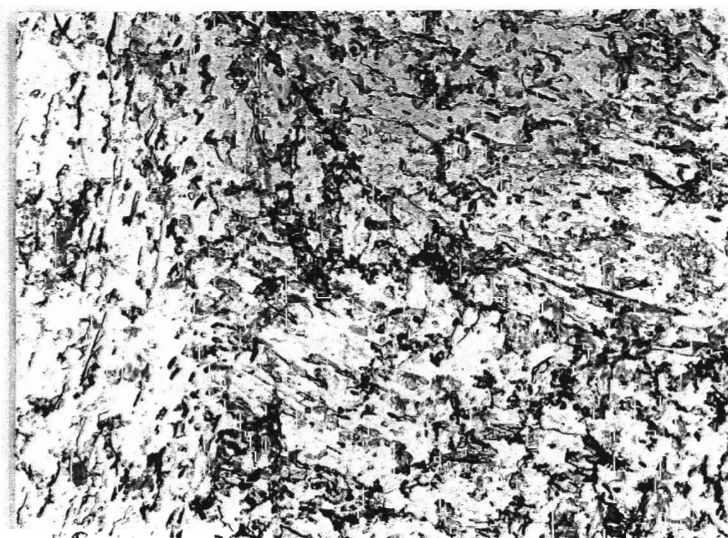


Figure 30. Microstructure produced by double thermal cycling: initial and second peak temperatures of 1275°C and 765°C respectively. Optical micrograph (a); carbon extraction replicas (b) and (c).

(a)
X 3,000



(b)
X 6,000



(c)
X 15,000

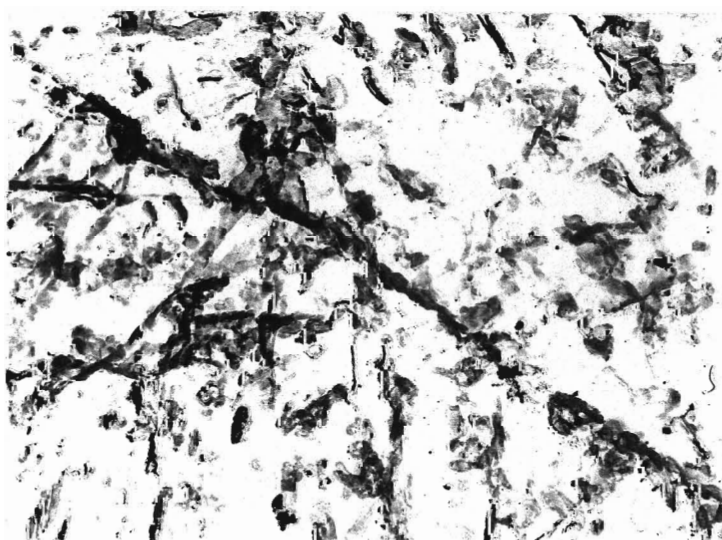


Figure 31. Microstructure produced by double thermal cycling: initial and second peak temperatures of 1275°C and 650°C respectively. Carbon extraction replicas.

6. 930°C (Figure 32)

A very fine ferrite - globular carbide aggregate was produced by this cycling treatment, (a). There was, however evidence (b), (c), of some transformation to auto-tempered martensite within regions bounded by the ferrite-carbide transformation product.

7. 930/765°C (Figure 33)

The second cycle, as in the case of 4, caused only partial austenitization and the resulting structure consisted of fine grained ferrite and regions containing a fine ferrite/carbide dispersion. There was some evidence (c) of only partial carbide solution in the austenite.

8. 930/650°C (Figure 34)

The tempering effect of the 650°C cycle was evidenced by an increased density of carbides when compared to 6. Regions containing high carbide (a) and low carbide (b) densities were apparent and in the former elongated carbides were found to exist (c).

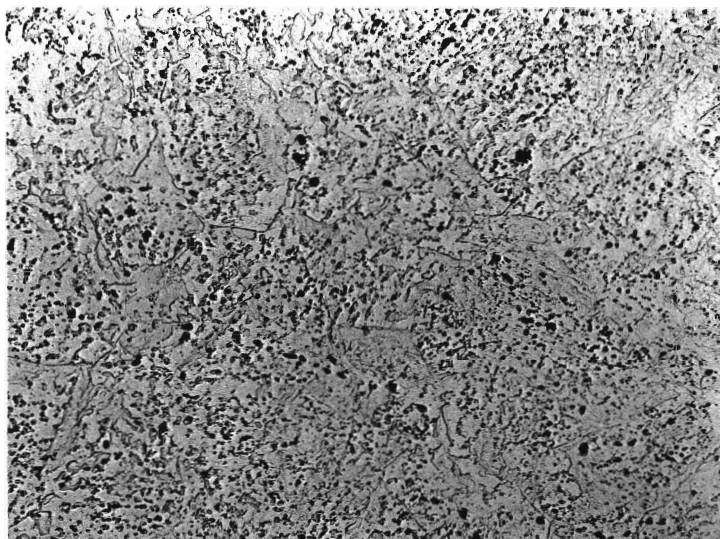
9. 765°C (Figure 35)

The structure was very similar to that of 7 but the density of carbides was much greater and in contrast to 7 many elongated carbides were present.

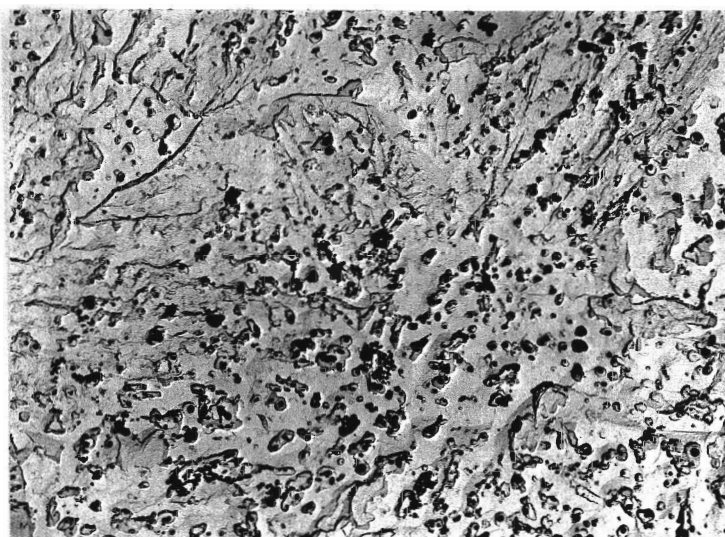
10. 765/650°C (Figure 36)

Fairly marked, general carbide precipitation was caused by the second cycle (a). The structure suggests, by the presence of colonies of small rod shaped carbides, (a), (c), that some auto-tempered martensite had been produced by the initial cycle in regions which had undergone austenitization.

(a)
X 3,000



(b)
X 6,000



(c)
X 6,000

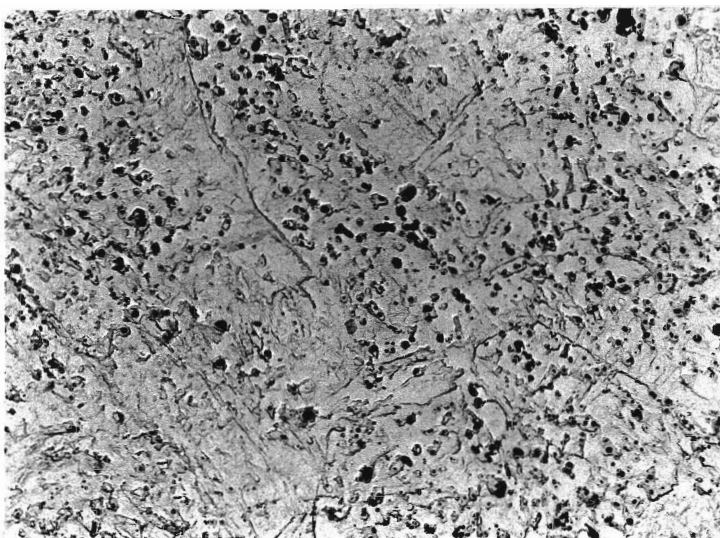
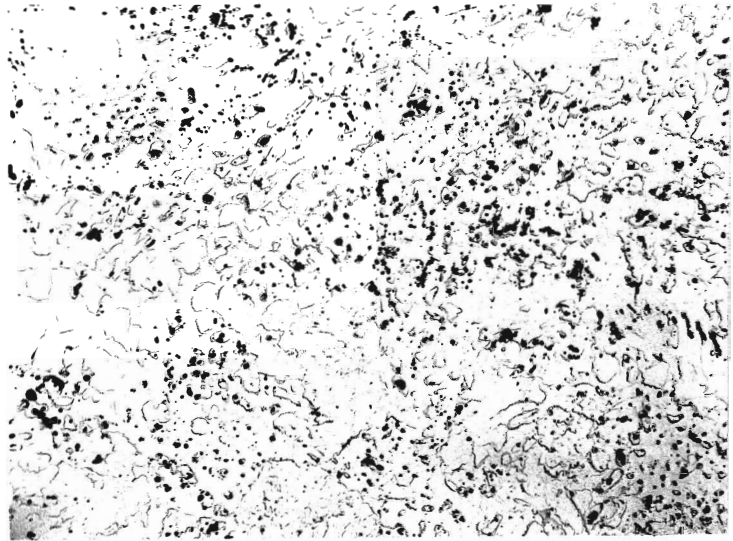
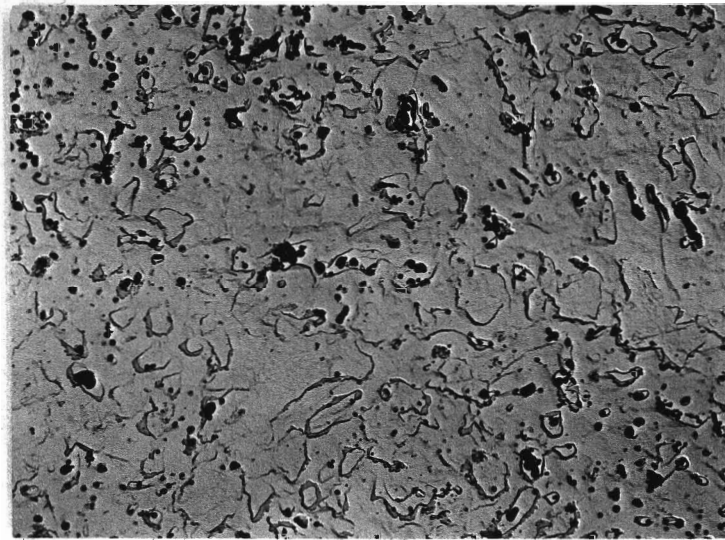


Figure 32. Microstructure produced by thermal cycling to a peak temperature of 930°C. Carbon extraction replicas.

(a)
X 3,000



(b)
X 6,000



(c)
X 15,000



Figure 33. Microstructure produced by double thermal cycling: initial and second peak temperatures of 930°C and 765°C respectively. Carbon extraction replicas.

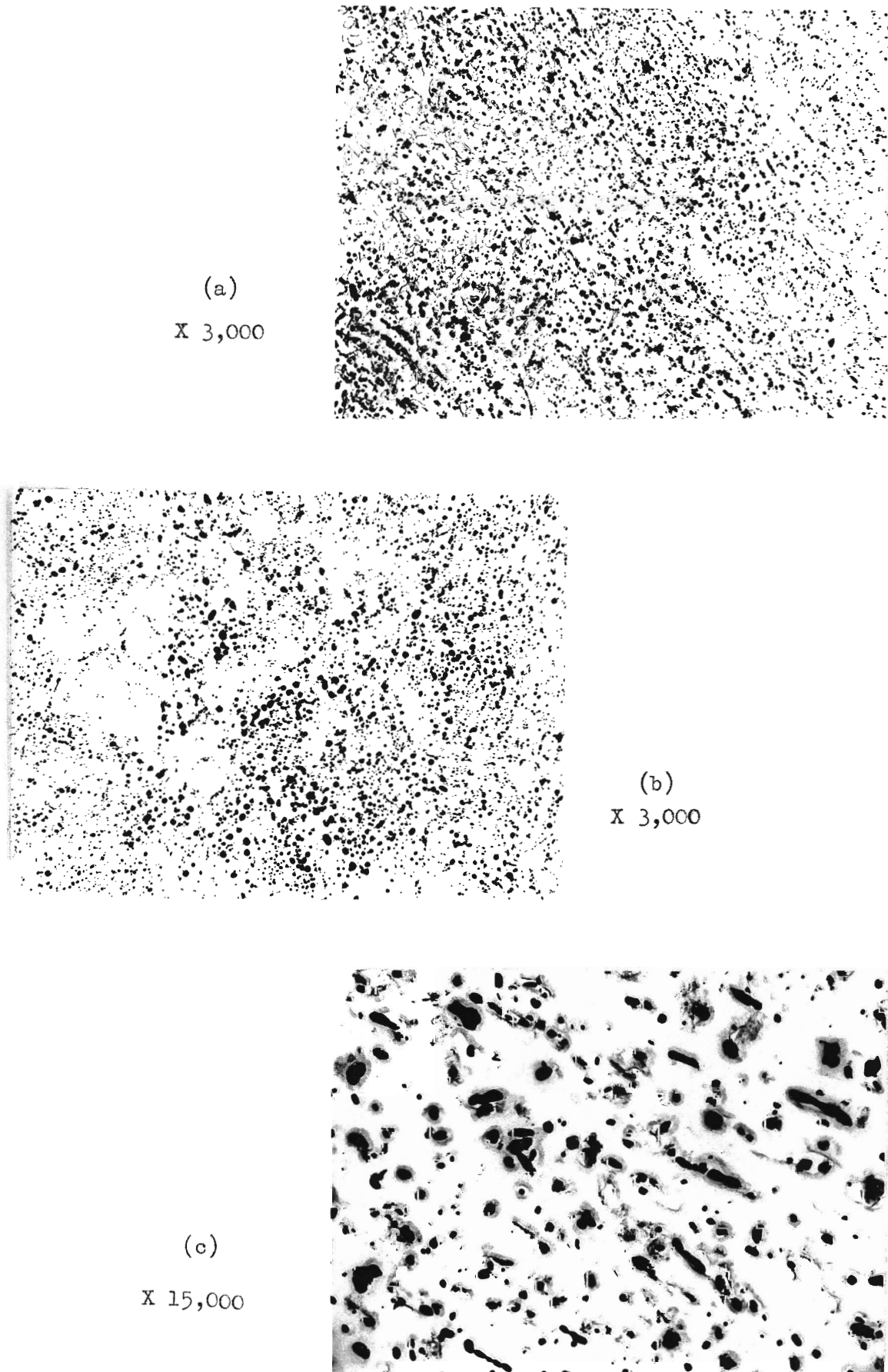


Figure 34. Microstructure produced by double thermal cycling: initial and second peak temperatures of 930°C and 650°C respectively. Carbon extraction replicas.

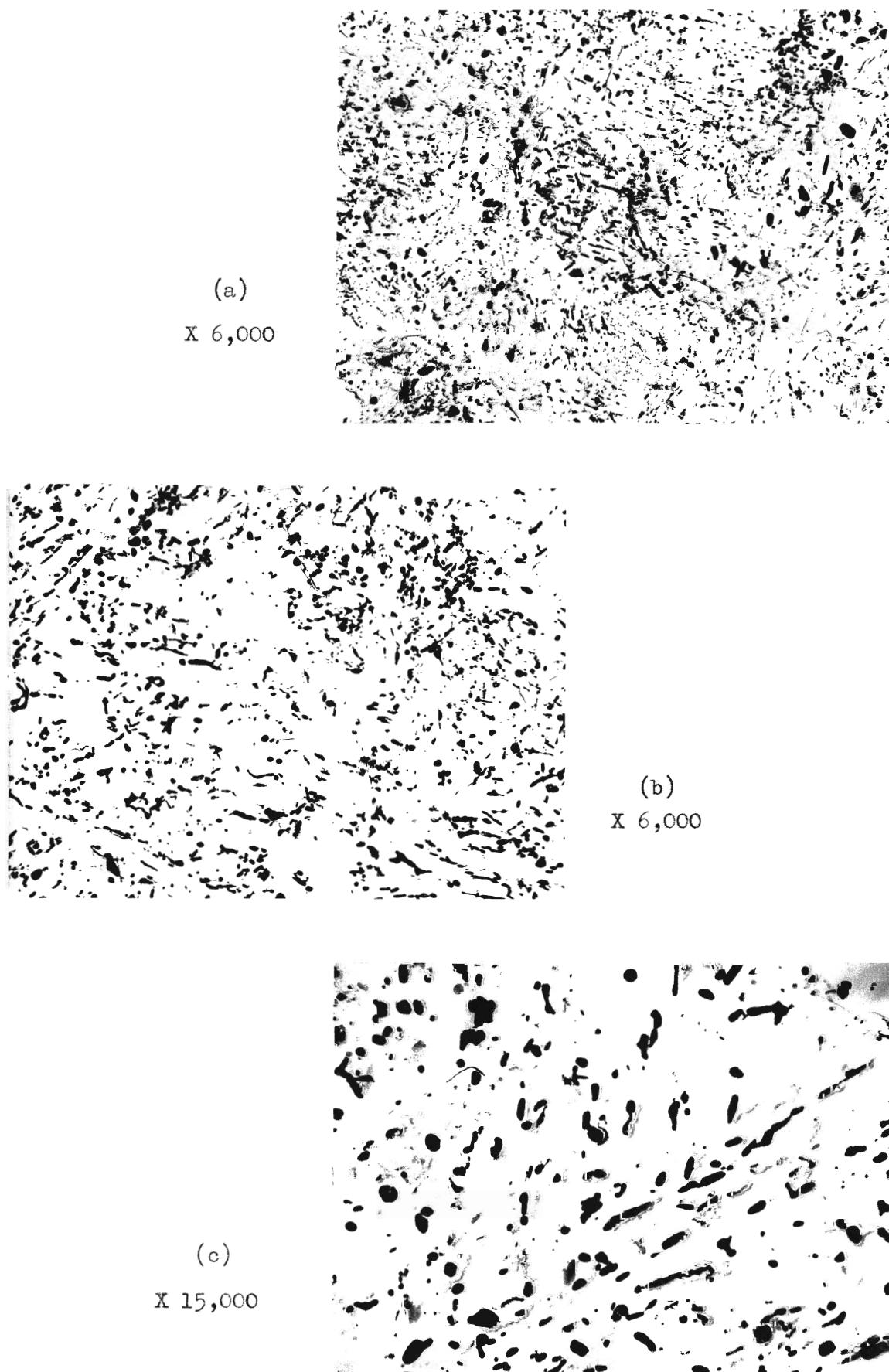


Figure 35. Microstructure produced by thermal cycling to a peak temperature of 765°C. Carbon extraction replicas.

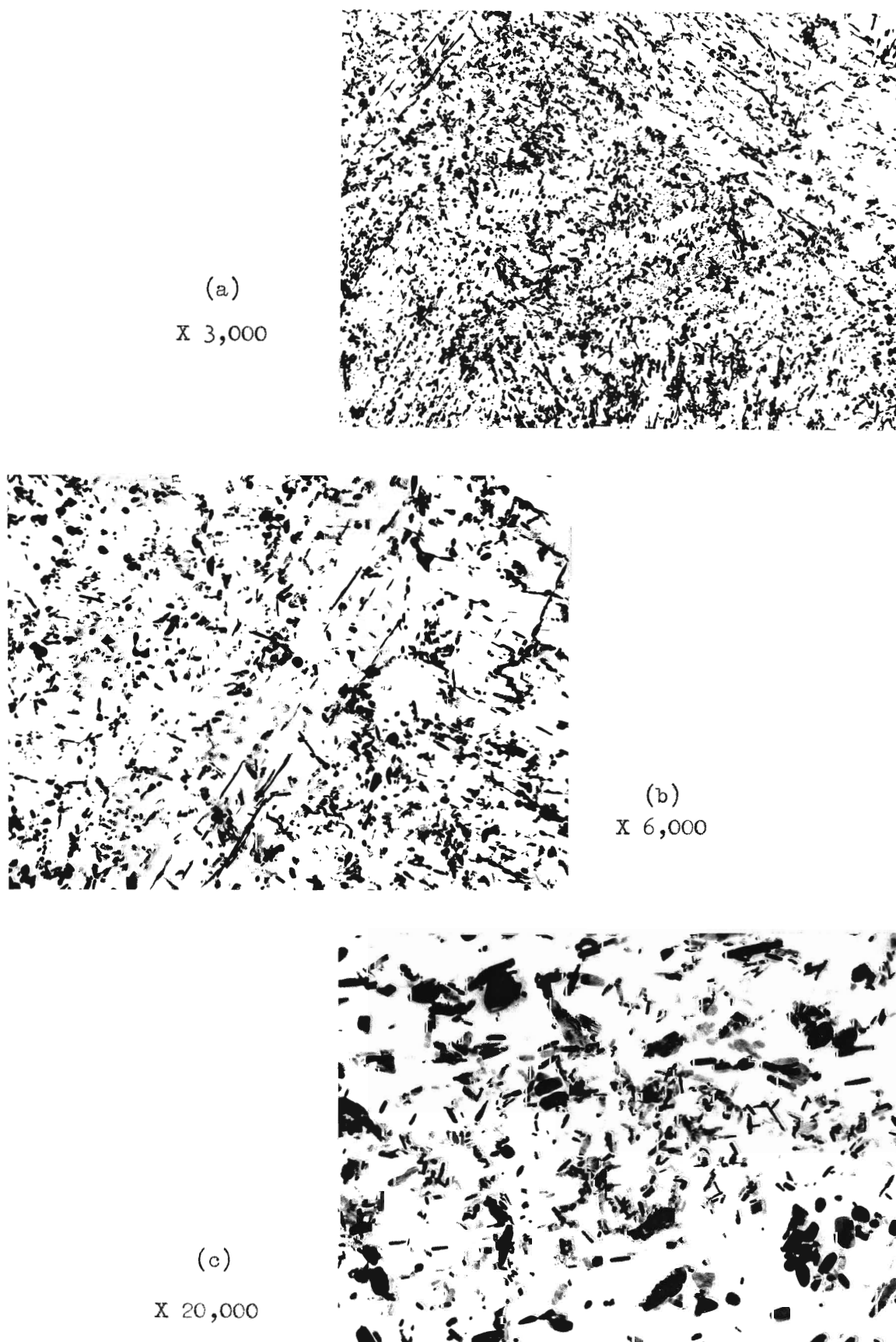


Figure 36. Microstructure produced by double thermal cycling: initial and second peak temperatures of 765 °C and 650°C respectively. Carbon extraction replicas.

4.5.2. Post cycle heat treated structures

Groups of these structures had features which were very similar. To save repetition, therefore, description is, in several instances, related to groups of, rather than individual, cycling treatments.

1. 1275°C, 1275/1275°C, 1275/650°C (Figure 37)

The structure produced by these treatments had the following features:

- (i) quite marked globular and stringer carbide precipitation at prior austenite grain boundaries, (a).
- (ii) within prior auto-tempered martensite laths a very fine Widmanstätten rod shaped carbide array existed (a), (c). Isolated globular carbides were also present within such regions, (c).
- (iii) discontinuous stringer carbides separating individual laths in ferrite lath arrays, (a), (b).

2. 1275/930°C (Figure 38)

The rather complex nature of the duplex prior austenite grain sizes due to this double cycle is defined by carbide precipitation at the prior austenite grain boundaries. Such carbides were mainly globular but also appeared in discontinuous stringer form. This latter carbide was also evident within the prior austenite grains in association with the small rod like precipitate noted in 1. (ii).

3. 1275/765°C (Figure 39)

Optical microscopy revealed a dense carbide network at prior austenite grain boundaries where, in the as-cycled condition, a predominantly auto-tempered martensite structure existed, (a). Electron microscopy revealed that the carbides present within such regions had three forms viz., globular, discontinuous stringer and small rod. The first named of these three forms was found between, the second at the edges and the third within, pre-existing auto-tempered martensite laths, (c). Fine carbide precipitation was also evident within the large lath form of ferrite within the grain boundary network and such laths had heavy carbide precipitation at their boundaries.

4. 930°C, 930/650°C (Figure 40)

The structure showed a dispersion of globular carbides in a ferrite matrix. Some regions showed a high carbide density, (a), (c) whereas in others, regions of ferrite were evident which were virtually devoid of carbide particles, (b).

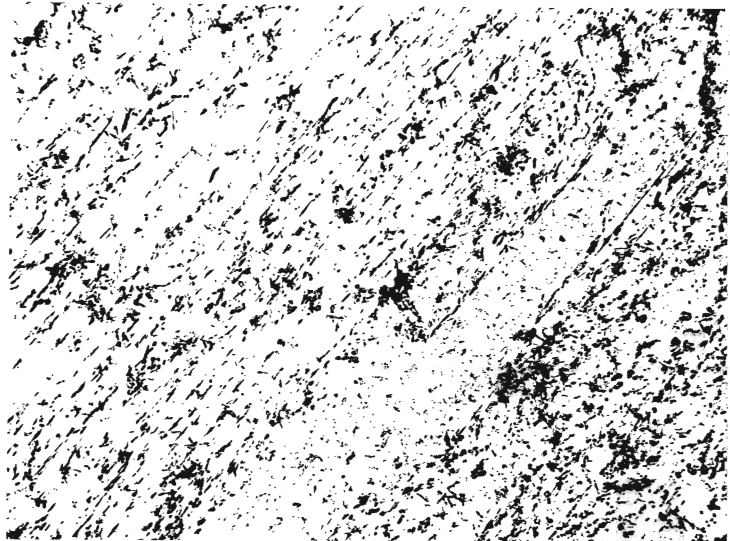
5. 930/765°C (Figure 41)

Compared with the structure described in 4, the carbide dispersion within the ferrite was, in this case, much more uniform and stringer as well as globular carbides were present. The latter carbide type was, however, predominant.

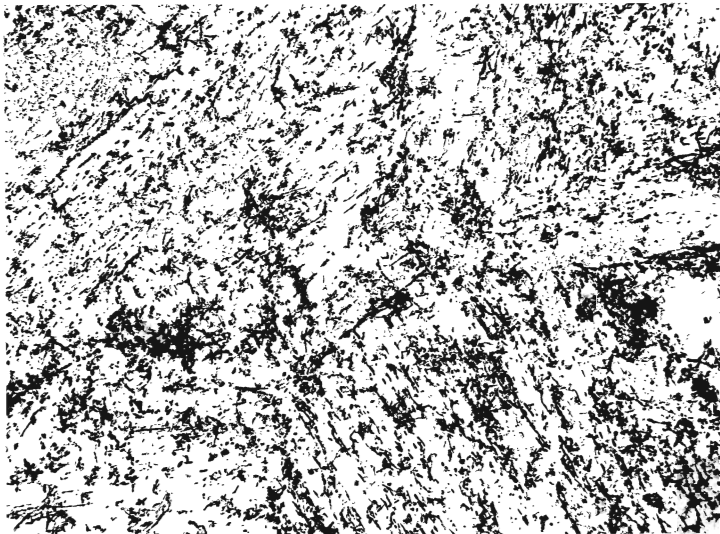
6. 765°C, 765/650°C (Figure 42)

This structure consisted of a fairly uniform carbide dispersion in a ferrite matrix, (a), but the ratio of stringer to globular carbide, (b), was greater than in 5. Fine rod shaped carbides were also present in some regions, (c).

(a)
X 3,000



(b)
X 3,000



(c)
X 15,000

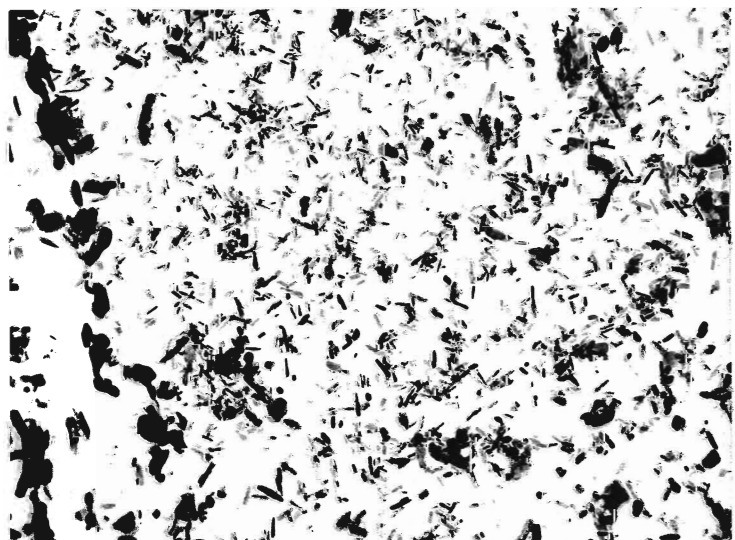


Figure 37. Post cycle heat treated microstructure typical of the thermal cycled 1275°C, 1275/1275°C and 1275/6500°C peak temperature conditions. Carbon extraction replicas.

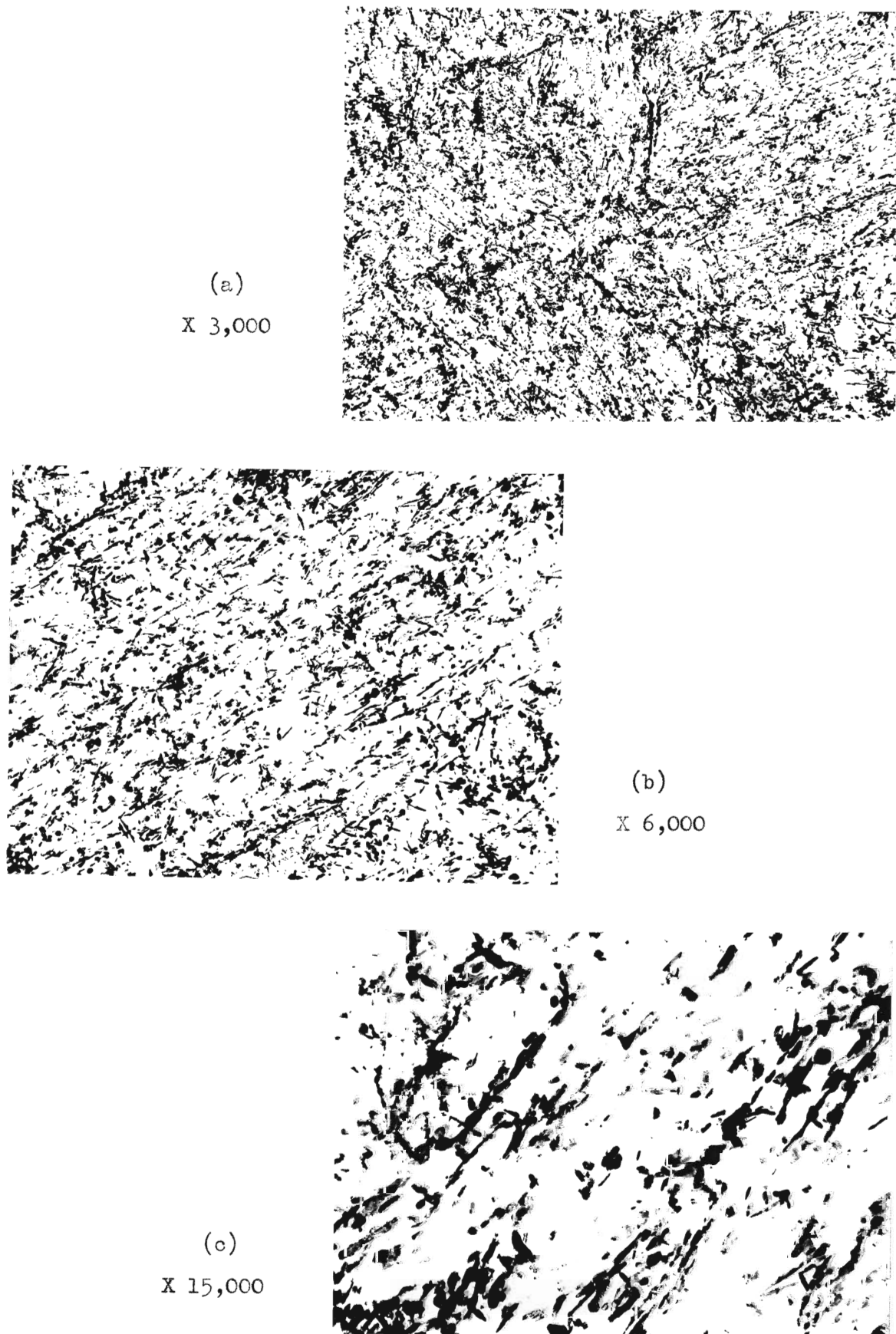


Figure 38. Post cycle heat treated microstructure of the double thermal cycled 1275/930°C peak temperature condition. Carbon extraction replicas.

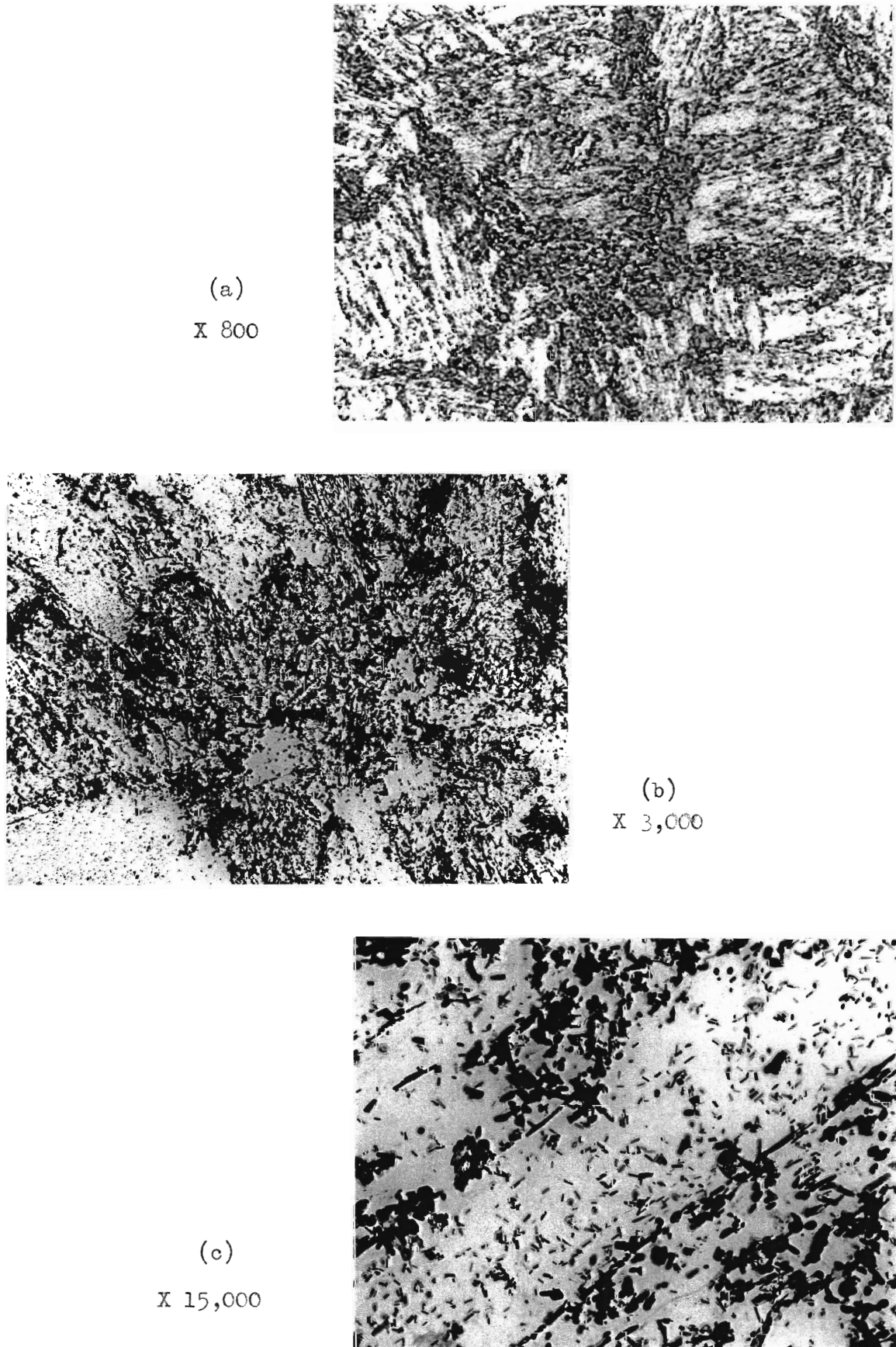


Figure 39. Post cycle heat treated microstructure of the double thermal cycled 1275/765°C peak temperature condition. Optical micrograph (a); carbon extraction replicas (b) and (c).

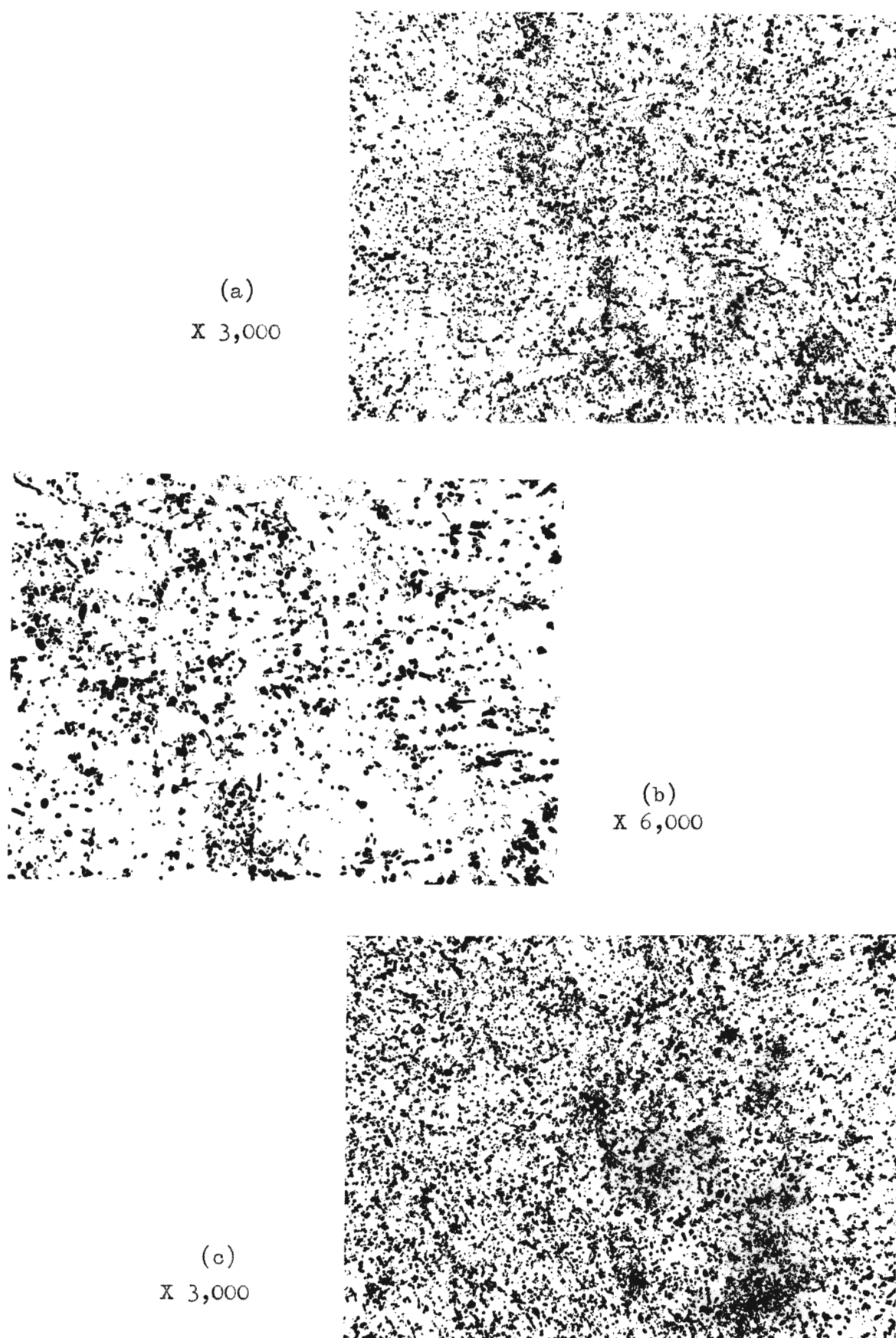
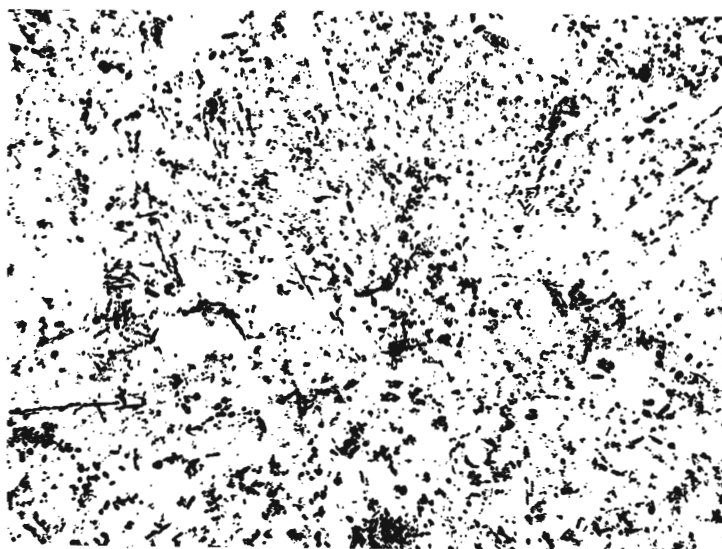
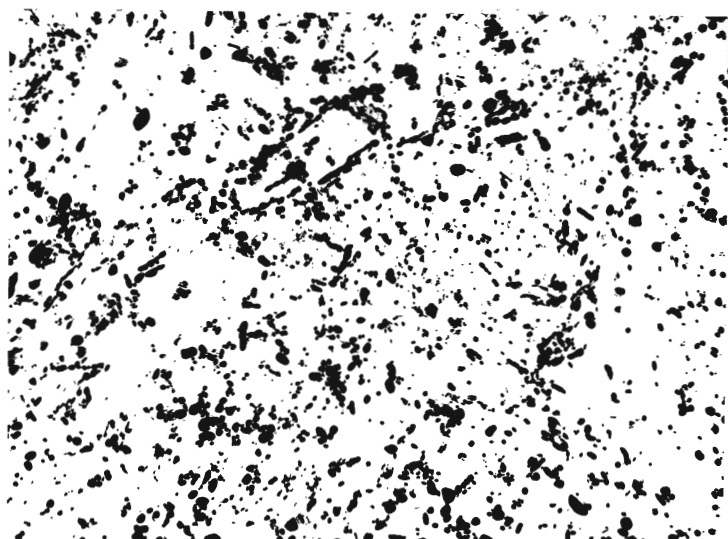


Figure 40. Post cycle heat treated microstructure typical of the thermal cycled 930°C and 930/650°C peak temperature conditions. Carbon extraction replicas.

(a)
X 3,000



(b)
X 6,000



(c)
X 15,000



Figure 41. Post cycle heat treated microstructure of the double thermal cycled 930/765°C peak temperature condition. Carbon extraction replicas.

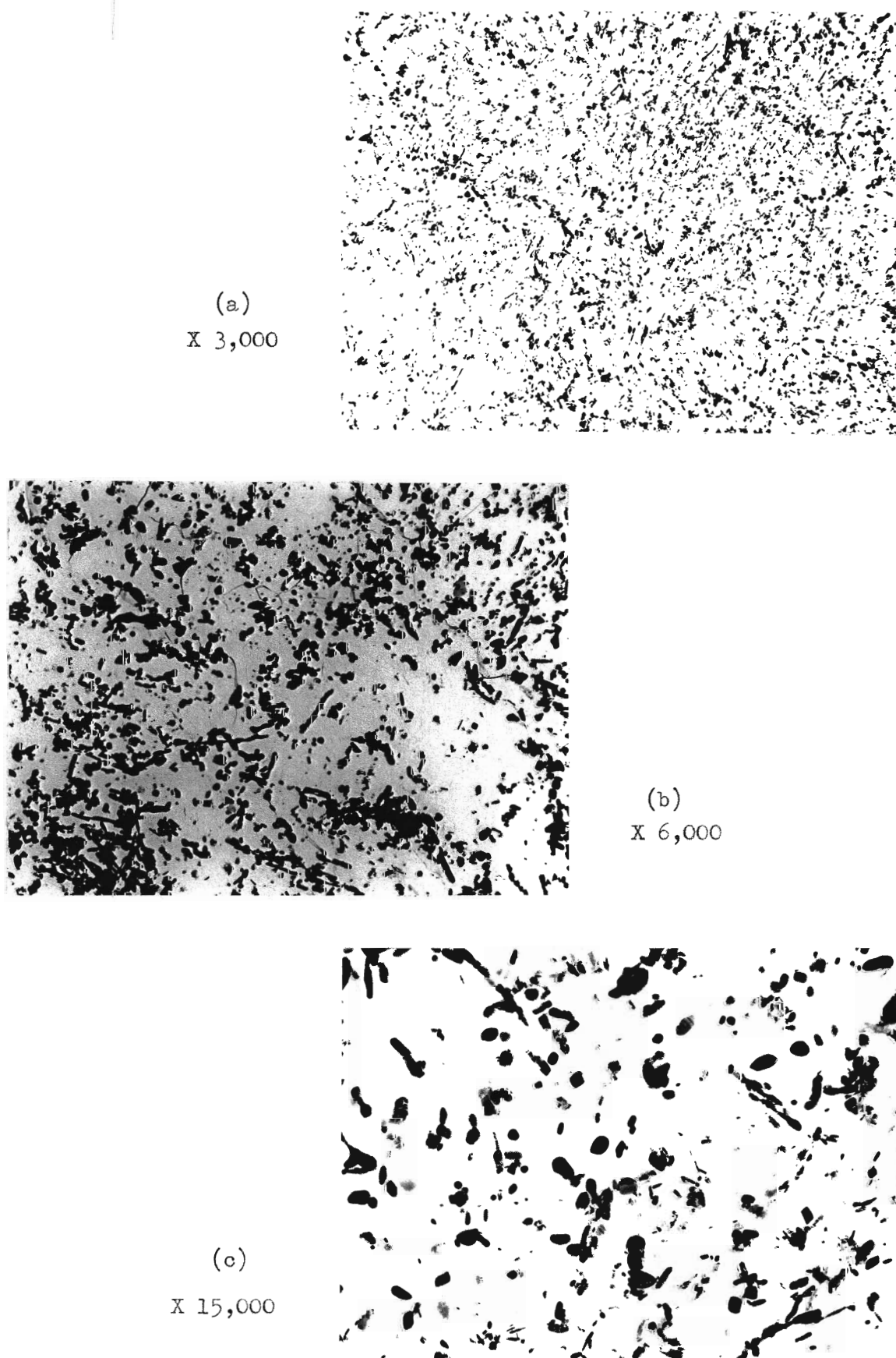


Figure 42. Post cycle heat treated microstructure typical of the thermal cycled 765°C and 765/650°C peak temperature conditions. Carbon extraction replicas.

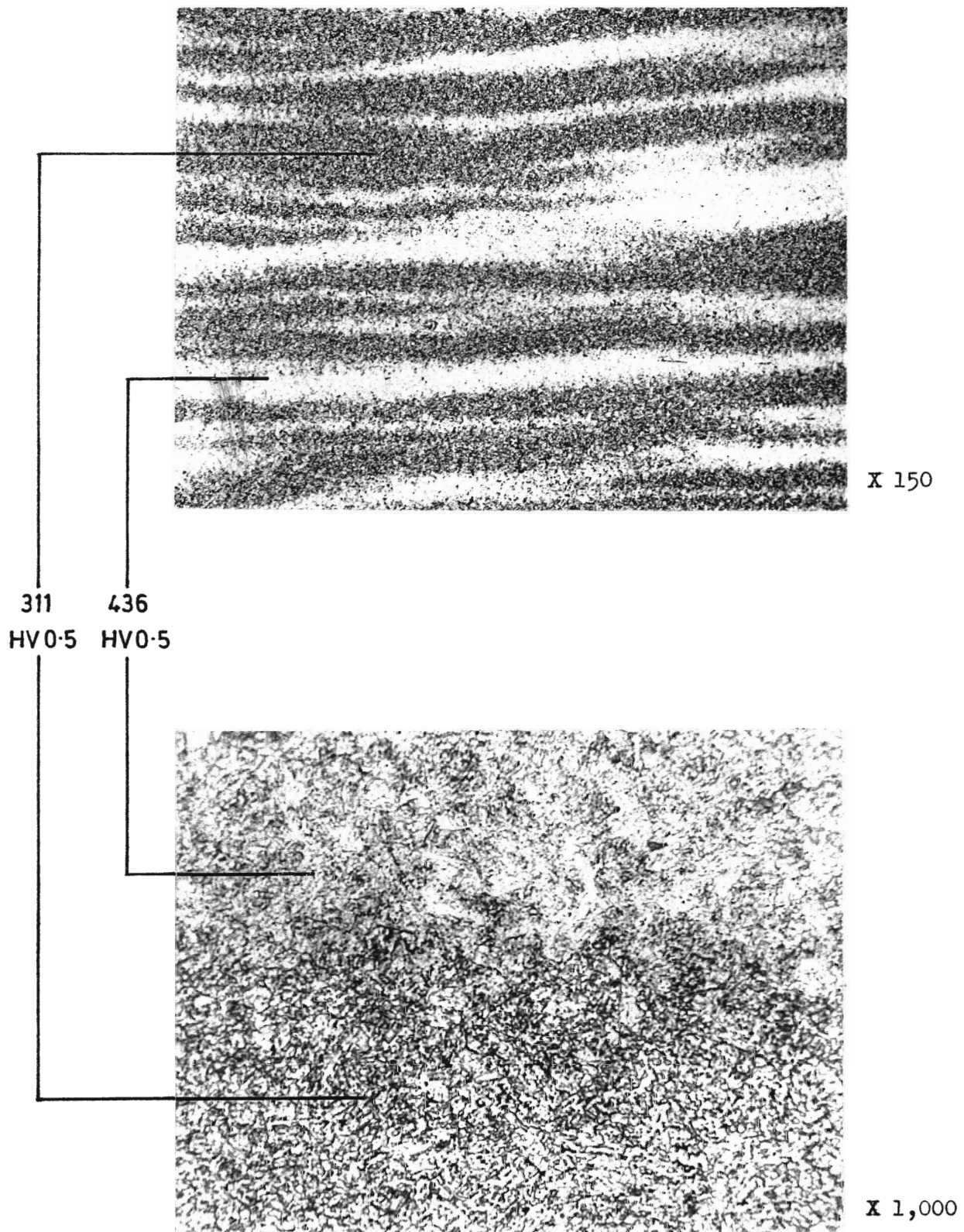


Figure 43. The nature of banding in the parent plate after thermal cycling to a peak temperature of 930°C. Optical micrographs.

4.6. PARENT PLATE BANDING

The existence of banding in the parent plate was not very evident in the as-received condition or after the 1275°C peak temperature cycle. In specimens cycled to peak temperatures of 765 and 930°C banding was very noticeable using low power optical microscopy. The banding in a 930°C peak temperature cycled specimen as revealed by etching in 2% nital is shown in figure 43. Hardness testing was carried out in a random manner using a 5kg. load. Inter-band hardness surveys were made using a 0.5kg. load, different bands being selected for determination within each group. A measurement traverse was made across the specimen width normal to the layer direction to estimate the relative widths of the light and dark etching layers. The results obtained were:

Band	Band width mm.		Hardness, HV 0.5		Hardness, HV 5	
	Mean	Std. Dev	Mean	Std. Dev	Mean	Std. Dev
Light Etching	0.056	0.031	436	7	353	17.4
Dark Etching	0.195	0.044	311	11		

Electron probe microanalysis across a typical region normal to the band length yielded the following results:

Element	Weight % in Band		
	Dark Etching	Light Etching	Dark Etching
Manganese	0.41	0.59	0.39
Nickel	2.2	3.5	2.31
Chromium	1.7	2.4	1.2
Molybdenum	0.2	0.26	0.18

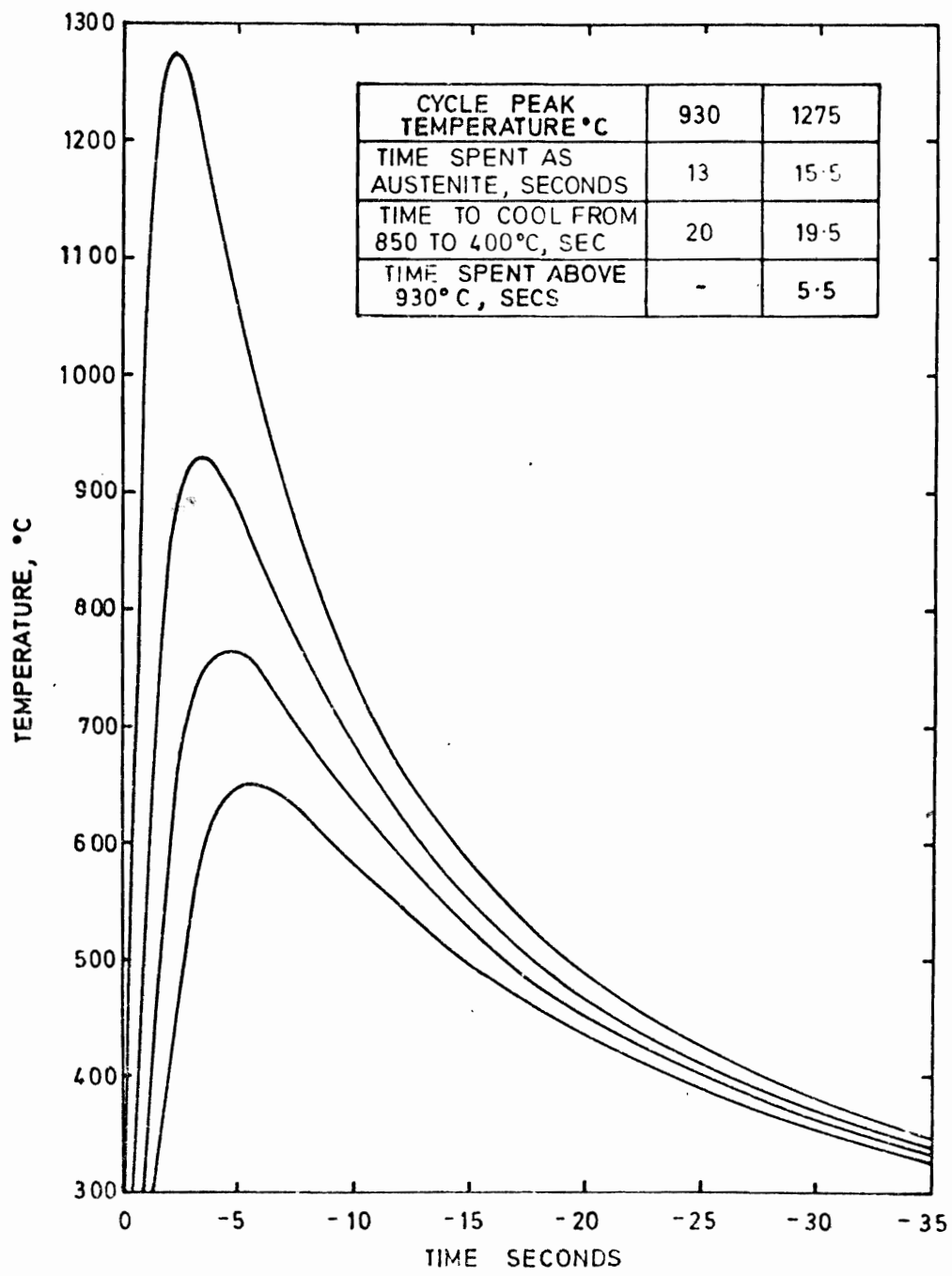


FIGURE 44 THERMAL CYCLE OBTAINED USING PROGRAMME VARTHERM 2

5. DISCUSSION OF RESULTS

5.1. COMPUTED THERMAL CYCLES

In program VARTHERM 1, page 188, the electrical input was multiplied by a factor η , the arc efficiency, to obtain the value of heat flow to be used in this and other programs of the series. There is no single clear definition of efficiency which finds general acceptance in the literature but it may simply be referred to as the ratio of the heat transferred to the weldment to the thermal equivalent of the electric power of the arc. Arc efficiencies are a function of many variables and to obtain reasonably accurate estimates there is no substitute for measurement. The efficiency of 0.98 used in this study was taken from a range of 0.91 to 0.99 given by CHRISTENSEN et al., (27) who measured the amount of heat leaving the bottom and sides of a plate during the submerged arc welding of mild steel. The quoted range does not include the heat loss from the top surface of the plate or that due to the part of the flux which is heated but not fused. However, in view of the large uncertainties inherent in measurement and the many variables involved, this probably causes no significant difference. The efficiency value of 0.98 employed for the computer study was considered reasonable since the results obtained were comparable with thermal cycles measured by SMITH (243) under controlled laboratory conditions.

The nature of the release of latent heat from the solidifying weld pool has received little attention in the literature. As the welding speed is increased molten material is forced behind the arc and the molten pool lengthens. This tends to give the same effect as another heat source of varying size located at a varying distance behind the arc. Due to the presence of convection currents within the molten weld pool the major proportion of latent heat release appears to occur at the tail of the pool. In the model used in this computer study the latent heat flow was assumed to originate from a second point source which followed the primary source, being separated from it by a distance equivalent to the weld pool length. The thermal cycle for a peak temperature of 1275°C, figure 15, page 86, measured by Smith, shows a marked inflection on cooling at about 1100°C. Such a curve was not produced by VARTHERM 2, figure 44, page 126, and by comparing this curve with the measured one and employing program VARTHERM 3, page 191, a sigmoidal function was suggested to account for latent heat flow. Actual comparison of computed and measured thermal cycles for the submerged arc welding of 1.5 in. thick plate is necessary for a wide range of welding conditions before a universal solution might be obtained to account for latent heat release. However, for the welding conditions used in this research the model employed gave very good correlation with practice and is worthy of acceptance until further

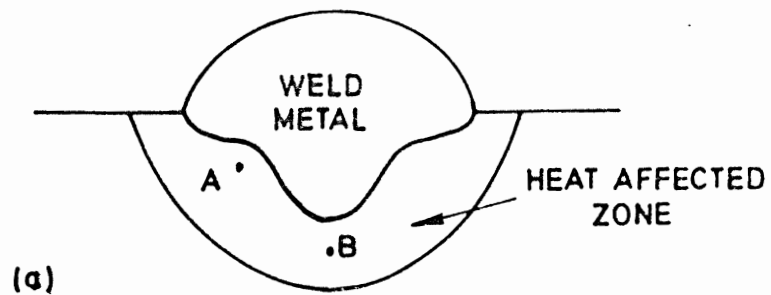
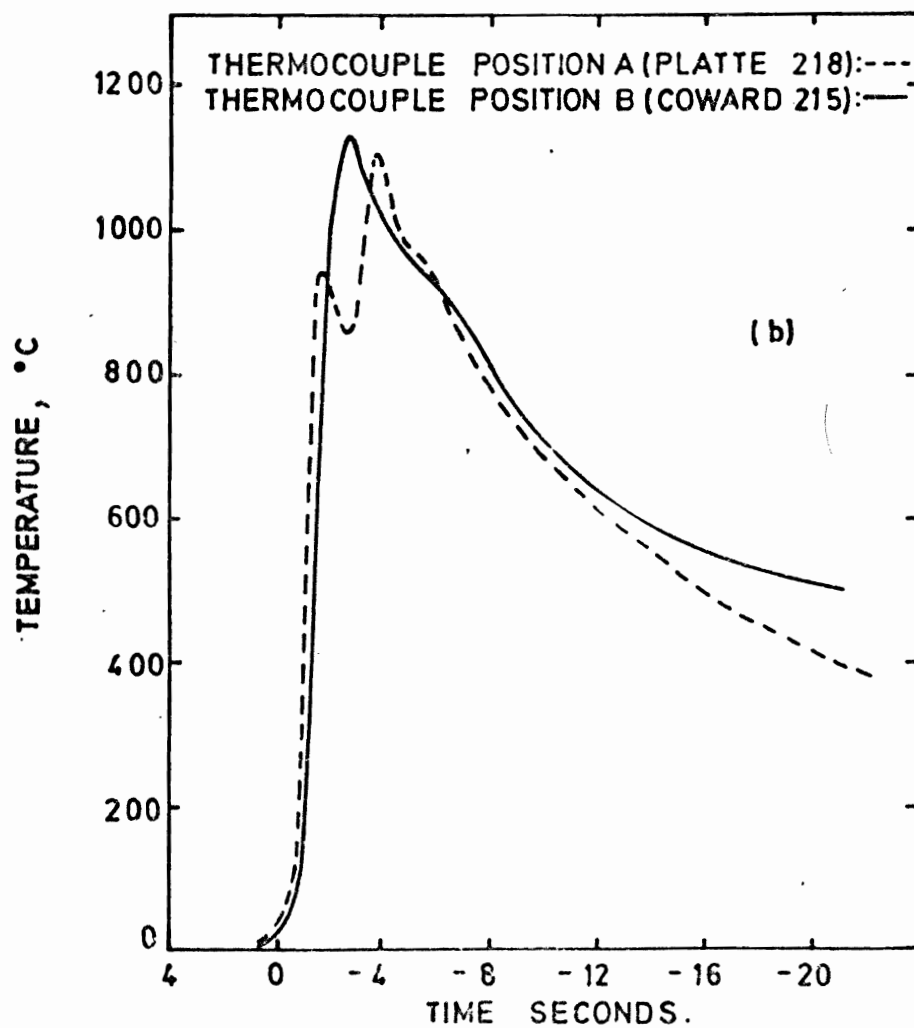


FIGURE 45 EFFECT OF THERMOCOUPLE POSITION ON THERMAL CYCLE SHAPE. POSITIONS OF THERMOCOUPLES IN THE HEAT AFFECTED ZONE (a) AND THERMAL CYCLES CORRESPONDS TO THESE POSITIONS (b)

investigation either substantiates or negates its use. It is rather imprudent to accept that all high temperature inflections in measured thermal cycles are associated with the release of latent heat from the molten weld pool as the actual shape of the cycle is extremely sensitive to thermocouple position. Figure 45(a), page 128, shows the cross-section of a weld with a fairly pronounced penetration 'nose' with thermocouples positioned at two points, A and B, within the heat affected zone. Assuming that heat flow occurs perpendicular to the fusion boundary it is possible for thermocouple A to experience heat flow from two major directions and possibly at different times due to its position relative to the fusion boundary. PLATTE (218) has shown that the thermal cycle produced by a thermocouple in such a position, figure 45(b), page 128, exhibits a double peak during the heating portion of the cycle. Increase in welding speed could give the 'double-hump' effect at a later stage in the cycle. It is interesting to note in Platte's curve the existence of a slight inflection during the cooling portion of the cycle which could be, in this case, attributed to latent heat release. Complex situations such as these require care in interpretation. Thermocouple B is directly below the point of maximum penetration and experiences major heat flow in one direction only and no discontinuities are apparent in the heating portion of the thermal cycle measured by COWARD (215), also shown in figure 45(b). An inflection in the high temperature range in the cooling portion of this curve can be attributed to latent heat flow from the solidifying weld pool due both to the thermocouple position and to the fact that the degree of penetration was relatively small in the weld studied. A similar but not nearly so marked inflection, as already noted, is present in Platte's curve and DOLBY (191) has reported inflections similar to those found by Coward and Smith. There is no evidence to the contrary at present to suggest that the high temperature inflection in the cooling portion of the weld thermal cycle for high peak temperatures is not due to the release of latent heat by the solidifying weld pool. It is clear that a high degree of discrimination must be applied in attaching significance to inflections appearing in measured weld thermal cycles. The latent heat inflection discussed has not appeared, for some reason or another, in many curves in the literature. GROTKE (210) used cycles measured by Platte for a heat input of 31 kJ. in^{-1} as a basis for a series of simulation studies on a range of high strength steels. He carried out some further studies on HY-150 steel using, quote, "thermal cycles based upon the heat flow relationships for welds in 1 in. thick plate deposited at 50 kJ. in^{-1} with 150°C preheat." The curves for this condition shown by Grotke included a latent heat inflection but how they were established using standard heat flow relationships is a point for conjecture. However, the awareness of such a latent heat effect and the use of cycles which exhibit it in simulation studies is encouraging.

The thermal conductivity function does not, as it stands, appear to have fundamental significance and is based on a comparison of measured and computed (VARTHERM 4(2)) cycles. That it did, however, give good correlation with the practical situation, figures 51 and 52, pages 177 and 178, is sufficient evidence for its application. As can be seen from figures 50 and 54, pages 174 and 182, the thermal conductivities of steels do vary with temperature and during transformation of undercooled austenite to martensitic or martensitic-bainitic structures during continuous cooling it appears reasonable to assume a further incremental change due to fundamental and morphological alteration in structure. The thermal conductivity of a martensitic or martensitic-bainitic structure, due to the inherently high degree of distortion present on the atomic level of dimensions and the complex structural morphology, should be lower than that of a tempered structure containing a fine fairly uniform carbide dispersion in a ferrite matrix. The intercept on the thermal conductivity axis, k_0 , of the thermal conductivity-temperature plot, figure 55, page 184, is a result of the function used. No data was available from which this intercept could be reliably obtained and very little attention has been given in the literature to the measurement at room temperature or 0°C of the thermal conductivities of a steel in a variety of structural conditions produced by different standard heat treatments let alone those due to thermal excursions of the type found in the weld heat affected zone. It was assumed in the formulation of the polynomial latent heat function that the k_0 value would lie between the extrapolated value for austenite and that for the fully tempered ferrite - carbide structure; the lower the value of k_0 the higher the temperature for any given time during the cycle below the B_s temperature. If room temperature thermal conductivities could be measured for the structural conditions produced by thermal cycling this would remove some doubt from the format and use of the thermal conductivity function. Thermal conductivity measurements during austenite transformation on continuous cooling would be the ultimate solution to the problem but a high degree of difficulty is envisaged in making such determinations. In program VARTHERM 5, page 195, the accuracy of temperature prediction is increased as time increments in the 'time loop' are decreased since the preceding temperature forms the basis of establishing new values of k , k_0 , m and λ for each time increment on cooling below the B_s temperature.

Since the cooling rate to the transformation start temperature appears to be more important than the cooling rate through the lower part of the transformation range more significance should be placed on the latent heat release function introduced than on that for thermal conductivity correction. Nevertheless variation of thermal properties over the whole cycle temperature range is necessary to produce the correct form of cycle and much confusion has arisen in previous work owing to the necessity to choose a constant value, for say, thermal conductivity in the

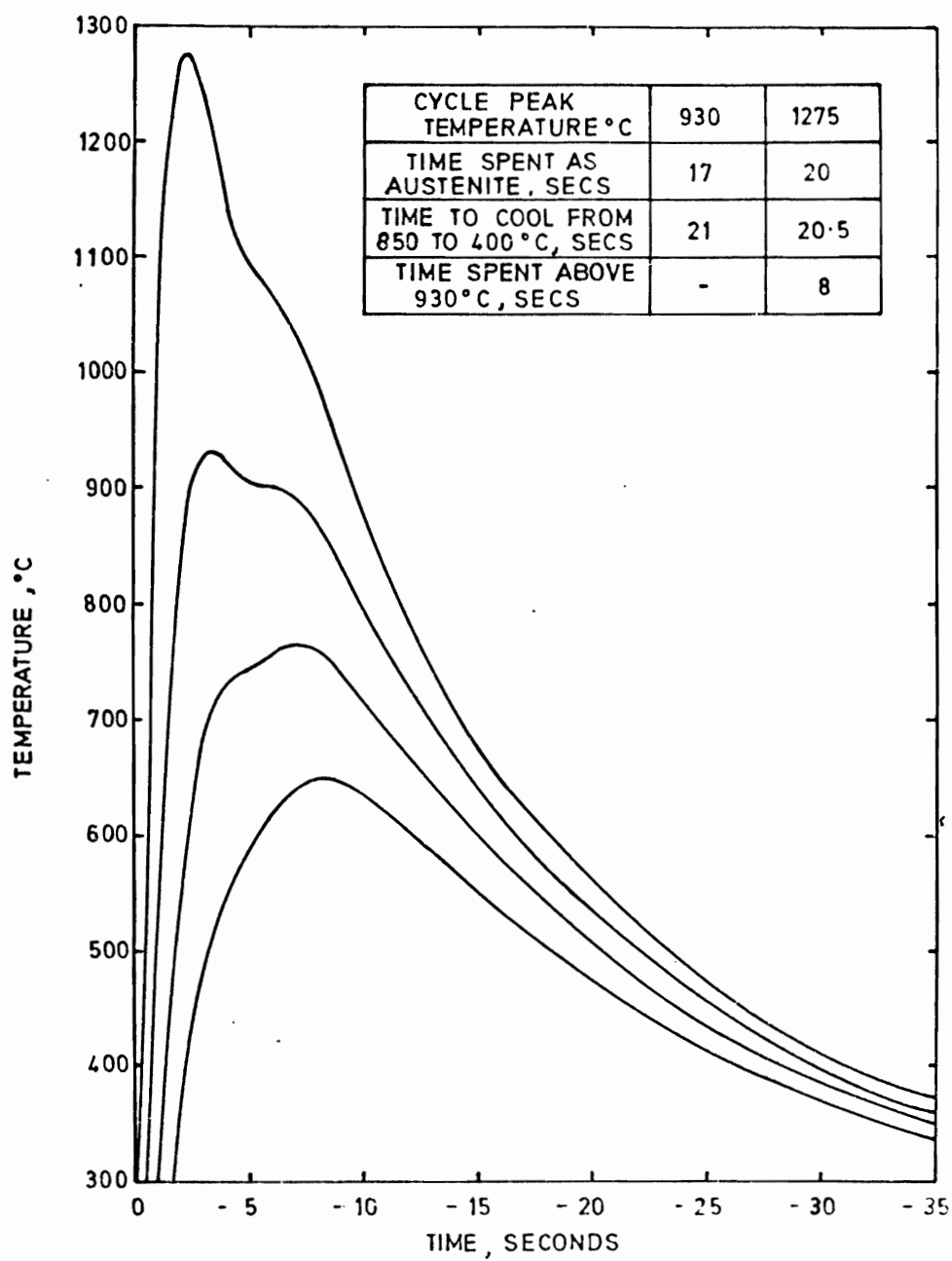


FIGURE 46 THERMAL CYCLES OBTAINED USING PROGRAMME VARTHERM 5.

theoretical prediction of weld thermal cycles. The importance of using calculated thermal cycles which include a latent heat correction may be ascertained from a study of figures 44 and 46, pages 126 and 131. The thermal cycles obtained by means of program VARTHERM 5, figure 46 indicate that for any peak temperature above A_{c3} , the time spent during cycling within the austenite field is greater than when the latent heat function is not considered as in the cycles in figure 44 produced by program VARTHERM 2. Lack of time during this research precluded a computer study, using these programs as a basis, to investigate the degree of carbide solution and austenite grain growth during thermal cycling; this work is, nevertheless, at present being carried out. If the 1275°C peak temperature thermal cycle is considered, then, on the basis of standard solid state diffusion and grain growth criteria, it may be inferred that the greater time spent, particularly in the higher temperature region, in the austenite field for the VARTHERM 5 cycle would produce a more homogeneous austenite but of somewhat larger grain size than would be produced by the VARTHERM 2 cycle. For the lower 930°C peak temperature cycle the time spent as austenite is again greater for the VARTHERM 5 cycle than for the VARTHERM 2 cycle but, as might be expected, is not as great in either case as that for the 1275°C peak temperature cycle. The VARTHERM 5 cycle should produce a greater degree of carbide solubility and a somewhat more homogeneous austenite than would be produced using the VARTHERM 2 cycle during simulation. The time to cool from 800 to 400°C is very similar for the cycles predicted by VARTHERMS 2 and 5 and both correlate well with measured cycles; for a given peak temperature cycle the difference is only of the order of one second, being greater for VARTHERM 5. In terms of the continuous cooling transformation (CCT) diagram due to INAGAKI et al., (220), figure 47, page 133, for any HY-80 type steel the structures produced for a given peak temperature are only insignificantly different if the cooling time from 800 to 400°C differs by only one to two seconds. The importance of cooling rate has been stressed by several workers but the form of the thermal cycle in the austenite field and particularly in the region of the peak temperature, in promoting austenite homogeneity and carbide solution, is usually neglected. D'ANDREA and ADAMS (114) have discussed cycle shape in terms of austenite homogeneity but do not refer to the prior austenite grain size produced by a given cycle although this is important in many structures in determining mechanical properties. INAGAKI and SEKIGUCHI (113) have shown, figure 5, page 32, that as the cycle peak temperature is increased the start of transformation lines in a CCT diagram, for a given cooling rate, are moved to longer times and lower temperatures. This suggests that the more homogeneous the austenite and the greater the degree of carbide solution in the austenite prior to transformation then the slower will be the critical cooling velocity necessary to produce an entirely martensitic structure and for a cooling rate slower than this a decrease in

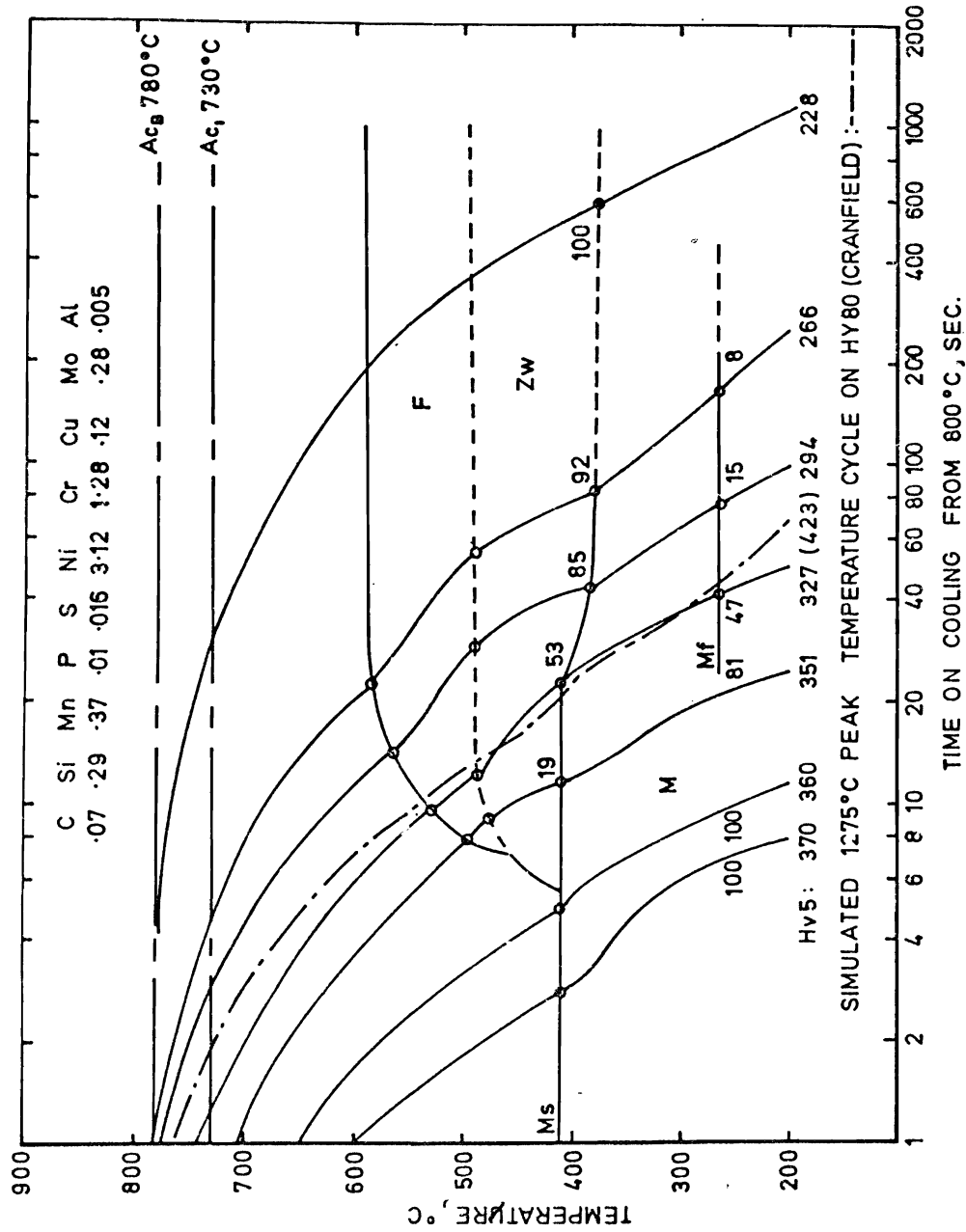


FIGURE 47 SH-CCT DIAGRAM FOR AN HY80 TYPE HT70 STEEL. MAXIMUM PEAK TEMPERATURE : 1350°C. (AFTER INAGAKI ET.AL.)

the degree of transformation occurring within the upper bainitic range.

Thermal cycles which include the latent heat correction would produce such an effect when compared with cycles which have no such modification. For a 1275°C peak temperature thermal cycle VARTHERM 5 would tend to produce in simulation studies, due to increased austenite homogeneity, a significantly greater proportion of auto-tempered martensite than would be produced using VARTHERM 2. Many simulation studies have been made using a slightly modified form of the Rosenthal solution (251) as a basis in which the cooling rate was taken as the ultimate criterion of thermal cycle efficiency. Simulation studies on low carbon, low alloy steels by NIPPES and SAVAGE (198), NIPPES et al., (205) and SAVAGE and OWCZARSKI (110) have used this approach and found the presence of regions of high carbon martensite in the simulated structures and this is of course a function of austenite heterogeneity with respect to carbon content prior to transformation. It is suggested herein that had the effect of latent heat release been considered in the modified equations employed to obtain the thermal cycles then the likelihood of occurrence of such regions, which have a detrimental effect on notch toughness properties, would have been very much less. Thus if a theoretical approach is made to obtain thermal cycles for simulation studies the latent heat effect must be included to ensure peak and overall cycle broadening to achieve good correlation with the practical situation. If theory predicts the correct cooling rate anomalous results can still be obtained if the time spent as austenite is relatively short.

As may be seen from figure 53, page 180, the predicted heat affected zone width using program VARTHERM 4 is approximately 0.35 mm. wider than when program VARTHERM 1 is employed. This increase is due to the effect of the latent heat function in increasing the peak temperature for a given distance from the point source for peak temperatures below about 850°C. The results show reasonable correlation with practice.

5.2. SIMULATED THERMAL CYCLES

The correlation between the measured and computed thermal cycles with those reproduced by the simulation equipment in specimen blanks was generally very good. The measured 1275°C cycle was rather difficult to reproduce in the region of the peak temperature, figure 48, page 135, due to the sharpness of the peak. The electrical characteristics of the simulation equipment caused a somewhat broader peak which could account for a slight increase in prior austenite grain size in the simulated structure compared with that in the region of the actual weld

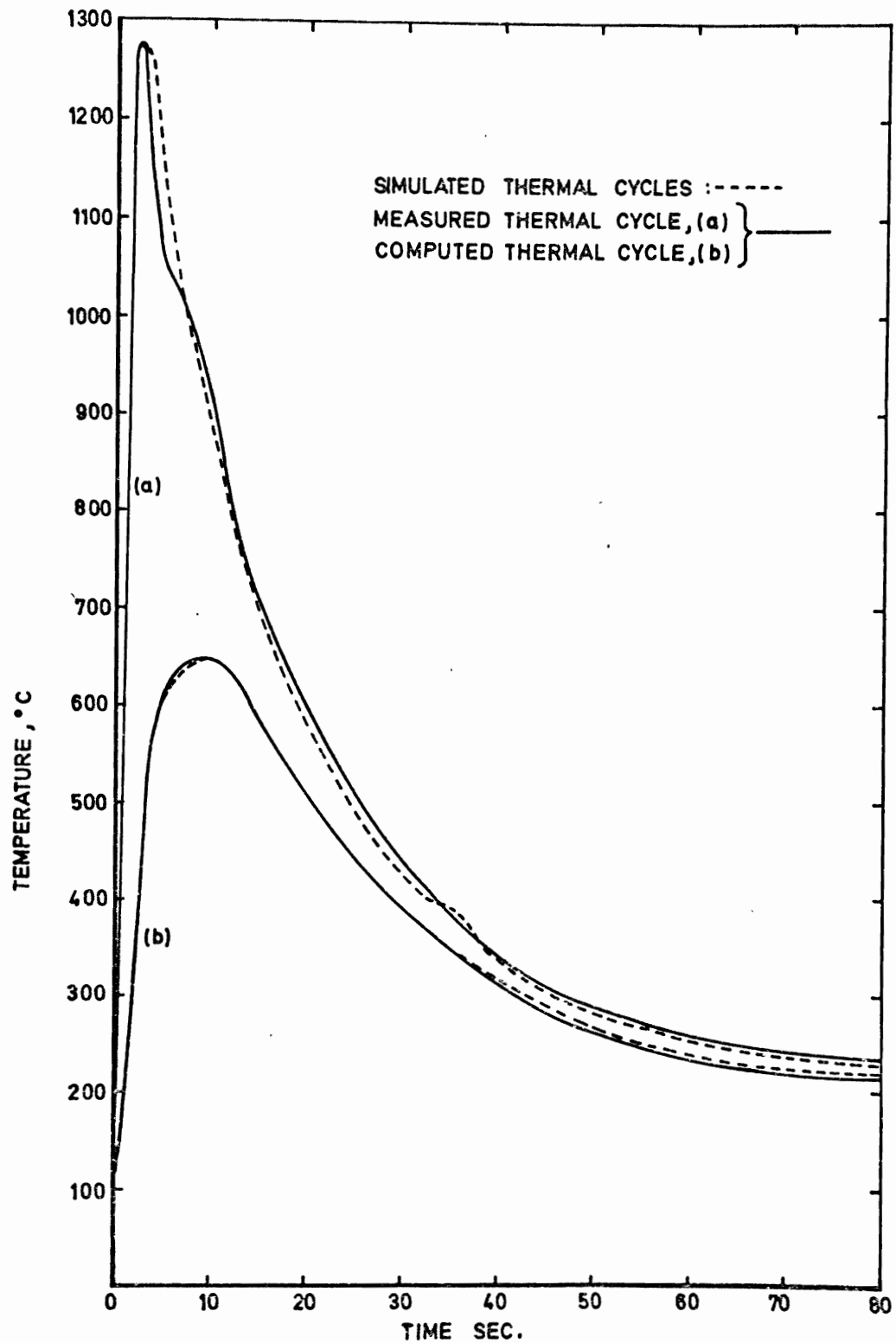


FIGURE 4.8 COMPARISON OF SIMULATED THERMAL CYCLES WITH THOSE PRODUCED BY MEASUREMENT AND COMPUTATIONAL METHOD.

heat affected zone subjected to a similar cycle. A more homogeneous austenite in the simulated structure may be inferred from the discussion in section 5.1. As is also shown in figure 48, very good reproducibility was obtained for the lower 650°C peak temperature cycle. This was also true of the 765 and 930°C peak temperature cycles.

As may be observed in figures 19 and 48, pages 93 and 135, inflections are apparent in the 1275°C peak temperature cycle in the region of 400°C but are not evident in the cycle measured by Smith as shown in figures 15 and 48, pages 86 and 135. This could be attributed to the much larger volume of material undergoing transformation simultaneously in the simulated specimens to that in the welding situation where the lack of transformational uniformity in the heat affected zone leads to averaged temperatures appearing in the cooling curve. This thermal arrest effect in simulated specimens has also been noted by Inagaki et al., (220) whose conclusions regarding its nature are similar to those suggested herein. The arrest is attributed to the exothermal transformation of austenite to martensite since specimens subjected to fast cooling rates under uncontrolled conditions in the water cooled simulator blocks showed similar behaviour.

The 930°C peak temperature cycle showed inflections in the cooling portion of the cycle at about 550°C and 400°C. It is tentatively suggested that, due to banding, the carbide dispersion present and non uniform alloying element distribution, one type of band started transformation to bainitic structural forms at about 550°C whereas some remaining austenite in these bands and the wholly austenitic structure in the other band type began transformation to martensite at 400°C which subsequently became auto-tempered on cooling to room temperature. The effect of microsegregation on the transformation characteristics of a low alloy steel was discussed in section 2.2.5.1., and illustrated by means of TTT diagrams in Figures 4a and 4b, page 30 .

In material which had been initially cycled to 1275°C then recycled to 930°C, the thermal arrest at about 550°C appeared to be almost non-existent but that at about 400°C was much more marked than the one produced by subjecting the parent material to a single 930°C peak temperature thermal cycle. A possible explanation for such behaviour is that the initial high temperature cycle caused complete temper carbide solution in the austenite which ultimately transformed on cooling to predominantly auto-tempered martensite. On subsequent cycling of this structure to a peak temperature of 930°C the small auto-tempered carbides were easily taken into solution in the austenite which on cooling transformed to a virtually complete auto-tempered martensitic structure. Banding, though still present to a degree was not nearly so marked after this double cycling treatment. These suggestions were substantiated by the results of

metallographic examination and hardness testing.

When a 765°C peak temperature thermal cycle was imposed upon material initially cycled to 1275°C the thermal arrest at about 400°C was evident. Single cycling of the parent material to a 765°C peak temperature gave only very slight evidence of such an arrest. This behaviour together with metallographic observations indicates only partial solution of the temper carbides in the parent plate on single cycling to peak temperatures of 765 and 930°C. Virtually complete carbide solution with the production of a relatively homogeneous austenite, prior to transformation, occurs when such cycles are applied to material which has had an initial 1275°C peak temperature cycle.

An important feature, not always appreciated in simulation studies on low carbon, low alloy steels, is the fact that the simulation cycle only lasts for a period of about 3 minutes before cooling to room temperature occurs. In practice however, the heat affected zone may exist in the region of 200°C for a much longer period of time which would allow some further auto-tempering to occur of any martensite, produced by cycling.

The discrepancy apparent in the region of the peak temperature in the 1275°C peak thermal cycle is attributed to the inherent characteristics of the simulation equipment. The output from the computer for a given set of welding conditions, including the latent heat function and thermal conductivity correction, could be obtained in a digital form on tape. If a tape-reader was incorporated in the simulator control apparatus and providing the response of the equipment was equal to the task then a much closer approximation to the required thermal cycle could be achieved than is at present obtained by manual resistor adjustment. The only current method available for the production of a 'sharp' peak in high temperature cycles is manual cut-out of the rotary potentiometer drive for a predetermined time period and the accuracy of reproducibility of this procedure is a function of the operator. This situation does not seem to be entirely satisfactory. In addition it is felt that due to the sensitivity of the equipment to simulator blank dimensions, surface finish etc., digital tape input would give better reproducibility as regards peak temperature and reduce the incidence of failures due to either overshoot or not attaining the required temperature range.

5.3. MECHANICAL TESTING

The errors involved in measuring mechanical properties using standard techniques are well documented. However, in testing to assess the properties of simulated heat affected zone structures caution must be exercised in the interpretation of the results

obtained. The standard errors involved in the Charpy V-notch impact test include a small amount of energy lost as friction of the moving machine parts, air resistance, vibration of the testing machine and setting the fractured specimen in motion. Due to the comparative nature of the test such errors are not usually taken into account, although they are recognised, since they are typical of the vast majority of testing machines. GROTKÉ (210) in his simulation study of T-1 and other low alloy steels thermally cycled to a peak temperature of approximately 1310°C to produce a martensitic region in the simulator blanks found that, on the subsequent Charpy V-notch impact testing of standard specimens machined from the blanks, the region of the test specimen in contact with the machine anvil showed evidence of deformation. The impact results obtained for the simulated specimens were corrected for this feature by testing a series of fully hardened specimens and cross-correlating the energy absorbed and lateral expansion measurements of the two specimen types. A discrepancy of approximately 25 ft lb. in the upper shelf energy was revealed by this procedure. Such discrepancies, which are a function of the structure of the specimen in contact with the anvil, are much more important than the so called 'standard' errors noted above. Such deformation was observed in this present research in the region of the specimens in contact with the anvil and the results obtained should be interpreted with this feature in mind. However, the adoption of Grotke's procedure was not feasible since the range of structures produced by different peak temperature cycles and combinations of these cycles could not be produced along the specimen length by standard heat treatment procedures. However, if the surface region of the Charpy specimens in contact with the machine anvil could be either flame or induction hardened, without detriment to the structure in the region which would undergo thermal cycling, then deformation during testing in this region would be virtually non-existent and more conservative assessments of notch toughness determined. Such a method was not, however, used in this present study due to the large number of specimens involved. Also the results obtained were intended for inter-cycle comparison and for correlation with the results obtained, without such correction, for other low alloy steels. It is suggested that such a procedure could be adopted with success in future studies.

The method of assessing percentage crystallinity of the Charpy fracture surface is open to investigator error and the results obtained can be expected to be accurate to $\pm 10\%$. This error could probably be improved upon by the use of a transparent ruled net in conjunction with low power examination of the fracture surface on the screen of a projection microscope. This, however, would be a time consuming operation and in a wide research programme involving a large number of specimens could probably not be justified. In such an instance, like this

present research, the $\pm 10\%$ error in measurement is acceptable.

Hardness traverse testing has long been employed as a method of obtaining an assessment of the variation of mechanical properties across the heat affected zone. However, as shown in figure 18, page 91, hardness as measured using a 5kg. load showed a fairly uniform curve but when a 0.5kg. load was employed marked peaking in the curve was evident. The peaking effect is attributed to the differential transformation characteristics of adjacent bands in the plate; higher load testing tends to average out the band-to-band hardness difference. This fact does not appear to be fully appreciated in the literature although a fairly recent catastrophic service failure (150) has been attributed to the presence of high hardness bands in the heat affected zone of a welded plate. The degree of scatter in hardness values is indicated in Tables 5 and 6, pages 169 and 170, by the standard deviation, for tests on simulated specimens using a 5kg. load. The highest degree of scatter was evident for cycling of the parent plate to a 930°C peak temperature when transformation of adjacent bands to martensitic and bainitic structures occurred. In the case of the duplex 1275/765°C cycle condition the degree of scatter is again relatively high and this is attributed, in the main, to the duplex ferrite - auto-tempered martensite structure present. As might be expected, post cycle heat treated and parent material structures which showed a fairly uniform carbide dispersion in a ferrite matrix had an inherently low degree of scatter in the hardness results. Where the carbide was in the form of a fairly concentrated and continuous network at the prior austenite grain boundaries e.g., the 1275/765°C post cycle heat treated condition, the scatter was somewhat higher.

The degree of scatter in the results for 0.2% proof stress, and ultimate tensile strength is low ($< 0.4 \text{ tonf. in}^{-2}$.) and the values quoted in Tables 5 and 6, pages 169 and 170, are reliable. The results quoted in past research (246) for % elongation are open to doubt since the elongation was taken as the overall measurement of the Instron testing machine crosshead during tensile testing and could not be directly related to the 0.3 in. gauge length of the specimens tested. For this reason elongation measurements have not been considered in this present study and ductility has been assessed by means of reduction of area measurement. The error involved in this procedure is only of the order of $\pm 2\%$.

5.4. BEAD-ON-PLATE WELD HEAT AFFECTED ZONE AND PARENT PLATE STRUCTURE

As discussed in section 4.2., the limit of resolution of the

optical microscope did not allow complete structural determination of the weld heat affected zone. Structural indications were revealed, figure 17, page 90, but the fine detail present could not be resolved. A 0.5kg. hardness traverse across the width of the heat affected zone showed the presence of banding while a traverse using a 5kg. load yielded a relatively smooth curve. This behaviour has been discussed in the previous section. The 5kg. load traverse was, however, consistent with the optical panorama at 100X in estimating the extent of the various structural regions within the heat affected zone. It is interesting to note from figure 18, page 91, a band within the grain refined region of similar hardness to that found in the grain coarsened region. Thus structures susceptible to heat affected zone cold cracking can be found within the grain refined region.

Due to the limitations of optical microscopy the structures produced by simulation were correlated with the heat affected zone structures using 5kg. load hardness testing.

The results of the investigation into parent plate banding in a specimen subjected to a 930°C peak temperature cycle indicated by means of hardness tests that the alloy enriched bands had a carbon content of approximately 0.2%; this value was estimated with the aid of figure 2, page 24. The 930°C peak temperature cycle condition was chosen since it gave the best optical contrast between the bands after etching, figure 43, page 124, and the highest inter-band hardness difference.

The impact properties of the as-received plate, Table 5, page 169, were consistent with the fine carbide dispersion in a ferrite matrix produced by tempering.

5.5. STRUCTURE AND PROPERTIES PRODUCED BY SINGLE CYCLE SIMULATION

5.5.1. 765°C peak temperature cycle

Partial austenitization occurred during cycling to this peak temperature which was within the intercritical range. Partial temper carbide solution was evident by virtue of a 'halo' effect around some carbides as shown in figure 35, page 114. The carbon extraction replica technique did not show any of the detail of the structure around such carbides but by the use of thin foil electron microscopy it is probable that this structure would be seen to be martensitic in nature. A high carbide density was evident in some regions of the structure whereas others were virtually devoid of carbide. Regions which had been austenitized during the thermal cycle appeared to have transformed to an

essentially mixed bainitic structure. The higher hardness compared with the parent material could be attributed to the presence of the fine carbide dispersion in the transformed regions and to the small 'envelopes' of martensitic transformation product assumed to be present around partially dissolved carbides. In addition, interstitial solid solution strengthening of the ferrite present is possible due to the cooling rate from the peak temperature allowing insufficient time for full precipitation. The fact that the impact properties were very similar to those of the parent plate, even though the hardness and 0.2% proof stress were somewhat higher, could be attributed to a refinement in ferrite grain size and an overall decrease in the size of the carbide particles present.

5.5.2. 930°C peak temperature cycle

According to the calculated A_{c3} temperature, section 3.3., this cycle should have caused complete austenitization. However, a 'halo' effect around some carbides, figure 32, page 111, once again suggested that the rapid thermal excursion caused only partial solution of the larger temper carbides in the initial structure with similar consequences to those noted in section 5.5.1. The carbides produced by this cycling treatment were mainly globular in character and there was some evidence of transformation to auto-tempered martensite within regions bounded by the ferrite-carbide transformation product. The interpretation of the complete structure was made rather difficult by the fact that alloy enriched bands, in the parent material, had transformed during cycling to a martensitic structure whilst adjacent alloy impoverished bands were mainly bainitic in character. The mechanism for the formation of such a duplex structure has been discussed in detail in section 2.2.5.1. This structural feature is in accordance with the results of optical microscopy, figure 43, page 124, and the transverse hardness survey and electron probe micro-analysis across the bands as detailed in section 4.6. In such a structure, through thickness notching tends to statistically average out the individual effects of both band types on fracture behaviour in the Charpy V-notch impact test. Within the degree of scatter encountered in the Charpy impact test (252) the upper shelf energy of 49 ft lb. virtually meets the overall 50 ft lb. requirement of MIL-S-16216G (228). In an actual weld heat affected zone if the 930°C peak temperature cycle was experienced by an alloy enriched band which transformed to a martensitic structure, then this region could reasonably be expected to have an upper shelf energy lower than that obtained for the 'composite' simulated structure. The converse would almost certainly be true for an alloy impoverished band which transformed to a mixed bainitic structure. The values for 0.2% proof stress, U.T.S., and hardness may be similarly considered. The poorer impact properties of this structure when compared with the parent plate may be attributed to the presence

of martensitic bands, a reduction in the proportion of free ferrite present and the presence of heterogeneous austenite prior to transformation leading to small envelopes of martensitic product around partially dissolved carbides in the bainitic bands.

5.5.3. 1275°C peak temperature cycle

This cycling treatment produced the poorest impact properties of the single cycled conditions studied, the upper shelf energy of 37 ft. lb. being well below the 50 ft lb. minimum specification requirement and the 20 ft lb. temperature being relatively high at -63 °C. The structure, figure 27, page 105, was predominantly auto-tempered martensite containing a dense but uniform dispersion of fine temper carbides within the martensite laths. The method of crack nucleation in such a structure has been discussed in sections 2.3.4.3., and 2.3.5.3. Crack propagation would tend to be hindered when lath boundaries and auto-tempered carbides are encountered causing the cleavage crack to change plane with associated plastic tearing and energy absorption. Virtually complete homogenisation of the austenite appeared to have occurred but local variation in carbon and other alloying element content is held responsible for the presence of isolated regions of mixed but predominantly upper bainite through which crack propagation would be relatively easier than in the surrounding auto-tempered martensite matrix. The rather poor impact properties in this cycled condition are most likely a function of the relatively large prior austenite grain size associated with the inherently high 'quenched-in' dislocation density within individual auto-tempered martensite laths. Compared with the 20 ft lb. temperature of + 40°C for the grain coarsened structure in QT-35 steel (224) obtained in simulated specimens thermally cycled to reproduce the conditions associated with a weld heat input of 108kJ. in⁻¹., the grain coarsened region in HY-80 steel was found to be very much superior as regards fracture toughness. In the case of QT-35, however, the structure is predominantly upper bainitic which, as discussed in section 2.3.4.1., has a poor resistance to cleavage fracture propagation. SMITH (243) is at present simulating structures in QT-35 steel using the same thermal cycles as those employed in this research and his results when compared with those presented herein may assist in the correlation between fracture behaviour and the relative proportions of upper bainite and auto-tempered martensite present in a low alloy steel, grain coarsened structure.

5.6. STRUCTURE AND PROPERTIES PRODUCED BY DOUBLE CYCLE SIMULATION

This section of the discussion considers the effects of various second cycles on the three initial cycles discussed in the previous section.

5.6.1. 1275°C peak temperature initial cycle

The structure produced by a second 1275°C peak temperature cycle, figure 28, page 106, was similar to that described in section 4.5.1., for the single cycled condition except that the prior austenite grain size was somewhat larger. This could account for the lower upper shelf energy of 26 ft lb. in the double cycled condition. The mode of fracture propagation appeared similar to that in the single cycled condition although this would require fractographic study for clarification. The hardness and U.T.S., were slightly lower than for the single cycled condition which could be accounted for by the larger prior austenite grain size produced by double cycling. The slightly higher 0.2% proof stress could be attributed to a higher 'quenched-in' dislocation density in the auto-tempered martensite laths associated with a somewhat greater interstitial solute content.

Employing a second 930°C peak temperature cycle produced a structure of auto-tempered martensite containing a few mixed bainitic colonies, figure 29, page 107. The auto-tempered martensite laths were smaller and more irregular in shape than those formed during the single 1275°C peak temperature cycle. Two ranges of prior austenite grain size were evident; a larger one due to the first high temperature cycle and a finer one which owed its existence to the presence of many more potential nucleation sites during the second lower temperature cycle. The highly irregular nature of the structure suggests the presence of a higher short range stress pattern than in the single cycled 1275°C condition which could give more potential sites to cause easier fracture initiation when long range stresses are imposed during testing. However the increased number of obstacles, in the form of small auto-tempered martensite lath boundaries, the fine carbide dispersion within such laths and the prior austenite grain boundary pattern still existing from the initial cycle could be expected to make fracture propagation more difficult. The impact transition curve for this condition must be interpreted as constituting a balance between these factors. With the non-instrumented Charpy test method employed in this research it is impossible to differentiate between the energy required for fracture initiation and that for propagation. Thus the reasons for the lower upper shelf energy in this double cycled condition and the 'cross-over' with the 1275°C single cycled impact transition curve at about - 38°C, figure 21, page 97, to give a lower 20 ft lb. temperature cannot be fully reconciled in terms of the structure observed. The increase in hardness and U.T.S., are most likely a result of the finer inherent grain size present.

A second cycle of 765°C gives the best overall properties produced by recycling material originally cycled to a peak temperature of 1275°C. The structure produced consisted of small,

irregular auto-tempered martensite laths in a virtually continuous network form delineating the prior austenite grain boundaries due to the initial cycle as shown in figure 30, page 108. Within this network internally veined ferrite laths were separated by an auto-tempered martensite product. The formation of this structure may be envisaged as follows. At the prior austenite grain boundaries, due to the first cycle, the ends of auto-tempered martensite laths from adjacent grains meet and form high energy sites for the nucleation of austenite during the second cycle. Diffusion of carbon from the auto-tempered carbides within the laths would be greatest at the lath tips, where the surfaces have a rounded form, into the austenite nucleated between adjacent grains. This gives a means for rationalising the existence of the auto-tempered martensite network. Austenite would also be nucleated along the lath boundaries, within the prior austenite grains, and by diffusion of carbon to these regions inter-lath auto-tempered martensite could also form. After tempering, as discussed in section 5.7.3., to follow, the ferrite laths, produced by this process, showed some very fine carbide precipitation which was indicative of the fact that not all the carbon from the auto-tempered carbides diffused to the lath boundaries during the second cycle treatment and that some remained to super saturate the ferrite. That the impact properties are superior to the single 1275°C peak temperature cycled condition, even in the presence of a prior austenite grain boundary constituent, may be considered to be a function of two factors:

- (a) the ferrite lath/auto-tempered martensite aggregate, within the auto-tempered martensite network, could act in a similar manner to upper bainitic structures in that crack nucleation would be relatively difficult (62). However, during the propagation phase the structures would behave differently since the small inter-ferritic lath auto-tempered martensite units would make crack propagation more difficult, as already discussed, whereas the virtually continuous stringer carbides at ferrite lath boundaries in upper bainite provide little resistance to propagation since they fracture readily with little energy absorption.
- (b) although crack initiation could occur in the auto-tempered martensite network, the fine, individual lath size and carbide dispersion present within the laths would make crack propagation difficult.

The 0.2% proof stress was lower than for the two duplex cycles already discussed and this could be attributed to the proportion and morphology of the relatively carbide-free ferrite present even if this, as has been suggested, is supersaturated to some extent with carbon. The relatively high standard deviation obtained in hardness testing is a function of the distribution of

the phase types in the duplex structure.

A 650°C peak temperature second cycle produced carbide precipitation at auto-tempered martensite lath boundaries, some degree of growth of auto-tempered carbides and additional carbide precipitation within the laths and an almost continuous network at the prior austenite grain boundaries as shown in figure 31, page 109. The 0.2% proof stress and hardness of this structure were higher and the impact properties poorer than for the 765°C peak temperature second cycle condition. The presence of the fine carbide dispersion within the laths and the precipitate at the lath boundaries may account for the increased tensile properties. That the impact properties are inferior can be attributed to the presence of both the inter-lath and prior austenite grain boundary carbide precipitates. It is tentatively suggested that, due to this carbide morphology, crack propagation could occur with less energy absorption than in the case of the 765°C second cycle condition.

5.6.2. 930°C peak temperature initial cycle

Only partial austenitization was produced by a second 765°C peak temperature cycle. In prior bainitic regions there was evidence of only partial carbide solution in the austenite. The structure of fine grained ferrite, containing regions of a fine ferrite-carbide transformation product, figure 33, page 112, produced an improvement in impact performance and a reduction in the 0.2% proof stress. A marked reduction in the proportion of the martensitic transformation product, when compared with the single cycled condition, could also have caused significant improvement when considered in conjunction with these structural changes. The lower standard deviation obtained on hardness testing indicated that the effect of banding was not nearly so marked in this condition.

Re-cycling to 650°C caused the 0.2% proof stress to increase over that produced by the 765°C peak temperature second cycle although the U.T.S., was similar. This behaviour may be accounted for by the tempering effect of this cycle causing a fine carbide dispersion to be formed in martensitic bands produced by the initial 930°C peak temperature cycle and the absence of regions of almost carbide-free ferrite, figure 34, page 113. The overall finer and more uniform carbide dispersion in this structure could make crack propagation more difficult and so explain the improved impact properties over the single 930°C or double 930/765°C cycled conditions.

5.6.3. 765°C peak temperature initial cycle

The 650°C peak temperature second cycle produced impact properties which were very similar to those of the parent material.

The reduction in 0.2% proof stress may be due to the precipitation of small rod-like carbides, figure 36 (c), page 115, within regions which had transformed to martensite during cooling after partial austenitization.

5.7. STRUCTURE AND PROPERTIES PRODUCED BY POST CYCLE HEAT TREATMENT.

Post cycle heat treatment in some instances produced similar structures and properties in groups of simulated cycle conditions and the discussion in this section will be in this context.

5.7.1. 1275°C, 1275/1275°C and 1275/650°C peak temperature conditions

The upper shelf energies for these conditions were all of the order of 70 ft lb. However, the double 1275/1275°C peak temperature cycle gave a 20 ft lb. temperature of -144°C, 12°C higher than that for the single 1275°C cycle. This somewhat poorer impact performance could be attributed to the larger austenite grain size produced by the double cycling treatment. There is little to choose between the 1275°C and 1275/650°C peak temperature conditions as regards impact properties and it is evident that the advantage obtained by the latter in the as-cycled form, section 5.6.1., is negated by the additional tempering period at 650°C. When compared with the parent material the poorer impact performance exhibited by these structures may be reconciled in terms of grain boundary carbide precipitation and the basically larger and more irregular carbide dispersion present, figure 37, page 118. The results obtained from tensile testing were very similar for the three conditions.

5.7.2. 1275/930°C peak temperature condition

The finer prior austenite grain size distribution within the coarser one due to the initial cycle and the form of carbide precipitation associated with these two grain sizes after tempering, figure 38, page 119, suggests greater resistance to crack propagation than the structures discussed in section 5.7.1. The impact properties as revealed by the respective transition curves, figure 21, page 97, are in fact better. The slightly higher yield strength exhibited by this structure is consistent with this concept.

5.7.3. 1275/765°C peak temperature condition

The improvement in impact properties over the structures discussed in sections 5.7.1. and 5.7.2. may be regarded as a function of the difference in carbide distribution after tempering.

A carbide network was evident at the prior austenite grain boundaries, due to the first cycle, figure 39, but could not be considered continuous although it was much more massive than that in the 1275/650°C post cycle heat treated condition. The nature of the carbide precipitation gave evidence for the fact that after cycling and before heat treatment, the grain boundary inter ferrite lath constituent was auto-tempered martensite. The fracture characteristics may be regarded as being the outcome of a balance of factors. The heavy carbide precipitation at the lath boundaries would tend to assist crack propagation as in an upper bainitic structure (62). However, precipitation within the laths themselves and the complex carbide distribution in the prior austenite grain boundary network would tend to hinder propagation.

5.7.4. 930°C, 930/765°C, 930/650°C

The impact properties produced by tempering these cycled conditions showed a definite improvement over those exhibited by structures associated with an initial 1275°C peak temperature cycle. The more uniform carbide distribution present in the 930/765°C post cycle heat treated condition, figure 41, page 122, could account for the slightly better impact performance when compared with the other two 930°C peak temperature cycle conditions.

5.7.5. 765°C, 765/650°C

The 0.2% proof stress and U.T.S., were slightly lower than for the parent material but still within the limit of the specification (228). The impact properties were very similar to those of the parent plate.

5.8. THE SIGNIFICANCE OF THE WELD HEAT AFFECTED ZONE

Welding can produce a large number of defects e.g. lack of penetration, lack of fusion, slag inclusions, porosity and cracking. Many service failures originate from gross defects in welded joints generally due to faulty welding techniques and unsuitable material. The toe of an edge bead in a multi-run weld is a point of appreciable stress concentration and it is in this locality of the weld, as a result of the presence of the grain coarsened region of the heat affected zone, that the worst notch toughness properties exist in HY-80 steel.

The single cycle simulation studies suggested that in an edge bead heat affected zone, such as that produced by bead 6 in figure 12a, page 70, the susceptibility to brittle fracture, as indicated by the Charpy V-notch impact test, decreased with distance from

the fusion boundary towards the parent plate. If a melting temperature of approximately 1500°C is assumed for HY-80 steel then the grain coarsened region which experienced thermal cycle peak temperatures within the range 1275 to 1500°C might reasonably be expected, by virtue of a larger prior austenite grain size, to exhibit poorer impact performance than that shown by the 1275°C peak temperature study. The 930°C peak temperature cycle met, within the limits of scatter inherent in the Charpy V-notch impact test, the 50 ft lb. minimum specification requirement, so that for cycles within the grain refined region, with higher peak temperatures, this minimum would most certainly not be achieved. Cycles with peak temperatures of 765 and 650°C gave impact properties well above the specification limit. The results for single 650°C thermal cycling have not been reported since they were virtually identical to those of the parent plate; this is not unreasonable when the thermal excursion experienced during cycling was of extremely short duration when compared with the specified 5 hour tempering treatment received by the plate. The edge bead heat affected zone, may, therefore, be considered to consist, in terms of thermal cycle peak temperatures, of a region which experiences peak temperatures in excess of about 930°C which exhibits increasing susceptibility to heat affected zone cold cracking as the peak temperature increases towards about 1500°C and a region in which the peak temperatures are below about 930°C which shows decreasing susceptibility as the cycle peak temperature decreases.

The double cycling simulation procedure enables predictions to be made regarding the effect of the tempering bead technique on the heat affected zone of an edge bead. All of the second cycles imposed on material initially subjected to a 1275°C peak temperature cycle, which, as noted above, corresponds to a cycle experienced in the parent plate close to the fusion boundary of an edge bead, did not return the impact properties to the 50ft lb. minimum specification requirement. A second 765°C peak temperature cycle, however, gave an impact energy which closely approached this value. The 650°C peak temperature second cycle gave impact properties which were slightly inferior to those produced by the 765°C cycle at the lower service temperature limit for submarine operation of -35°C . The degree of tempering generally claimed to be produced by such a sub-critical cycle was not obtained. The most surprising result, however, is that a 930°C peak temperature second cycle caused a deterioration in impact properties and could lead to an increased susceptibility to brittle failure within the edge bead heat affected zone grain coarsened region. Misalignment of a tempering bead suggests, via the results of this research, that if thermal cycles with peak temperatures in excess of A_{c3} or lower than about 650°C are experienced within the grain coarsened region of the edge bead heat affected zone, they would be particularly ineffective in improving fracture toughness in the region of the toe of the weld. In the former instance an increased susceptibility to heat

affected zone cold cracking would be produced while the latter treatment, although providing some increase in such resistance, would not accomplish what was desired of it in terms of the specification requirement. The effect of second peak temperature cycles of 650 and 765°C on the grain refined structure of the edge bead heat affected zone would improve fracture toughness for initial peak temperatures in the region of 930°C and below. The effect of such cycles on the grain refined region due to initial cycles in excess of 930°C has not been studied but it may be inferred from the results obtained that the higher the initial peak temperature the lower will be the effectiveness of a sub-Ac₃ second cycle in improving susceptibility to brittle fracture. Towards the grain refined - grain coarsened region boundary the fracture toughness would approach that produced by sub-Ac₃ peak temperature second cycles on grain coarsened structures.

If means were available for the accurate positioning of the tempering bead during deposition, which is highly unlikely in practice, then, for the welding conditions studied in this research, the cycle peak temperature range which should be experienced within the grain coarsened region of the edge bead heat affected zone, due to the tempering bead, would appear to be $725 \pm 25^\circ\text{C}$ for the maximum effect. If this temperature range was experienced within the grain coarsened region then lower peak temperature cycles would be experienced within the original edge bead grain refined region. The results suggest that provided a second cycle peak temperature of at least 650°C could be experienced towards the grain refined - grain coarsened region boundary then the tempering bead technique would confer properties to meet the specification requirement within the grain refined region subjected to initial peak temperature cycles in excess of 930°C. The deposition of the tempering bead using a higher heat input than that used for the previous multi-run layer deposition could be beneficial in that the time spend during cycling in the region of 725°C within the original grain coarsened region would be greater and, due to the wider heat affected zone associated with the higher energy input, the possibility of producing high sub-critical peak temperature cycles in the initial high peak temperature grain refined region would be increased. However, accurate positioning of the tempering bead in such an instance would be of extreme importance in order to alleviate the possible occurrence of undesirable structures, caused by slower rates of cooling, for peak temperatures above the critical range, should these occur.

The occurrence of the double 1275°C peak temperature cycle sequence, with the production of a structure with inherently poor resistance to brittle fracture, can only be envisaged in an edge bead heat affected zone if for the same energy input used for edge bead deposition the tempering bead completely overlaps the edge

bead or in the case of a higher energy input only partial overlap occurs. However, it is possible that such a cycle sequence could be produced in the heat affected zone in the interior of a weld due to the overlap of a subsequent run on a pre-existing one. This situation could occur for example due to the overlap of bead 6 on bead 3 in figure 12a, page 70. Thus even within the weld itself regions highly susceptible to cold cracking can exist within the heat affected zone; this point is not always appreciated when multi-run weld heat affected zones are considered. The other double cycle sequences investigated are also possible and the results already discussed for these are equally applicable in such a situation. The heat affected zone in the weld interior presents a complex situation where the multiplicity of cycle sequences which can occur are strongly dependent upon bead geometry, run sequence, energy input of successive runs etc.

Post cycle heat treatment of both the single and double cycled simulated specimens indicates that if sub-critical heat treatment was carried out subsequent to welding, marked improvement in the properties of the edge bead or edge bead plus tempering bead heat affected zones would be achieved. This could be accomplished by laying electric strip contact resistance heaters on the heat affected zone, at the plate surface, along the weld length and maintaining a slowly decreasing temperature gradient, away from the visible heat affected zone, in the unaffected parent material. Where possible in practice this technique is suggested for use due to its superiority and certainty of producing the required edge bead heat affected zone fracture toughness when compared with the tempering bead technique. The only method of ensuring that the whole of the heat affected zone due to a multi-run weld in HY-80 steel has a high degree of resistance to brittle fracture would be to resort to furnace heat treatment of the welded fabrication.

Attention has recently been drawn by WINN (253) to the properties of the weld metal. Information regarding the performance of a heat affected zone in connection with brittle fracture is completely satisfactory if the weld metal has adequate strength, good impact properties and freedom from defects. However, if the weld metal does not meet these requirements then no matter how good the properties of the heat affected zone in the parent plate, failure could still occur. To obtain a complete understanding of the performance of a weldment made using a given set of conditions the characteristics of the weld metal must be ascertained in addition to those of the heat affected zone.

6. CONCLUSIONS

1. Single cycle simulation showed that as far as susceptibility to heat affected zone cold cracking was concerned high peak temperature cycling, as would be experienced within the grain coarsened region of the heat affected zone of a single weld run, had a detrimental effect. Compared with QT-35 steel, however, the results obtained for single cycled conditions in HY-80 steel were markedly superior.
2. The use of the double cycling technique has shown that single cycle simulation is somewhat inadequate in predicting the structures and associated properties in steels such as HY-80 which are normally multi-run welded using the tempering bead technique.
3. Peak temperature cycles in excess of about 900°C due to the application of a misaligned tempering bead would cause an increase in the susceptibility to heat affected zone cold cracking within the grain coarsened region of an edge bead heat affected zone. Although such susceptibility would be decreased by sub-Ac₃ peak temperature cycling the 50ft lb. minimum energy requirement is unlikely to be attained as shown by recycling to peak temperatures of 765°C and 650°C material initially cycled to 1275°C. The second 765°C peak temperature cycle gave a close approach to the specification minimum and it is suggested that, for maximum benefit, the edge bead grain coarsened region should undergo peak temperature cycles of the order of $725 \pm 25^\circ\text{C}$ when a tempering bead is applied. However, it is unlikely in practice that even a correctly aligned tempering bead would cause sufficient tempering to occur in the grain coarsened region of an edge bead heat affected zone.
4. Post cycle heat treatment at 650°C gave a marked improvement in impact properties for all the cycled conditions studied. The 0.2% proof stress was higher than for the parent plate except for the 765°C and 765/650°C post cycle heat treated conditions which, nevertheless, had values greater than the minimum specification requirement. It is recommended that wherever possible HY-80 weldments should receive post-weld heat treatment at this temperature for a time dependent upon plate thickness. This treatment should be employed even if the tempering bead technique has been used during welding to ensure adequate notch toughness in the edge bead heat affected zone adjacent to the toe of the weld.
5. In this study the ultimate structure produced by a given thermal cycle has been found to be markedly affected by the nature of the structure which existed prior to cycling. To ensure consistency of results in single cycle simulation studies of a given steel the initial structure should always be the same.

6. In analysing the results of simulation studies banding in the parent plate must be taken into consideration, since for a given peak temperature cycle a particular structure is found at a plane in an actual weld heat affected zone and not a volume of layered aggregate as in a simulated specimen. This is particularly so if the thermal cycle conditions are such as to cause transformation of adjacent bands to bainitic and martensitic structures. Caution must be applied if hardness traverses are used to obtain an estimate of weld heat affected zone properties since the use of too high a testing load will not reveal the presence of banding.

7. The use of electron microscopy is essential in the examination of weld heat affected zone structures in low alloy steels. The fine structural detail present cannot be resolved using optical microscopy and this study has shown that while the carbon extraction replica technique serves a definite purpose in determining the general nature of a given structure, thin foil electron microscopy could be useful in many instances to obtain an accurate assessment of the mode of transformation occurring.

8. The series of computer programs developed to obtain weld heat affected zone thermal cycles represents a significant advance in the theoretical prediction of such cycles. Functions representing the release of latent heat from the solidifying weld pool and the variation of thermal conductivity during the transformation of undercooled austenite have been successfully incorporated in the programs. Latent heat release must be considered if calculated thermal cycles are to be used as a basis for simulation studies due to its effect in increasing austenite homogeneity and degree of carbide solution prior to transformation in cycles which involve thermal excursion into the austenite region.

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APPENDIX 1

Tables

TABLE 1 Symbols used in heat flow equations

Symbol	Description	Units
x, y, z	cartesian co-ordinates	cm.
r	$(x^2 + y^2)^{\frac{1}{2}}$	cm.
R	$(x^2 + y^2 + z^2)^{\frac{1}{2}}$	cm.
g	plate thickness	cm.
t	time	sec.
Q'	rate of heat flow from line source	cal.sec. ⁻¹
Q	rate of heat flow from point source	cal.sec. ⁻¹
T	temperature	°C
T_o	preheat temperature	, °C
k	thermal conductivity	cal.cm. ⁻¹ sec. ⁻¹ °C ⁻¹
ρ	density	gm.cm ⁻³ .
c	specific heat	cal.gm ⁻¹ °C ⁻¹
α	thermal diffusivity	cm ² .sec ⁻¹ .
λ	0.5x (reciprocal of thermal diffusivity)	sec.cm. ⁻²
$K_o(\lambda vr)$	modified Bessel function of the second kind and zero order with argument λvr	
subscript o as in k_o and $(\rho c)_o$	evaluated at reference temperature	
η	arc efficiency	

TABLE 2 Comparison of the chemical composition requirements for HY-80 steel specified by MIL-S-16216D (229) and MIL-S-16216G (228).

Specification Requirements		Military Specification		
		Weight Percentage*		
		MIL-S-16216D ⁺		MIL-S-16216G
Nominal Plate Thickness lb.ft ⁻²		<56.1	>51.0	-
Major Alloying Elements	C	0.22	0.23	0.18
	Mn	0.10-0.40	0.10-0.40	0.10-0.40
	S	0.040	0.040	0.025
	P	0.035	0.035	0.025
	Si	0.15-0.35	0.15-0.35	0.15-0.35
	Ni	2.00-2.75	2.50-3.25	2.00-3.25
	Cr	0.90-1.40	1.35-1.85	1.00-1.80
	Mo	0.23-0.35	0.30-0.60	0.20-0.60
Residual Elements	Ti	-	-	0.02
	V	-	-	0.03
	Cu	-	-	0.25
Microstructure (Minimum)		-		80% † Martensite

* Maximum percentage unless a range is quoted.

+ For nominal thicknesses between 51.0 and 56.1 lb.ft⁻², either composition may be supplied.

† Criterion at mid-plate thickness.

TABLE 3 Mechanical property requirements for HY-80 steel plate specified by MIL-S-16216G (228).

		Nominal Plate Thickness, lb.ft ⁻²	
		<20.4	>20.4
Ultimate Tensile Strength,* lbf.in ⁻²		-	-
Yield Strength at 0.2% Offset, lbf.in ⁻²		80,000 to 100,000	80,000 to 95,000
Minimum Elongation on 2 in, %		19	20
Reduction in area, %	Longitudinal	-	55
	Transverse	-	50

*Supplied by manufacturer for information only.

TABLE 4 Charpy V-notch impact requirements for HY-80 steel plate specified by MIL-S-16216G (228).

Nominal Plate Thickness lb.ft ⁻²	Specimen Size, mm.	Energy, ft.lb. ⁺ (Minimum)	Test Temperature, °C [†]
10.2 to 20.4 exclusive	10 x 5 half size	- *	-84
20.4 to 61.2 inclusive	10 x 10 standard	50	-84
Over 6.12	10 x 10 standard	30	-84

⁺ Average of three test values. No single value should be more than 5 ft.lb., below the minimum average required.

* Impact tests required for information only. Tests are not required for plate less than 10.2 lb.ft⁻²

[†] Test temperature shall be -84°C $\pm$ 3°C.

TABLE 5 Mechanical property data for as-cycled simulated specimens and the parent material.

Material Condition		Tensile Properties			Impact Test Criteria			Hardness, HV5	
1st Cycle	2nd Cycle	0.2% Proof Stress -2 tonf.in	UTS -2 tonf.in	Reduction of Area %	Upper Shelf Energy ft lb.	20ft lb. Temperature °C	50% FATT °C	Mean*	Standard† Deviation ±
1275	-	61.0	86.3	51	37	-63	-42	423	7.3
1275	1275	65.1	82.6	53	26	-64	-27	412	6.9
1275	930	66.9	87.7	58	30	-120	-71	437	7.2
1275	765	51.5	66.4	57	43	-134	-83	308	17.6
1275	650	60.4	68.1	61	45	-112	-62	352	5.9
930	-	59.1	70.1	62	49	-133	-81	353	17.4
930	765	43.9	57.4	62	53	-156	-114	318	10.7
930	650	50.1	57.2	66	68	-194	-127	279	11.5
765	-	39.1	49.0	72	96	-178	-132	278	8.9
765	650	37.5	46.8	71	93	-176	-142	234	7.1
Parent	Transverse	37.4	46.7	72	91	-176	-128	228	3.9
	Longitudinal	38.6	47.8	70	116	-189	-139	230	3.7

$$*\text{Mean} = \bar{x} = \frac{\sum x_i}{n}$$

$$†\text{Standard deviation} = s = \frac{\sum (x_i - \bar{x})^2}{n}$$

TABLE 6 Mechanical property data for post cycle heat treated simulated specimens.

Material Condition		Tensile Properties			Impact Test Criteria			Hardness, HV5	
1st Cycle	2nd Cycle	0.2% Proof Stress ₂ tonf.in	UTS tonf.in ⁻²	Reduction of Area %	Upper Shelf Energy ft lb.	20ft lb. Temperature °C	50% FATT °C	Mean*	Standard† Deviation ±
1275	-	47.8	54.4	70	70	-152	-106	268	4.9
1275	1275	48.8	53.7	72	71	-144	-101	272	5.2
1275	930	50.0	53.6	63	72	-166	-133	273	4.3
1275	765	45.8	52.8	70	73	-171	-126	255	6.1
1275	650	48.3	53.0	68	68	-155	-108	273	3.9
930	-	39.5	45.9	74	85	-175	-131	231	2.9
930	765	38.6	45.3	73	89	-181	-146	238	4.8
930	650	39.4	45.8	72	93	-176	-136	232	2.8
765	-	36.7	45.8	78	94	-173	-134	230	3.1
765	650	36.1	44.1	73	92	-175	-143	228	3.3

$$*\text{Mean} = \bar{x} = \frac{\sum x_i}{n} \quad + \quad \text{Standard Deviation} = \frac{\sum (x_i - \bar{x})^2}{n}$$

APPENDIX 2

The Calculation of Weld Heat Affected Zone Thermal
Cycles using an ICL 1905 series Computer.

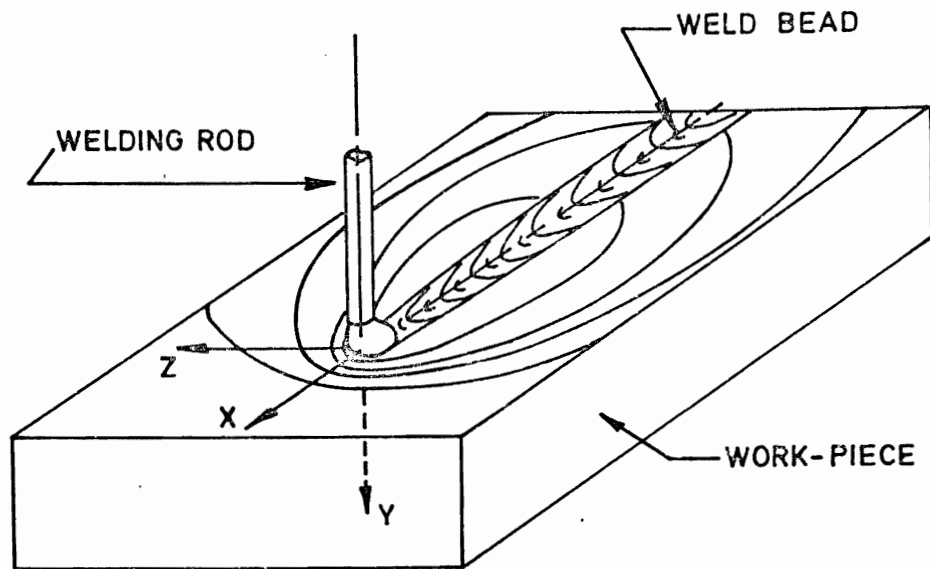
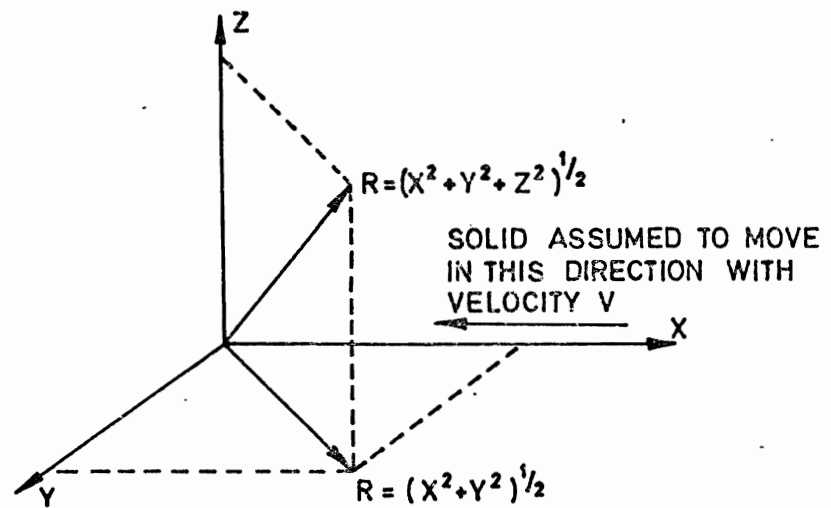


FIGURE 49 MOVING CO-ORDINATE SYSTEM IN THE WORK-PIECE DURING WELDING

1. Introduction

As discussed in section 3.3. it was necessary in this research to employ computational techniques to obtain thermal cycles typical of those experienced at certain points within the heat affected zone. The analytical solution, due to GROSH et al., (12) to the partial differential equation for 3-dimensional heat flow was chosen for computation since it has the inherent characteristic of permitting variation of the thermal properties with temperature. This solution is recalled from section 2.1.:

$$T = \frac{1}{m} \left\{ \left[\frac{Q_m}{2\pi k_0 R} \exp(-\lambda v x) \exp(-\lambda v R) + (1 + m T_0)^2 \right]^{\frac{1}{2}} - 1 \right\} \quad 1$$

The co-ordinate system to which this solution refers is given in figure 49 and the symbols used are in accordance with Table 1, page 165.

Since $x = vt$, equation 1 may be written as:

$$T = \frac{1}{m} \left\{ \left[\frac{Q_m}{2\pi k_0 R} \exp(-\lambda v (vt + R)) + (1 + m T_0)^2 \right]^{\frac{1}{2}} - 1 \right\} \quad 2$$

This form of the solution was incorporated in the computer programs to be discussed in sections 2 to 6 inclusive.

As discussed in section 2.1., the thermal properties were assumed to vary as:

$$k = k_0 f'(T) \quad 3$$

$$\text{and } \rho c = (\rho c)_0 f'(T) \quad 4$$

where $f'(T)$ is the temperature derivative of some function of temperature. Linear variation of the thermal properties was chosen such that:

$$k = k_0 (1 + mT) \quad 5$$

$$\text{and } \rho c = (\rho c)_0 (1 + mT) \quad 6$$

Since the initial stage of the computational phase of this study involved comparison with curves measured by SMITH (243), at points in the heat affected zone directly below the weld centre line, it was decided to set z equal to zero so that:

$$R = (x^2 + y^2)^{\frac{1}{2}} = (v^2 t^2 + y^2)^{\frac{1}{2}} \quad 7$$

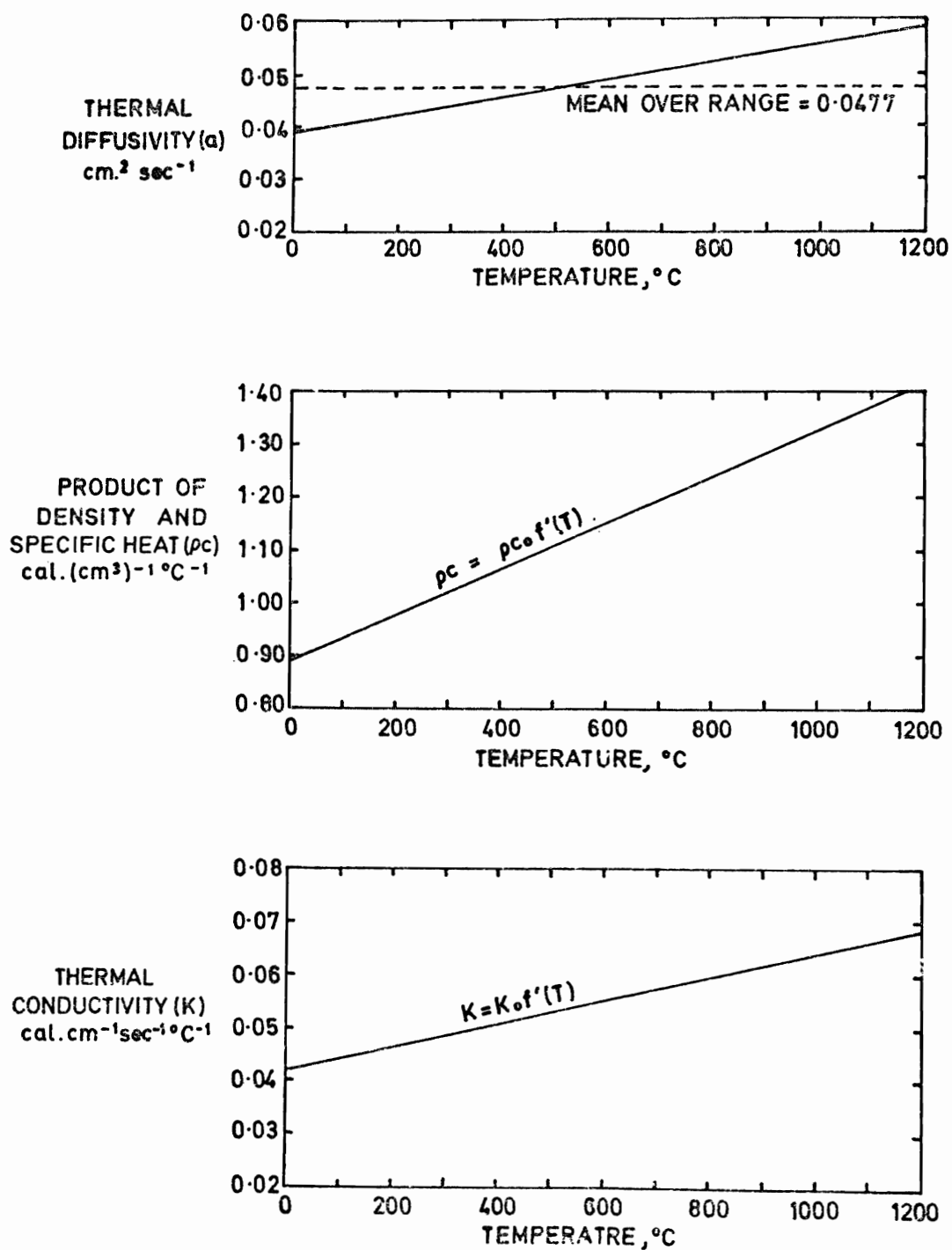


FIGURE 50 THERMAL CONDUCTIVITY, PRODUCT OF DENSITY AND SPECIFIC HEAT, AND THERMAL DIFFUSIVITY FOR AN AUSTENITIC STEEL.

The heat flow from the point source was given by:

$$Q = 0.24 UI\eta \quad 8$$

where U is the arc voltage (volt), I the welding current (amp.), η the efficiency of heat transfer to the weldment and 0.24 a conversion factor (cal. watt sec⁻¹). CHRISTENSEN et al., (27) quoted an efficiency range of 0.91 to 0.99 for the submerged arc welding of mild steel. Since the measurements made by Smith were carried out under controlled laboratory conditions this present computer study employed an efficiency of 0.98.

In the initial stages of computation the steel was assumed to be austenitic over the range of temperatures concerned and the data of POWELL (25), as shown in Figure 50, were used. A mean value for the thermal diffusivity, α , was found to be 0.0477 over the range of temperature concerned. This gave:

$$\lambda = \frac{1}{2\alpha} = 10.482$$

The slope of the thermal conductivity - temperature relation is given by $k_0 m$ which yields a value of 0.000514 for m. Including the welding conditions employed the values used in this computer study were:

U volt	I amp.	V in.min. ⁻¹	η	λ sec.cm. ⁻²	m	k_0	T_0 °C
30	500	16.66	0.98	10.482	0.000514	0.042	120

V is the electrode travel speed, T_0 the preheat temperature.

The VARTHERM series of computer programs which will be described in the sections which follow are written in Algol 60 programming language (245) and were used in conjunction with an ICL 1905 digital computer, having 32K words of core store and card and tape input/output facilities.

2. Program VARTHERM 1.

A system was employed whereby a print-out of thermal cycle peak temperatures with their corresponding distances from the point source was obtained. Program VARTHERM 1, pages 188 and 189 satisfies this requirement and computes 90 values of y corresponding to 90 peak temperature values. In this, as in the other programs which follow a 'rogue-value' format was used such that any amount of required data could be read by the computer. The data cards presented to the computer were:

card 1: 0.000514

card 2: 30" 500" 0.98" 16.66" 0.042" 120" 10.482

The symbol ' ' indicates a punch space on the card.

From these data cards the print-out given on page 197 was obtained and from this, values of y corresponding to peak temperatures of interest were ascertained;

Cycle Peak Temperature, °C	1275	930	765	650
y , cm.	0.637	0.792	0.908	1.015

The following additional information was given:

Heat flow Q Cal.sec ⁻¹	Speed V cm.sec ⁻¹	Nominal Heat Input, Q_n kJ.in ⁻¹	Actual Heat Input, Q_I kJ.in ⁻¹
3528	0.7056	54.0	52.92

3. Program VARTHERM 2.

Using values for y corresponding to peak temperatures of 1275, 930, 765 and 650°C, obtained from VARTHERM 1, program VARTHERM 2, page 190, was run to obtain complete thermal cycles according to equation 2. The data cards fed to the computer were;

card 1: 0.0000514

card 2: 3528" 0.042" 10.482" 0.7056" 120

card 3: 0.637" 0.792" 0.908" 1.015" 1.159

The print-out obtained is given on pages 198 to 201. Figures 51 and 52 compare cycles computed by VARTHERM 2 and other programs with those measured by Smith for cycle peak temperatures of 1275 and 765°C. This comparison shows a major discrepancy in the region of 1100°C for the 1275°C peak temperature cycle and a generally poor correlation for both peak temperature cycles.

4. Program VARTHERM 3.

As is evident from figure 51, for any time t the temperature indicated by the measured curve is higher than the temperature calculated by VARTHERM 2 and a form of inflection occurs at about

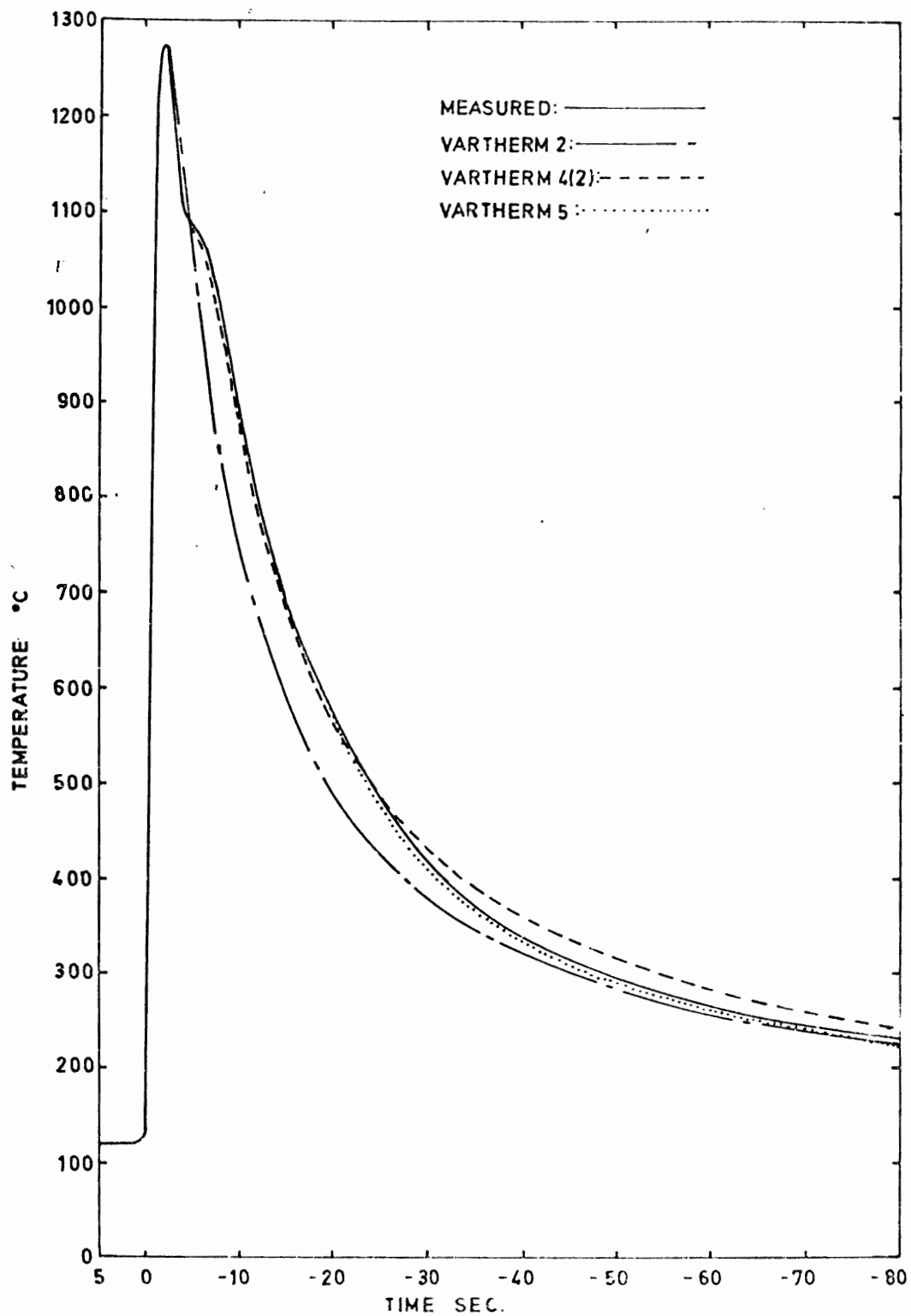


FIGURE 51 COMPARISON OF MEASURED AND COMPUTED THERMAL CYCLES FOR 1275°C PEAK TEMPERATURE.

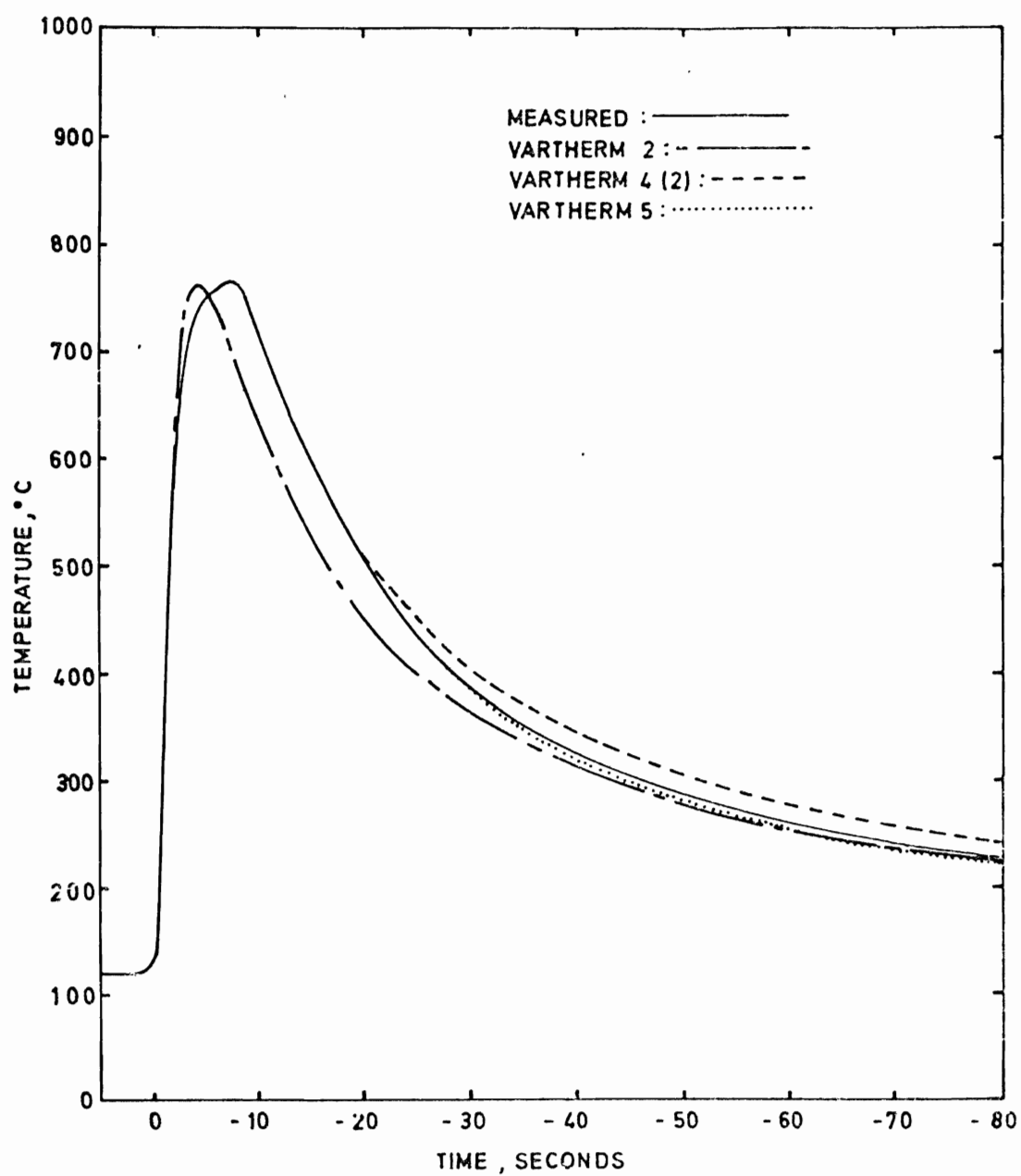


FIGURE 52 COMPARISON OF MEASURED AND COMPUTED THERMAL CYCLES FOR 765°C PEAK TEMPERATURE.

1100°C which causes the two curves to deviate. A model was invoked to provide an explanation for this effect viz., that a secondary point source, separated by some distance x' , follows the primary source but that the heat flow due to this former source is small by comparison with the latter and is released in a gradual but increasing manner until some maximum value is attained. APPS and MILNER (10) suggested that, due to convection currents within the molten weld pool, the major proportion of latent heat release occurred at the 'tail' of the pool.. This concept was included in the model and the secondary point source was assumed to be due to latent heat release. Program VARTHERM 3, page 191 was devised to investigate the nature of the secondary heat flow.

A modified version of equation 2 is employed in VARTHERM 3:

$$Q = \frac{2 \pi k_o R \exp(\lambda v(vt+R))}{m} \left[(1+mT)^2 - (1+mT_o)^2 \right] \quad 9$$

where T and T_o are the temperatures indicated by the measured and computed (VARTHERM 2) cycles at some time t . T_o is used as an 'effective preheat temperature' such that Q is the increment of heat flow required at any time t to raise the temperature to T for a given value of y . The results obtained from feeding 36 values of T and T_o , obtained from figure 51, into the computer suggested that the secondary or latent heat flow was released according to the sigmoidal function:

$$L = L_M (1 - \exp(-t_D^n/J)) \quad 10$$

where L is the instantaneous and L_M the maximum values, respectively, of secondary heat flow, t_D the time, J a positive constant and n a positive integer.

The delay time between the sources t_s is given by x'/v and for computing, equation 10 is modified to:

$$L = L_M (1 - \exp(-(t + t_s)^n/J)) \quad 11$$

5. Program VARTHERM 4.

VARTHERM 1 was modified to include the latent heat function, equation 11, and entitled VARTHERM 4, page 192. This program was used to obtain another plot of cycle peak temperature and corresponding distance from the point source, and to investigate any variation due to the heat flow from the secondary source. Initially, for some time t , the temperature is calculated using equation 2 and then the delay time t_s is incorporated. As suggested in section 4,

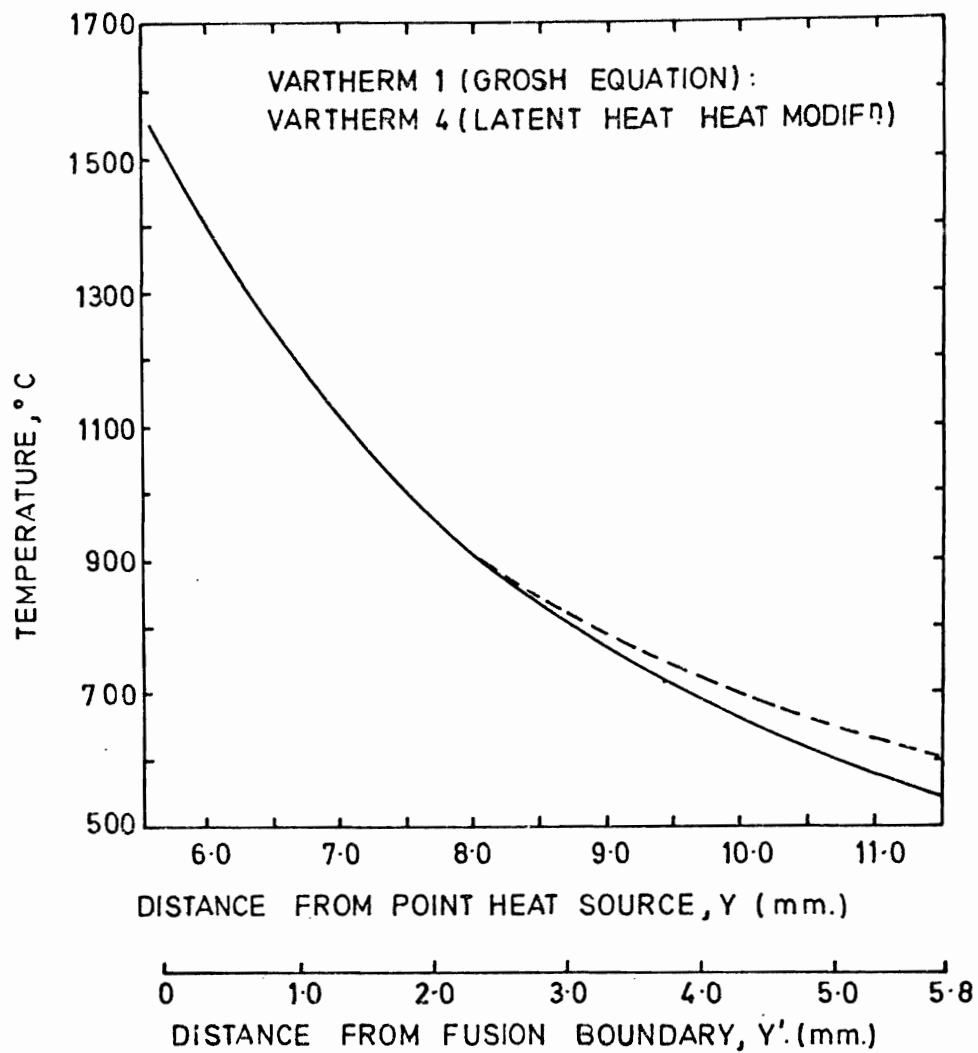


FIGURE 53. PREDICTED EFFECT OF LATENT HEAT OF FUSION ON THE PEAK TEMPERATURE DISTRIBUTION IN THE PARENT MATERIAL.

$$t_D = t + t_s$$

In VARTHERM 4, t_D may only be introduced into equation 11 if $(t + t_s)$ is equal to or less than zero and equation 11 is only applicable if the exponent n is a positive, even integer. If $t_D \leq 0$ the temperature obtained initially is recalculated by first employing it as the preheat temperature in equation 2 and obtaining modification via the latent heat function, equation 11. By a process of comparison within the programme the maximum temperature is found.

The data cards supplied to the computer were:

card 1: 0.000514

card 2: 0.042" 0.7056" 10.482" 120" 3528" 650" 3.5" 10.

In terms of this data, the form of equation 11 used in VARTHERM 4 was:

$$Q_2 = 650 (1 - \exp((t + 3.5)^2/10)) \quad 12.$$

Figure 53 compares the peak temperature - distance distributions obtained from programs VARTHERM 1 and VARTHERM 4. Two distance scales are incorporated in this figure, one to indicate distance from the point source and one the distance from the assumed fusion boundary. The distances corresponding to the peak temperatures of interest were inferred from the print-out given on page 202 and were:

Peak Temperature, °C	1275	930	765	650
y, cm.	0.637	0.792	0.931	1.075

6. Program VARTHERM 4(2)

This program incorporates the latent heat function, equation 11, and was run, using values of y obtained from VARTHERM 4, to obtain complete thermal cycles. Program VARTHERM 4(2) is given on page 194 and the print-outs obtained on pages 203 to 206. The data cards used were:

card 1: 0.000514

card 2: 3528" 0.042" 10.482" 0.7056" 120" 650" 10

card 3: 0.637" 0.792" 0.931" 1.075" 1.09

Figures 51 and 52 compare cycles calculated by this program, for peak temperatures of 1275 and 765°C with those computed by VARTHERM 2 and those measured by Smith. For the 1275°C peak

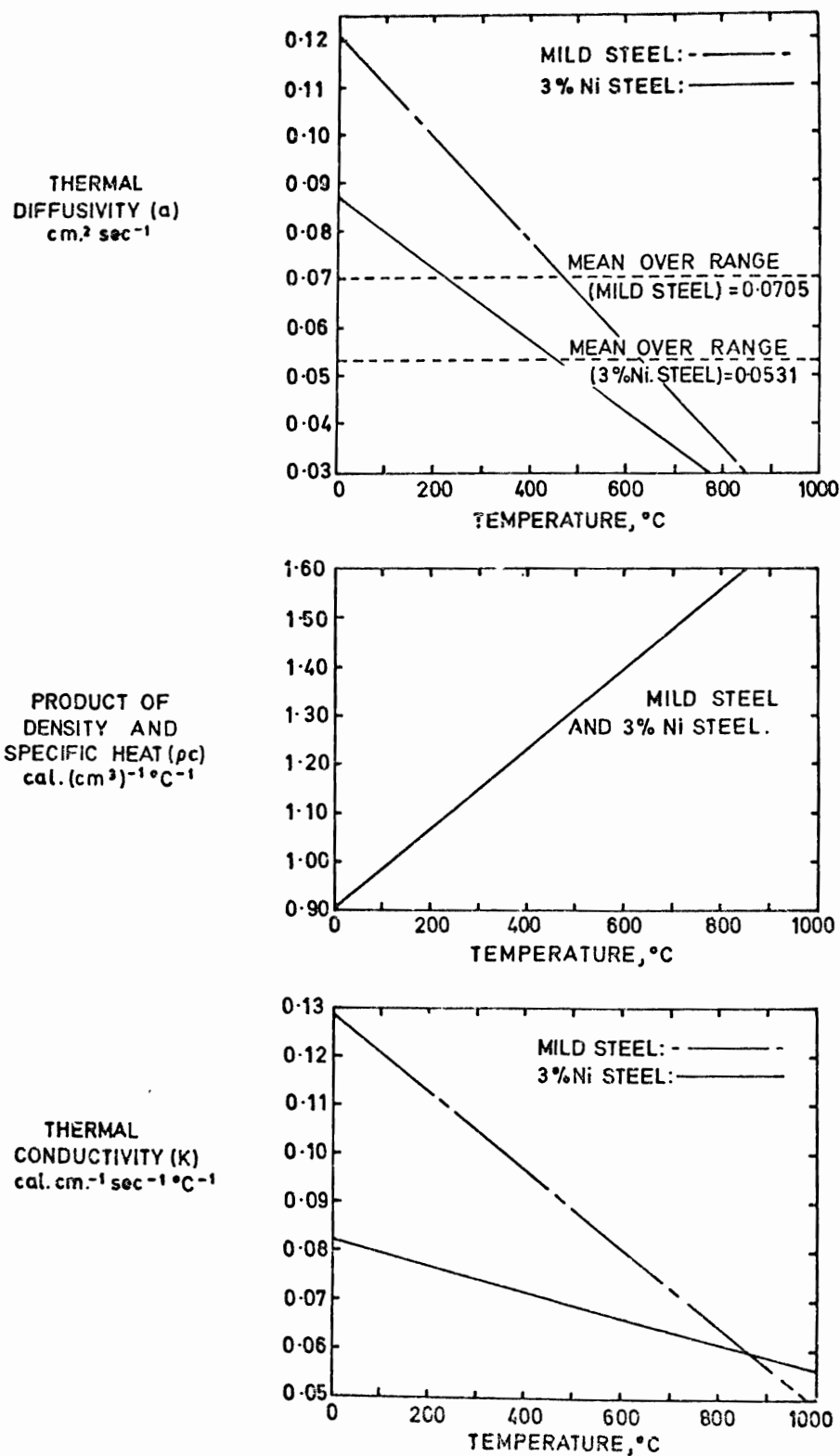


FIGURE 54. THERMAL CONDUCTIVITY, PRODUCT OF DENSITY AND SPECIFIC HEAT, AND THERMAL CONDUCTIVITY FOR MILD STEEL AND A 3% NICKEL STEEL.

temperature cycle very good correlation is obtained, including the inflection at about 1100°C, down to about 500°C after which the computed temperatures are higher than those obtained by measurement.

7. Program VARTHERM 5

It was evident from the results obtained from running program VARTHERM 4(2) that some form of program modification was necessary in order to obtain correlation with the measured 1275°C peak temperature cycle for temperatures below about 500°C. This is essentially a 'curve fitting' exercise but, in order to have a theoretical basis upon which future development could be made, it was assumed that a variation in thermal conductivity occurred at low temperatures during the transformation of undercooled austenite to bainitic and martensitic structures. In the absence of any published data to the contrary, this assumption appears valid. For austenite starting to transform at the bainite start temperature, B_s , then the thermal conductivity between B_s and room temperature is taken to vary according to the polynomial function:

$$k = AT^n + BT^{n-1} + CT^{n-2} + DT^{n-3} + \dots \quad 13$$

A straightforward quadratic form of this function was found to be suitable:

$$k = AT^2 + BT + C \quad 14$$

Thermal property data for a 3% nickel steel (246) are shown in figure 54; data for mild steel (247) are included for comparison. This data is markedly different from that for an austenitic steel given in figure 50. Figure 55 combines the data of figures 50 and 54 and shows the assumed variation of thermal conductivity with temperature during transformation to bainitic and martensitic structures below B_s . K_1 represents the thermal conductivity variation of a tempered ferrite-carbide aggregate on heating, K_2 the heating and cooling, including extrapolation for undercooling, of austenite and K_3 the transformation of austenite on cooling to bainitic - martensitic structures.

If the thermal conductivity of austenite is related to temperature by equation 5 viz.,

$$k = k_0 + k_0 m T$$

then the slope of the line at any temperature T is:

$$\frac{dk}{dT} = k_0 m \quad 15$$

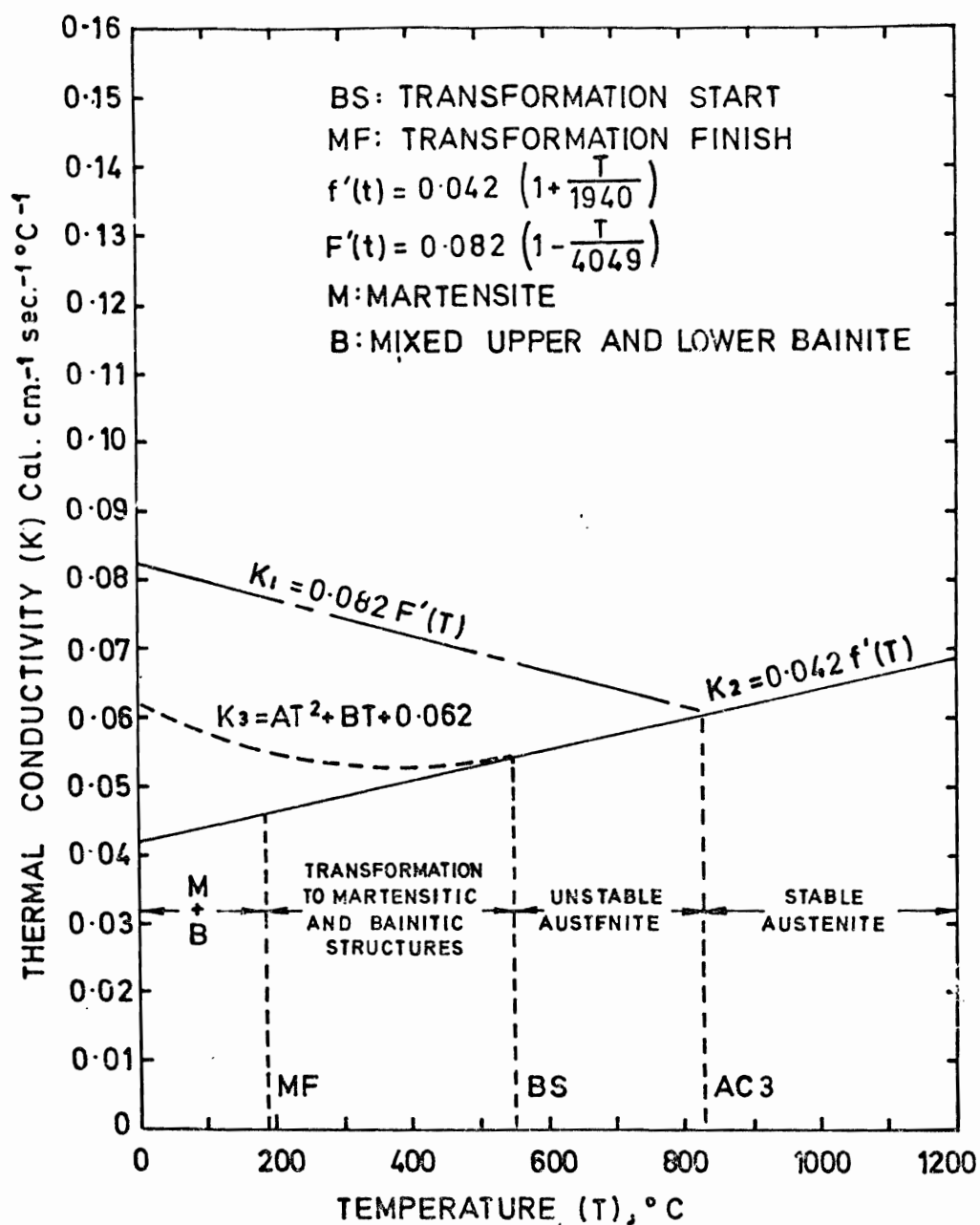


FIGURE 55 ASSUMED THERMAL CONDUCTIVITY VARIATION DURING HEATING (K_1 AND K_2) AND DURING SUBSEQUENT COOLING AND TRANSFORMATION OF AUSTENITE (K_2 AND K_3)

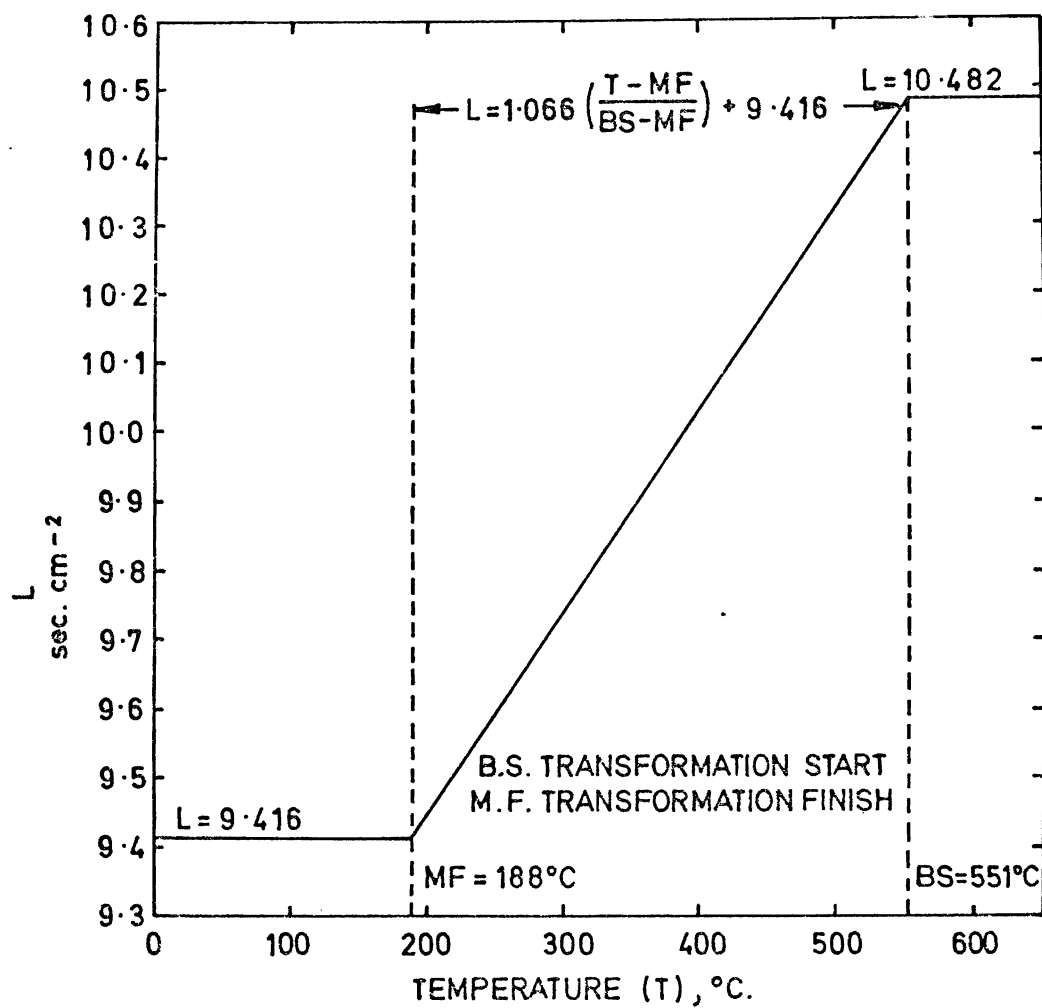


FIGURE 56. ASSUMED VARIATION OF L ($0.5 \times$ RECIPROCAL OF THERMAL DIFFUSIVITY, α) WITH TEMPERATURE IN THE RANGE OF AUSTENITE TRANSFORMATION.

At 0°C, the constant C in the quadratic function, equation 14, is seen to be merely the intercept k_{o2} on the thermal conductivity axis. The slope of the quadratic at any temperature T is:

$$\frac{dk}{dT} = 2AT + B \quad 16$$

In figure 55, the austenite line is tangent to the quadratic function at the B_s temperature. Thus:

$$k_o + k_{om} B_s = AB_s^2 + BB_s + k_{o2} \quad 17$$

$$\text{also } k_{om} = 2AB_s + B \quad 18$$

$$\text{so that } A = \frac{k_{o2} - k_o}{B_s^2} \quad 19$$

$$\text{and } B = k_{om} - 2AB_s \quad 20$$

By invoking relations 19 and 20 and noting that $C = k_{o2}$, the constants in the quadratic function may be found.

Program VARTHERM 5, pages 195 and 196 considers both the latent heat function and the low temperature thermal conductivity variation due to austenite transformation. Linear variation of λ (L in the computer program), figure 56, was assumed between B_s and the martensite reaction finish temperature, M_f . For $T > B_s$ and $T < M_f$ constant values for λ of 10.482 and 9.416 were obtained from figures 50 and 54 respectively. Using empirical formulae due to STEVEN and HAYNES (38) the B_s and M_f temperatures for the heat of HY-80 steel used in this research are:

$$B_s = 830 - 270C - 90Mn - 37Ni - 70Cr - 83Mo = 571^\circ\text{C}$$

Correction for continuous cooling gives $B_s = 551^\circ\text{C}$.

$$M_s = 561 - 474C - 33Mn - 17Ni - 17Cr - 21Mo = 403^\circ\text{C}.$$

Using the relation (38) $M_f = M_s - 215$, $M_f = 188^\circ\text{C}$.

The calculated values for B_s and M_f are quoted as giving 90% confidence to $\pm 20^\circ\text{C}$ of experimental values.

Thus in the temperature range of austenite transformation:

$$\lambda = 1.066 \left(\frac{T - M_f}{B_s - M_f} \right) + 9.416 \quad 21$$

It should be noted that the constants A and B are calculated as an initial stage during the running of the program.

VARATHERM 5 incorporates a thermal conductivity 'block' for use during cooling below the B_s temperature. Within this 'block' new values for k , k_0 , m and λ are calculated on the basis of the previously calculated temperature. This is a source of error but provided sufficiently small time intervals are employed in calculation it is virtually insignificant. The thermal conductivity k is first calculated in this loop using equation 14 followed by the calculation of the slope at some temperature T by equation 16. The new value of k_0 is obtained from:

$$k_0 = k - (2AT + B)T \quad 22$$

k_0 is the intercept at 0°C , on the thermal conductivity axis, made by the tangent to the quadratic function at temperature T . New values for m and λ are then obtained from equations 15 and 21 respectively. These values are used in the next time step in the 'time loop' to calculate another temperature and the whole process reiterated, for a given value of y , until the 'time loop' is terminated. After the print-out from data stored in the computer 'memory' and before further calculation using another y value, the initial values of m , k_0 and λ are reassigned.

To obtain the thermal cycles required in this research the data cards used were:

card 1: 0.000514

card 2: 3528" 0.042" 10.482" 0.7056" 120" 3.5" 650" 2" 10"
551" 188" 0.062

card 3: 0.637" 0.792" 0.931" 1.075" 1.09

The print-outs for VARATHERM 5 obtained using this data are given on pages 207 to 210. The quadratic function had the form: $K = 0.0000000659 T^2 - 0.0000510073T + 0.062$. Figures 51 and 52 compare computed and measured cycles for peak temperatures of 1275 and 765°C respectively. The close correlation shown in these figures provided the justification for using VARATHERM 5 to obtain the 930 and 650°C peak temperature cycles, shown in figure 15, for the simulation phase of this research.

8. Applicability of the VARATHERM series of Programs.

At present this program series is limited by the fact that for a given set of welding conditions and plate thickness, a high peak temperature thermal cycle must first be measured to permit calculation of constants for the latent heat and thermal conductivity functions. After such measurement and computation using the VARATHERM programs 1 to 4(2) inclusive, VARATHERM 5 may be employed to calculate the thermal cycle occurring at any point in the heat affected zone of the parent material due to welding.

Programme VARTHERM 1.

06/08/69

COMPILED BY XALE MK. 3C

```

'SEND TO'(ED,PROGRAM FILE,.STOPE)
'PROGRAM'(T03F)
'INPUT'1=CR0
'OUTPUT'2=LPO
'BEGIN''REAL'M,E,I,EF,Q,QI,SP,V,G,KO,T0,R,U,S,W,L,D,K,QN;
      'INTEGER'F,P,PR,Z,H;
      'REAL''ARRAY'TIM[1:310],TEMP[1:310],TIMMAX[1:90],
      TEMMAX[1:90],Y[1:90];
      SELECT INPUT(1);
      SELECT OUTPUT(2);
      WRITETEXT('('G.T.B.KELLOCK JOB.NO.T03F')');
      NEWLINE(1);
      WRITETEXT('('VARTHERM.1.PEAK TEMPERATURE DISTRIBUTION GROSH
      AND TRARANT')');
      NEWLINE(1);
L1:M:=READ; 'IF'M=3.0'THEN'GOTO'L7'ELSE'
      E:=READ;I:=READ;EF:=READ;SP:=READ;KO:=READ;T0:=READ;L:=READ;
      Q:=0.24*I*E*EF;
      QI:=(I*E*EF*6)/(SP*100); QN:=(I*E*6)/(SP*100);
      V:=(SP*2.54)/60;
      WRITETEXT('(' T0='');PRINT(T0,3,2);SPACE(2);
      WRITETEXT('(' Q=''); PRINT(Q,5,2);
      WRITETEXT('(' V='');PRINT(V,2,4);
      WRITETEXT('(' M='');PRINT(M,2,8);
      WRITETEXT('(' KO='');PRINT(KO,1,4);
      NEWLINE(1);
      WRITETEXT('(' QN='');PRINT(QN,5,2);SPACE(2);
      WRITETEXT('(' QI=''); PRINT(QI,5,2);
      WRITETEXT('(' I=''); PRINT(I,4,2);
      WRITETEXT('(' E=''); PRINT(E,4,2);
      WRITETEXT('(' EF='');PRINT(EF,1,3);
      WRITETEXT('(' SP='');PRINT(SP,3,3);
      NEWLINE(2);
      WRITETEXT('('Y[F]');WRITETEXT('(' TEMMAX');
      WRITETEXT('(' TIMMAX');
      NEWLINE(1);
      F:=1;H:=1;
      'FOR'G:=0.5'STEP'0.01'UNTIL'1.39'DO'
      'BEGIN'Y[F]:=G;
          P:=1;
          'FOR'D:=0'STEP'-0.1'UNTIL'-30.9'DO'
          'BEGIN'TIM[P]:=D;
              R:=SQRT(Y[F]2+V2*TIM[P]2);
              S:=(1+M*T0)2;
              U:=Q*M/(2*3.14159*KO);
              W:=U/R*EXP(-L*V*(V*TIM[P]+R));
              TEMP[P]:=(SQRT(W+S)-1)/M;
              'IF'P<2'THEN'GOTO'L4'ELSE'
              'IF'TEMP[P]>TEMP[P-1]'THEN'GOTO'L4'ELSE'
              'IF'TEMP[P]<TEMP[P-1]'THEN'
              'BEGIN' TEMMAX[H]:=TEMP[P-1];
                  TIMMAX[H]:=TIM[P-1];
              'END';
              'GOTO'L5;
              L4:P:=P+1;
          L5:'END';
          F:=F+1;H:=H+1;

```

VARTHERM 1 contd.

```
      'END';  
      'FOR'PR:=1'STEP'1'UNTIL'30'DO'  
        'BEGIN'K:=0;Z:=1;  
        L6:PRINT(Y[PR+K],2,2);SPACE(2);  
        PRINT(TEMMAX[PR+K],4,2);SPACE(2);  
        PRINT(TIMMAX[PR+K],3,2);SPACE(2);  
        K:=K+30; Z:=Z+1;  
        'IF'Z<4'THEN''GOTO'L6;  
        NEWLINE(1);  
        'END';  
        PAPERTHROW;NEWLINE(1);  
        'GOTO'L1;  
    L7:NEWLINE(1);PAPERTHROW;  
  'END'
```

Programme VARTHERR 2.

06/08/69

COMPILED BY XALE MK. 3C

```
'SEND TO'(ED,PROGRAM FILE,.STORE)
'PROGRAM'(T03L)
'INPUT'1=CR0
'OUTPUT'2=LP0
'BEGIN'REAL'M,Q,KO,L,V,T0,R,U,S,W;
      'INTEGER'P,K,PR,Z,D,F;
      'REAL'ARRAY'TEMP[1:170],TIM[1:170],Y[1:5];
      SELECT INPUT(1);
      SELECT OUTPUT(2);
      WRITETEXT('('G.T.B.KELLOCK JOB.NO.T03L')');
      NEWLINE(1);
      WRITETEXT('('VARTHERR.2.THERMAL CYCLES.GROSH AND TRABANT')');
      NEWLINE(1);
L1:M:=READ;'IF'M=3.0'THEN''GOTO'L4'ELSE'
      Q:=READ;KO:=READ;L:=READ;V:=READ;T0:=READ;
      'FOR'F:=1'STEP'1'UNTIL'5'DO'Y[F]:=READ;
      F:=1;
L2:WRITETEXT('('Y[F]=')');PRINT(Y[F],2,3);SPACE(1);
      WRITETEXT('('Q=')');PRINT(Q,5,3);SPACE(1);
      WRITETEXT('('V=')');PRINT(V,2,4);SPACE(1);
      WRITETEXT('('T0=')');PRINT(T0,4,0);SPACE(1);
      WRITETEXT('('KO=')');PRINT(KO,1,4);SPACE(1);
      WRITETEXT('('L=')');PRINT(L,2,3);SPACE(1);
      NEWLINE(1);
      WRITETEXT('('TIME TEMPERATURE')');
      NEWLINE(2);
      P:=1;
      S:=(1+M*T0)2; U:=Q*M/(2*3.14159*KO);
      'FOR'D:=6'STEP'-1'UNTIL'-163'DO'
      'BEGIN'TIM[P]:=D/2;
            R:=SQRT(Y[F]2+V2*TIM[P]2);
            W:=U/R*EXP(-L*V*(V*TIM[P]+R));
            TEMP[P]:=(SQRT(W+S)-1)/M;
            P:=P+1;
      'END';
      'FOR'PR:=1'STEP'1'UNTIL'34'DO'
      'BEGIN'K:=0;Z:=1;
            L3:PRINT(TIM[PR+K],4,2);
            PRINT(TEMP[PR+K],5,2);
            K:=K+34;Z:=Z+1;
            'IF'Z<6'THEN''GOTO'L3;
            NEWLINE(1);
      'END';
      PAPERTHROW;NEWLINE(1);
      F:=F+1;'IF'F'LE'5'THEN''GOTO'L2'ELSE''GOTO'L1;
L4:NEWLINE(1);PAPERTHROW;
'END'
```

Programme VARTHERM 3.

11/09/79

COMPILED BY XAIE MA. 30

```

'SEND TO' (ED, PROGRAM FILE, .STORE)
'PROGRAM' (T031)
'LIST' 1=CRQ
'OUTPRIT' 2=LPO
'BEGIN' 'REAL' 'J,Q,V,Y,P,U,S,K,TIM,H,K,QP,N,D;
        'INTEGER' 'F,G;
        'REAL' 'APRAY' 'TL1:30],TOL1:30];
        SELECT INPUT(1);
        SELECT OUTPUT(2);
        WRITETEXT('G.T.B.KELLOCK JOB NO.T03F');
        NEWLINE(1);
        WRITETEXT('VARTHERM.3.LATENT HEAT FLOW');
        M:=READ;Y:=READ;V:=READ;QP:=READ;H:=READ;K:=READ;
        'FOR' 'F:=1' 'STEP' 1 'UNTIL' 30 'DO' 'T[F]:=READ;
        'FOR' 'F:=1' 'STEP' 1 'UNTIL' 30 'DO'
            'BEGIN' PRINT(T[F],10,0);
                    NEWLINE(1);
            'END';
        'FOR' 'G:=1' 'STEP' 1 'UNTIL' 30 'DO' 'TOLG]:=READ;
        F:=1;
        G:=1;
        NEWLINE(1);
        WRITETEXT('TIME PREHEAT TEMP. TEMP. LATENT HEAT FLOW');
        NEWLINE(1);
        WRITETEXT('Y='); PRINT(Y,2,5);
        WRITETEXT('V='); PRINT(V,2,4);
        WRITETEXT('QP='); PRINT(QP,5,2);
        WRITETEXT('H='); PRINT(H,2,2);
        WRITETEXT('K='); PRINT(K,2,2);
        NEWLINE(1);
        'FOR' 'D:=H' 'STEP' 0.5 'UNTIL' 'K' 'DO'
            'BEGIN' TIM:=D;
                    U:=(3.14159*0.042)/4;
                    R:=SQRT(V*2*TIM*2+Y*2);
                    S:=(m*2*(T[F]*2-TOLG*2));
                    W:=(T[F]-TOLG)*2*4;
                    N:=EXP(10.452*(V*2*TIM+V*R));
                    Q:=D*N*R*(S+W);
                    F:=F+1;
                    G:=G+1;
                    PRINT(TIM,2,2);PRINT(TOLG,4,2);PRINT(T[F],4,2);
                    PRINT(Q,5,2);
                    NEWLINE(1);
            'END';
        NEWLINE(1);PAPERTHROW;
'END'

```

```

'SEND TO'(ED,PROGRAM FILE,,STORE)
'PROGRAM'(T03M)
'INPUT'1=CRO
'OUTPUT'2=LPO
'BEGIN' 'REAL'M,Q,Q2,QM,KO,V,L,T0,TD,TS,G,R,R2,U,U2,W,W2,S,S2,D,K,J;
        'INTEGER'F,P,PR,Z,H;
        'REAL' 'ARRAY'TIN[1:310],TEMP[1:310],TIMMAX[1:90],
                TEMMAX[1:90],Y[1:90];
        SELECT INPUT(1);
        SELECT OUTPUT(2);
        WRITETEXT('('G.T.B.KELLOCK JOB.NO.T03M'))');
        NEWLINE(1);
        WRITETEXT('('VARTHERM.4.PEAK TEMPERATURE DISTRIBUTION LATENT
                HEAT MODIFICATION'))');
        NEWLINE(1);
L1:M:=READ;'IF'M=3.0'THEN'GOTO'L5'ELSE'
        KO:=READ;V:=READ;L:=READ;T0:=READ;Q:=READ;QM:=READ;TS:=READ;
        J:=READ;
        WRITETEXT('(' Q='));PRINT(Q,5,2);SPACE(1);
        WRITETEXT('(' V='));PRINT(V,2,4);SPACE(1);
        WRITETEXT('(' L='));PRINT(L,2,5);SPACE(1);
        WRITETEXT('(' KO='));PRINT(KO,1,5);SPACE(2);
        NEWLINE(1);
        WRITETEXT('(' TS='));PRINT(TS,2,2);SPACE(2);
        WRITETEXT('(' QM='));PRINT(QM,5,2);SPACE(2);
        WRITETEXT('(' T0='));PRINT(T0,3,1);SPACE(2);
        WRITETEXT('(' J='));PRINT(J,3,2);
        NEWLINE(2);
        WRITETEXT('('Y[F]))');WRITETEXT('(' TEMMAX'))');
        WRITETEXT('(' TIMMAX'))');
        NEWLINE(2);
        F:=1;H:=1;
        'FOR'G:=0.5'STEP'0.01'UNTIL'1.39'DO'
        'BEGIN'Y[F]:=G;
                P:=1;
                'FOR'D:=0'STEP'-0.1'UNTIL'-30.9'DO'
                'BEGIN'TIM[P]:=D;
                        R:=SQRT(Y[F]2+V2*TIM[P]2);
                        S:=(1+M*T0)2;
                        U:=Q*M/(2*3.14159*KO);
                        W:=U/R*EXP(-L*V*(V*TIM[P]+R));
                        TEMP[P]:=(SQRT(W+S)-1)/M;
                        TD:=TIM[P]+TS;
                        'IF'TD>-0.1'THEN'GOTO'L6'ELSE'
                        Q2:=QM*(1-EXP(-(TD*TD/J)));
                        U2:=Q2*M/(2*3.14159*KO);
                        S2:=(1+M*TEMP[P])2;
                        R2:=SQRT(Y[F]2+V2*TD*TD);
                        W2:=U2/R2*EXP(-L*V*(V*TD+R2));
                        TEMP[P]:=(SQRT(W2+S2)-1)/M;
                L6:'IF'P<2'THEN'GOTO'L2'ELSE'
                        'IF'TEMP[P]>TEMP[P-1]'THEN'GOTO'L2'ELSE'
                        'IF'TEMP[P]<TEMP[P-1]'THEN'
                'BEGIN' TEMMAX[H]:=TEMP[P-1];
                        TIMMAX[H]:=TIM[P-1];
                'END';
                'GOTO'L3;

```


Programme VARTHERM 4 (continued).

```

L2:P:=P+1;
  'END';
  L3:F:=F+1;H:=H+1;
  'END';
    'FOR'PR:=1'STEP'1'UNTIL'30'DO'
      'BEGIN'K:=0;Z:=1;
      L4:PRINT(Y[PR+K],2,2);SPACE(2);
          PRINT(TEMPAX[PR+K],4,2);SPACE(2);
          PRINT(TIMMAX[PR+K],3,2);SPACE(2);
          K:=K+30;Z:=Z+1;
          'IF'Z<4'THEN''GOTO'L4;
          NEWLINE(1);
        'END';
        PAPERTHROW;NEWLINE(1);
        'GOTO'L1;
    L5:NEWLINE(1);PAPERTHROW;
  'END'

```

11/09/69

COMPILED BY XATF 4X. 30

```

'END ID' (CP, PROGRAM FILE, .STORE)
'BEGIN' (103)
'INPUT' 1=10
'OUTPUT' 2=10
'BEGIN' 'REAL' M, Q1, Q2, QM, K0, L, V, T0, T1, T2, R1, R2, U1, U2, S1, S2, W1, W2,
      TD, TS, N, G, H, SLOPE, KL, J;
      'INTEGER' K, Z, D, PR, F, P;
      'REAL' 'ARRAY' TEMP[1:170], TIM[1:170], Y[1:5];
      SELECT INPUT(1);
      SELECT OUTPUT(2);
      WRITE TEXT('G.T.E. KEILLOCK JOB.NO.103K');
      WRITE TEXT('GRUSH, TRABANT, VARTHERR 4, LATENT HEAT MOD2. ');
      NEWLINE(1);
      L1: G:=READ; IF M=3.0 THEN GOTO L4 ELSE
      Q1:=READ; K0:=READ; L:=READ; V:=READ; T0:=READ;
      QM:=READ; G:=READ;
      FOR F:=1 STEP 1 UNTIL 5 DO Y[F]:=READ;
      F:=1;
      L2: FOR TS:=3.0 STEP 0.5 UNTIL 3.5 DO
      'BEGIN' WRITE TEXT('TIME TEMPERATURE');
      WRITE TEXT('Y[F]= '); PRINT(Y[F], 2, 3);
      WRITE TEXT('Q1= '); PRINT(Q1, 5, 3);
      WRITE TEXT('T0= '); PRINT(T0, 4, 0);
      WRITE TEXT('QM= '); PRINT(QM, 4, 2);
      WRITE TEXT('TS= '); PRINT(TS, 2, 1);
      WRITE TEXT('G= '); PRINT(G, 2, 2);
      NEWLINE(2);
      P:=1;
      FOR D:=6 STEP -1 UNTIL -163 DO
      'BEGIN' TIM[P]:=D/2;
      S1:=(1+N*T0)2;
      U1:=Q1*M/(2*3.14159*K0);
      R1:=SQRT(Y[F]*Y[F]+(V*V)*TIM[P]*TIM[P]);
      W1:=U1/R1*EXP(-L*M*(V*TIM[P]+R1));
      T1:=(SQRT(W1+S1)-1)/M;
      TEMP[P]:=T1;
      IF TIM[P] GE -TS THEN GOTO L8;
      TD:=TIM[P]+TS;
      Q2:=QM*(1-EXP(-(TD+TD/G)));
      U2:=Q2*M/(2*3.14159*K0);
      S2:=(1+N*TEMP[P])2;
      R2:=SQRT(Y[F]*Y[F]+(V*V)*TD*TD);
      W2:=U2/R2*EXP(-L*M*(V*TD+R2));
      T2:=(SQRT(W2+S2)-1)/M;
      TEMP[P]:=T2;
      L8: P:=P+1;
      L6: 'END';
      FOR PR:=1 STEP 1 UNTIL 34 DO
      'BEGIN' K:=0; Z:=1;
      L3: PRINT(TIM[PR+K], 4, 2); PRINT(TEMP[PR+K], 5, 2);
      K:=K+34; Z:=Z+1;
      IF Z<6 THEN GOTO L3;
      NEWLINE(1);
      'END';
      PAPERTHROW;
      'END';
      F:=F+1; IF F<6 THEN GOTO L2 ELSE GOTO L1;
      L4: NEWLINE(1); PAPERTHROW;
      'END'

```

Programme VARTHERR 5.

11/09/69

COMPILED BY XALE MK. 3C

```

'SEND TO'(CD,PROGRAM FILE,.STORE)
'PROGRAM'(103N)
'INPUT'1=CR0
'OUTPUT'2=LPU
'BEGIN' 'REAL'M,Q1,Q2,QM,K0,L,V,T0,TS,TD,R1,R2,U1,U2,S1,S2,W1,W2,
      N,G,H,SLOPE,KL,J,BS,MF,KO2,A,B;
      'INTEGER'K,Z,D,PR,F,P;
      'REAL' 'ARRAY' TEMP[1:170],TIM[1:170],Y[1:5];
      SELECT INPUT(1);
      SELECT OUTPUT(2);
      WRTTEXT('('G.T.B.KELLOCK JOB.NO.103N')');
      NEWLINE(1);
      WRTTEXT('('VARTHERR.5. THERMAL CYCLES LATENT HEAT AND
      THERMAL CONDUCTIVITY MODIFICATION')');
      NEWLINE(1);
L1:M:=READ: 'IF'M=3.0'THEN''GOTO'L4'ELSE'
      Q1:=READ;K0:=READ;L:=READ;V:=READ;T0:=READ;TS:=READ;
      QM:=READ;N:=READ;G:=READ;BS:=READ;MF:=READ;KO2:=READ;
      'FOR'F:=1'STEP'1'UNTIL'5'DO'Y[F]:=READ;
      F:=1;
      A:=(KO2-K0)/BS↑2;
      B:=K0*M-2*A*BS;
L2:WRTTEXT('('Y[F]=')');PRINT(Y[F],2,3);SPACE(1);
      WRTTEXT('('Q1=')');PRINT(Q1,5,3);SPACE(1);
      WRTTEXT('('T0=')');PRINT(T0,4,0);SPACE(1);
      WRTTEXT('('QM=')');PRINT(QM,4,2);SPACE(1);
      WRTTEXT('('TS=')');PRINT(TS,2,1);SPACE(1);
      WRTTEXT('('N=')');PRINT(N,4,2);SPACE(1);
      WRTTEXT('('G=')');PRINT(G,2,2);SPACE(1);
      NEWLINE(1);
      WRTTEXT('('K0=')');PRINT(K0,1,5);SPACE(1);
      WRTTEXT('('KO2=')');PRINT(KO2,1,5);SPACE(2);
      WRTTEXT('('H=')');PRINT(M,1,7);SPACE(1);
      WRTTEXT('('BS=')');PRINT(BS,3,0);SPACE(1);
      WRTTEXT('('A=')');PRINT(A,1,10);SPACE(1);
      WRTTEXT('('B=')');PRINT(B,1,10);SPACE(1);
      WRTTEXT('('MF=')');PRINT(MF,3,0);
      NEWLINE(2);
      P:=1;
      'FOR'D:=6'STEP'-1'UNTIL'-163'DO'
'BEGIN' TIM[P]:=0/2;
      S1:=(1+M*T0)↑2;
      U1:=Q1*M/(2*3.14159*K0);
      R1:=SQRT(Y[F]↑2+V↑2*TIM[P]*TIM[P]);
      W1:=U1/R1*EXP(-L*V*(V*TIM[P]+R1));
      TEMP[P]:=(SQRT(W1+S1)-1)/M;
      TD:=TIM[P]+TS;
      'IF'TD>-0.1'THEN''GOTO'L7'ELSE'
      H:=-TD;
      Q2:=QM*(1-EXP(-(H↑N/G)));
      U2:=Q2*M/(2*3.14159*K0);
      S2:=(1+M*TEMP[P])↑2;
      R2:=SQRT(Y[F]↑2+V↑2*H↑2);
      W2:=U2/R2*EXP(-L*V*(V*TD+R2));
      TEMP[P]:=(SQRT(W2+S2)-1)/M;
      L7:'IF'P<2'THEN''GOTO'L5'ELSE'
      'IF'TEMP[P]'GE'TEMP[P-1]'THEN'

```

Programme VARTHERM 5 (continued).

```

      'GOTO' L5 'ELSE'
      'IF' TEMP[P] > BS 'THEN' 'GOTO' L5 'ELSE'
      'BEGIN' KL := A * TEMP[P] + 2 + B * TEMP[P] + K02;
      SLOPE := 2 * A * TEMP[P] + B;
      K0 := KL - SLOPE * TEMP[P];
      M := SLOPE / K0;
      L := ((TEMP[P] - MF) / (BS - MF)) * 1.066 + 9.416;
      'END';
      L5: P := P + 1;
    'END';

    'FOR' PR := 1 'STEP' 1 'UNTIL' 34 'DO'
    'BEGIN' K := 0; Z := 1;
      L3: PRINT(TIME[PR + K], 4, 2);
      PRINT(TEMP[PR + K], 5, 2);
      K := K + 34; Z := Z + 1;
      'IF' Z < 6 'THEN' 'GOTO' L3;
      NEWLINE(1);
    'END';
    PAPERTHROW; NEWLINE(1);
    M := 0.000514; L := 10.482; K0 := 0.042;
    F := F + 1; 'IF' F < 6 'THEN' 'GOTO' L2 'ELSE' 'GOTO' L1;
    L4: PAPERTHROW; NEWLINE(1);
  'END'

```

G.T.B.KELLOCKJOB.NO.T03F
VARTHERM.1.PEAKTEMPERATUREDISTRIBUTIONGROSHANDTRABANT
T0= 120.00 Q= 3528.00 V= 0.7056 M= 0.00051400 KO= 0.0420
QN= 54.00 QI= 52.92 I= 500.00 E= 30.00 EF= 0.980 SP= 16.667

[illegible]

Programme VAPTHERM 1: print-out of peak temperature distribution.

G.I.B. REFLECTOR, INC. 1031
 VARTHERM. 2: THERMAL CYCLES. PROS AND 1774 BANT
 YUFJ= 0.037 0= 3528.000 V= 0.7056
 TIME TEMPERATURE

TO= 120 KO= 0.0420 L= 10.482

3.00	120.00	-14.00	608.69	-51.00	373.14	-48.00	290.75	-65.00	248.83
2.50	126.00	-14.50	595.71	-51.50	369.60	-48.50	289.13	-65.50	247.91
2.00	120.00	-15.00	583.40	-52.00	366.15	-49.00	287.55	-66.00	247.00
1.50	120.00	-15.50	571.70	-52.50	362.80	-49.50	285.99	-66.50	246.10
1.00	120.00	-16.00	560.57	-53.00	359.54	-50.00	284.46	-67.00	245.21
0.50	122.00	-16.50	549.98	-53.50	356.37	-50.50	282.95	-67.50	244.34
0.00	237.00	-17.00	539.88	-54.00	353.28	-51.00	281.48	-68.00	243.48
-0.50	602.00	-17.50	530.24	-54.50	350.26	-51.50	280.03	-68.50	242.63
-1.00	1009.99	-18.00	521.04	-55.00	347.33	-52.00	278.61	-69.00	241.79
-1.50	1209.38	-18.50	512.23	-55.50	344.47	-52.50	277.21	-69.50	240.97
-2.00	1273.16	-19.00	503.80	-56.00	341.68	-53.00	275.84	-70.00	240.15
-2.50	1271.06	-19.50	495.73	-56.50	338.95	-53.50	274.49	-70.50	239.35
-3.00	1259.95	-20.00	487.99	-57.00	336.30	-54.00	273.16	-71.00	238.56
-3.50	1197.00	-20.50	480.55	-57.50	333.71	-54.50	271.86	-71.50	237.77
-4.00	1150.25	-21.00	473.42	-58.00	331.18	-55.00	270.58	-72.00	237.00
-4.50	1105.50	-21.50	466.56	-58.50	328.71	-55.50	269.32	-72.50	236.24
-5.00	1058.50	-22.00	459.95	-59.00	326.30	-56.00	268.08	-73.00	235.49
-5.50	1016.02	-22.50	453.60	-59.50	323.94	-56.50	266.86	-73.50	234.74
-6.00	976.50	-23.00	447.48	-40.00	321.63	-57.00	265.66	-74.00	234.01
-6.50	939.54	-23.50	441.58	-40.50	319.38	-57.50	264.48	-74.50	233.29
-7.00	905.02	-24.00	435.89	-41.00	317.18	-58.00	263.32	-75.00	232.57
-7.50	873.17	-24.50	430.39	-41.50	315.02	-58.50	262.18	-75.50	231.87
-8.00	843.50	-25.00	425.08	-42.00	312.91	-59.00	261.05	-76.00	231.17
-8.50	816.11	-25.50	419.95	-42.50	310.85	-59.50	259.95	-76.50	230.48
-9.00	790.02	-26.00	415.00	-43.00	308.83	-60.00	258.86	-77.00	229.80
-9.50	766.67	-26.50	410.20	-43.50	306.85	-60.50	257.78	-77.50	229.13
-10.00	744.59	-27.00	405.55	-44.00	304.91	-61.00	256.73	-78.00	228.47
-10.50	723.54	-27.50	401.05	-44.50	303.02	-61.50	255.69	-78.50	227.81
-11.00	704.01	-28.00	396.69	-45.00	301.16	-62.00	254.66	-79.00	227.16
-11.50	685.57	-28.50	392.47	-45.50	299.34	-62.50	253.66	-79.50	226.52
-12.00	668.42	-29.00	388.37	-46.00	297.55	-63.00	252.66	-80.00	225.89
-12.50	652.18	-29.50	384.59	-46.50	295.80	-63.50	251.68	-80.50	225.27
-13.00	636.56	-30.00	380.53	-47.00	294.09	-64.00	250.72	-81.00	224.65
-13.50	622.58	-30.50	376.73	-47.50	292.40	-64.50	249.77	-81.50	224.04

Programme VARTHERM 2: print-out for 1275°C peak temperature.

Y[F]= 0.792 Q= 3528.000 V= 0.7056 T0= 120 K0= 0.0420 L= 10.482
 TIME TEMPERATURE

3.00	120.00	-14.00	572.90	-31.00	364.28	-48.00	286.81	-65.00	246.61
2.50	120.00	-14.50	561.97	-31.50	360.99	-48.50	285.27	-65.50	245.72
2.00	120.00	-15.00	551.53	-32.00	357.79	-49.00	283.75	-66.00	244.84
1.50	120.00	-15.50	541.56	-32.50	354.67	-49.50	282.27	-66.50	243.97
1.00	120.01	-16.00	532.02	-33.00	351.64	-50.00	280.81	-67.00	243.12
0.50	120.88	-16.50	522.89	-33.50	348.68	-50.50	279.37	-67.50	242.27
0.00	142.59	-17.00	514.14	-34.00	345.79	-51.00	277.96	-68.00	241.44
-0.50	276.03	-17.50	505.76	-34.50	342.98	-51.50	276.58	-68.50	240.62
-1.00	512.11	-18.00	497.71	-35.00	340.24	-52.00	275.22	-69.00	239.81
-1.50	713.13	-18.50	489.99	-35.50	337.56	-52.50	273.88	-69.50	239.01
-2.00	836.89	-19.00	482.57	-36.00	334.94	-53.00	272.57	-70.00	238.23
-2.50	900.93	-19.50	475.44	-36.50	332.39	-53.50	271.27	-70.50	237.45
-3.00	927.40	-20.00	468.57	-37.00	329.90	-54.00	270.00	-71.00	236.66
-3.50	931.61	-20.50	461.96	-37.50	327.47	-54.50	268.75	-71.50	235.92
-4.00	922.95	-21.00	455.60	-38.00	325.09	-55.00	267.53	-72.00	235.18
-4.50	907.00	-21.50	449.46	-38.50	322.76	-55.50	266.32	-72.50	234.44
-5.00	887.11	-22.00	443.54	-39.00	320.49	-56.00	265.13	-73.00	233.71
-5.50	865.28	-22.50	437.83	-39.50	318.27	-56.50	263.96	-73.50	232.99
-6.00	842.72	-23.00	432.31	-40.00	316.09	-57.00	262.81	-74.00	232.28
-6.50	820.15	-23.50	426.98	-40.50	313.96	-57.50	261.68	-74.50	231.58
-7.00	798.01	-24.00	421.82	-41.00	311.88	-58.00	260.56	-75.00	230.88
-7.50	776.56	-24.50	416.84	-41.50	309.85	-58.50	259.46	-75.50	230.20
-8.00	755.94	-25.00	412.01	-42.00	307.85	-59.00	258.38	-76.00	229.52
-8.50	736.20	-25.50	407.34	-42.50	305.90	-59.50	257.32	-76.50	228.86
-9.00	717.37	-26.00	402.81	-43.00	303.98	-60.00	256.27	-77.00	228.20
-9.50	699.44	-26.50	398.42	-43.50	302.11	-60.50	255.24	-77.50	227.54
-10.00	682.39	-27.00	394.17	-44.00	300.27	-61.00	254.22	-78.00	226.90
-10.50	666.19	-27.50	390.04	-44.50	298.47	-61.50	253.22	-78.50	226.26
-11.00	650.79	-28.00	386.03	-45.00	296.71	-62.00	252.23	-79.00	225.64
-11.50	636.15	-28.50	382.14	-45.50	294.98	-62.50	251.26	-79.50	225.01
-12.00	622.22	-29.00	378.37	-46.00	293.28	-63.00	250.30	-80.00	224.40
-12.50	608.98	-29.50	374.69	-46.50	291.62	-63.50	249.36	-80.50	223.79
-13.00	596.37	-30.00	371.12	-47.00	289.98	-64.00	248.43	-81.00	223.19
-13.50	584.35	-30.50	367.65	-47.50	288.38	-64.50	247.51	-81.50	222.60

Programme VARTHERM 2: print-out for 930°C peak temperature.

VLF1= 0.908 O= 3528.000 V= 0.7056 T0= 120 KO= 0.0420 L= 10.482
 TIME/TEMPERATURE

3.00	120.00	-14.00	543.11	-31.00	356.64	-48.00	283.38	-65.00	244.66
2.50	120.00	-14.50	533.83	-31.50	353.57	-48.50	281.90	-65.50	243.80
2.00	120.00	-15.00	524.90	-32.00	350.57	-49.00	280.45	-66.00	242.95
1.50	120.00	-15.50	516.32	-32.50	347.66	-49.50	279.02	-66.50	242.11
1.00	120.01	-16.00	508.07	-33.00	344.81	-50.00	277.62	-67.00	241.28
0.50	120.35	-16.50	500.13	-33.50	342.03	-50.50	276.25	-67.50	240.46
0.00	128.39	-17.00	492.48	-34.00	339.32	-51.00	274.89	-68.00	239.66
-0.50	184.27	-17.50	485.12	-34.50	336.68	-51.50	273.56	-68.50	238.86
-1.00	315.48	-18.00	478.03	-35.00	334.10	-52.00	272.26	-69.00	238.08
-1.50	466.75	-18.50	471.19	-35.50	331.58	-52.50	270.97	-69.50	237.30
-2.00	588.17	-19.00	464.60	-36.00	329.11	-53.00	269.71	-70.00	236.54
-2.50	670.36	-19.50	458.24	-36.50	326.71	-53.50	268.47	-70.50	235.78
-3.00	720.66	-20.00	452.11	-37.00	324.35	-54.00	267.25	-71.00	235.04
-3.50	748.41	-20.50	446.18	-37.50	322.05	-54.50	266.04	-71.50	234.30
-4.00	760.96	-21.00	440.45	-38.00	319.80	-55.00	264.86	-72.00	233.57
-4.50	763.49	-21.50	434.91	-38.50	317.60	-55.50	263.70	-72.50	232.86
-5.00	759.46	-22.00	429.56	-39.00	315.45	-56.00	262.55	-73.00	232.15
-5.50	751.21	-22.50	424.38	-39.50	313.34	-56.50	261.43	-73.50	231.45
-6.00	740.29	-23.00	419.36	-40.00	311.28	-57.00	260.32	-74.00	230.76
-6.50	727.74	-23.50	414.51	-40.50	309.26	-57.50	259.22	-74.50	230.07
-7.00	714.26	-24.00	409.80	-41.00	307.29	-58.00	258.15	-75.00	229.40
-7.50	700.32	-24.50	405.24	-41.50	305.35	-58.50	257.09	-75.50	228.74
-8.00	686.26	-25.00	400.82	-42.00	303.45	-59.00	256.05	-76.00	228.08
-8.50	672.28	-25.50	396.53	-42.50	301.59	-59.50	255.02	-76.50	227.43
-9.00	658.53	-26.00	392.36	-43.00	299.77	-60.00	254.01	-77.00	226.79
-9.50	645.10	-26.50	388.32	-43.50	297.99	-60.50	253.01	-77.50	226.15
-10.00	632.05	-27.00	384.40	-44.00	296.24	-61.00	252.03	-78.00	225.52
-10.50	619.42	-27.50	380.58	-44.50	294.52	-61.50	251.06	-78.50	224.90
-11.00	607.23	-28.00	376.87	-45.00	292.84	-62.00	250.11	-79.00	224.29
-11.50	595.48	-28.50	373.27	-45.50	291.18	-62.50	249.17	-79.50	223.69
-12.00	584.16	-29.00	369.76	-46.00	289.56	-63.00	248.24	-80.00	223.09
-12.50	573.28	-29.50	366.35	-46.50	287.97	-63.50	247.33	-80.50	222.50
-13.00	562.82	-30.00	363.02	-47.00	286.41	-64.00	246.43	-81.00	221.91
-13.50	552.77	-30.50	359.79	-47.50	284.88	-64.50	245.54	-81.50	221.33

Programme VARTHERM 2: print-out for 765°C peak temperature.

YLF)= 1.075 U= 3528.000 V= 0.7056 T0= 120 K0= 0.0420 L= 10.482
TIME REPORT

3.00	126.00	-14.00	514.00	-51.00	348.91	-48.00	279.06	-65.00	242.66
2.50	120.00	-14.50	508.26	-51.50	346.05	-48.50	278.45	-65.50	241.83
2.00	120.00	-15.00	498.76	-52.00	343.27	-49.00	277.07	-66.00	241.01
1.50	120.00	-15.50	491.50	-52.50	340.55	-49.50	275.70	-66.50	240.19
1.00	120.00	-16.00	484.47	-53.00	337.89	-50.00	274.36	-67.00	239.39
0.50	120.15	-16.50	477.67	-53.50	335.30	-50.50	273.05	-67.50	238.60
0.00	123.50	-17.00	471.08	-54.00	332.77	-51.00	271.75	-68.00	237.82
-0.50	167.70	-17.50	464.70	-54.50	330.29	-51.50	270.48	-68.50	237.05
-1.00	218.73	-18.00	458.52	-55.00	327.67	-52.00	269.23	-69.00	236.29
-1.50	520.27	-18.50	452.55	-55.50	325.50	-52.50	268.00	-69.50	235.54
-2.00	420.45	-19.00	446.75	-56.00	323.19	-53.00	266.79	-70.00	234.80
-2.50	501.17	-19.50	441.14	-56.50	320.93	-53.50	265.59	-70.50	234.06
-3.00	559.02	-20.00	435.74	-57.00	318.72	-54.00	264.42	-71.00	233.34
-3.50	595.72	-20.50	430.44	-57.50	316.55	-54.50	263.27	-71.50	232.63
-4.00	625.77	-21.00	425.54	-58.00	314.43	-55.00	262.13	-72.00	231.92
-4.50	640.28	-21.50	420.59	-58.50	312.35	-55.50	261.01	-72.50	231.23
-5.00	647.72	-22.00	415.58	-59.00	310.32	-56.00	259.91	-73.00	230.54
-5.50	649.77	-22.50	410.92	-59.50	308.33	-56.50	258.83	-73.50	229.86
-6.00	648.10	-23.00	406.40	-60.00	306.38	-57.00	257.76	-74.00	229.19
-6.50	645.54	-23.50	402.01	-60.50	304.47	-57.50	256.71	-74.50	228.53
-7.00	637.23	-24.00	397.75	-61.00	302.60	-58.00	255.67	-75.00	227.87
-7.50	629.05	-24.50	393.60	-61.50	300.77	-58.50	254.65	-75.50	227.23
-8.00	621.19	-25.00	389.58	-62.00	298.97	-59.00	253.65	-76.00	226.59
-8.50	612.27	-25.50	385.67	-62.50	297.21	-59.50	252.66	-76.50	225.96
-9.00	602.92	-26.00	381.86	-63.00	295.48	-60.00	251.69	-77.00	225.33
-9.50	593.49	-26.50	378.16	-63.50	293.78	-60.50	250.72	-77.50	224.72
-10.00	584.15	-27.00	374.55	-64.00	292.12	-61.00	249.78	-78.00	224.11
-10.50	574.63	-27.50	371.05	-64.50	290.49	-61.50	248.84	-78.50	223.50
-11.00	565.30	-28.00	367.63	-65.00	288.88	-62.00	247.92	-79.00	222.91
-11.50	556.25	-28.50	364.31	-65.50	287.31	-62.50	247.02	-79.50	222.32
-12.00	547.34	-29.00	361.07	-66.00	285.77	-63.00	246.12	-80.00	221.74
-12.50	538.55	-29.50	357.91	-66.50	284.25	-63.50	245.24	-80.50	221.16
-13.00	530.19	-30.00	354.84	-67.00	282.76	-64.00	244.37	-81.00	220.59
-13.50	521.97	-30.50	351.84	-67.50	281.30	-64.50	243.51	-81.50	220.03

Programme VARTHERM 2: print-out for 650°C peak temperature.

Q= 3528.00 V= 0.7056 L= 10.48200 K0= 0.04200
 TS= 3.50 QM= 650.00 T0= 120.0 J= 10.00

Y[F]TEMPMAXTIMMAX

0.50	1794.18	-1.40	0.80	918.44	-3.40	1.10	632.67	-8.40
0.51	1745.70	-1.40	0.81	901.97	-3.50	1.11	625.69	-8.40
0.52	1700.02	-1.50	0.82	885.99	-3.60	1.12	618.81	-8.50
0.53	1655.27	-1.60	0.83	870.47	-3.70	1.13	612.02	-8.60
0.54	1613.36	-1.60	0.84	855.43	-3.80	1.14	605.33	-8.70
0.55	1572.38	-1.70	0.85	840.87	-3.90	1.15	598.72	-8.80
0.56	1533.37	-1.70	0.86	826.80	-4.00	1.16	592.22	-8.80
0.57	1495.78	-1.80	0.87	813.31	-4.20	1.17	585.81	-8.90
0.58	1459.36	-1.80	0.88	812.35	-6.70	1.18	579.49	-9.00
0.59	1424.80	-1.90	0.89	802.99	-6.80	1.19	573.27	-9.10
0.60	1391.16	-2.00	0.90	793.76	-6.90	1.20	567.13	-9.20
0.61	1358.88	-2.00	0.91	784.66	-6.90	1.21	561.09	-9.30
0.62	1327.88	-2.10	0.92	775.69	-7.00	1.22	555.13	-9.40
0.63	1297.76	-2.20	0.93	766.83	-7.10	1.23	549.27	-9.50
0.64	1268.90	-2.20	0.94	758.08	-7.20	1.24	543.50	-9.50
0.65	1241.08	-2.30	0.95	749.44	-7.20	1.25	537.82	-9.60
0.66	1214.07	-2.40	0.96	740.94	-7.30	1.26	532.22	-9.70
0.67	1188.06	-2.40	0.97	732.53	-7.40	1.27	526.72	-9.80
0.68	1163.04	-2.50	0.98	724.22	-7.50	1.28	521.30	-9.90
0.69	1138.75	-2.60	0.99	716.02	-7.50	1.29	515.97	-10.00
0.70	1115.20	-2.70	1.00	707.94	-7.60	1.30	510.72	-10.20
0.71	1092.61	-2.70	1.01	699.96	-7.70	1.31	505.56	-10.30
0.72	1070.71	-2.80	1.02	692.07	-7.80	1.32	500.49	-10.40
0.73	1049.47	-2.90	1.03	684.30	-7.80	1.33	495.50	-10.50
0.74	1028.86	-3.00	1.04	676.65	-7.90	1.34	490.59	-10.60
0.75	1009.01	-3.00	1.05	669.05	-8.00	1.35	485.76	-10.70
0.76	989.78	-3.10	1.06	661.57	-8.10	1.36	481.01	-10.80
0.77	971.12	-3.20	1.07	654.20	-8.10	1.37	476.35	-11.00
0.78	953.01	-3.30	1.08	646.93	-8.20	1.38	471.77	-11.10
0.79	935.44	-3.40	1.09	639.76	-8.30	1.39	467.26	-11.20

Programme VARTHERM 4: print-out of peak temperature distribution.

TIMETEMPERATUREDIFF= 0.657 Q1= 3528.000 T0= 120 QM= 650.00 TS= 3.5 G= 10.00									
3.00	120.00	-14.00	709.74	-31.00	421.89	-48.00	323.04	-65.00	273.00
2.50	120.00	-14.50	693.64	-31.50	417.63	-48.50	321.11	-65.50	271.90
2.00	120.00	-15.00	678.39	-32.00	413.48	-49.00	319.21	-66.00	270.82
1.50	120.00	-15.50	663.93	-32.50	409.45	-49.50	317.35	-66.50	269.75
1.00	120.00	-16.00	650.19	-33.00	405.53	-50.00	315.52	-67.00	268.69
0.50	122.91	-16.50	637.14	-33.50	401.71	-50.50	313.72	-67.50	267.65
0.00	207.05	-17.00	624.71	-34.00	397.99	-51.00	311.96	-68.00	266.63
-0.50	602.14	-17.50	612.87	-34.50	394.37	-51.50	310.23	-68.50	265.62
-1.00	1009.90	-18.00	601.57	-35.00	390.85	-52.00	308.53	-69.00	264.62
-1.50	1209.08	-18.50	590.78	-35.50	387.41	-52.50	306.86	-69.50	263.63
-2.00	1273.16	-19.00	580.46	-36.00	384.06	-53.00	305.22	-70.00	262.66
-2.50	1271.06	-19.50	570.58	-36.50	380.79	-53.50	303.61	-70.50	261.71
-3.00	1239.95	-20.00	561.13	-37.00	377.61	-54.00	302.03	-71.00	260.76
-3.50	1197.00	-20.50	552.06	-37.50	374.50	-54.50	300.47	-71.50	259.83
-4.00	1151.88	-21.00	543.36	-38.00	371.46	-55.00	298.94	-72.00	258.91
-4.50	1116.52	-21.50	535.00	-38.50	368.50	-55.50	297.44	-72.50	258.00
-5.00	1093.46	-22.00	526.97	-39.00	365.60	-56.00	295.96	-73.00	257.11
-5.50	1077.67	-22.50	519.24	-39.50	362.77	-56.50	294.50	-73.50	256.22
-6.00	1064.17	-23.00	511.80	-40.00	360.01	-57.00	293.07	-74.00	255.35
-6.50	1046.71	-23.50	504.64	-40.50	357.31	-57.50	291.66	-74.50	254.49
-7.00	1030.29	-24.00	497.73	-41.00	354.67	-58.00	290.28	-75.00	253.64
-7.50	1008.62	-24.50	491.07	-41.50	352.09	-58.50	288.92	-75.50	252.79
-8.00	984.26	-25.00	484.64	-42.00	349.56	-59.00	287.58	-76.00	251.97
-8.50	958.14	-25.50	478.42	-42.50	347.09	-59.50	286.26	-76.50	251.15
-9.00	931.20	-26.00	472.42	-43.00	344.67	-60.00	284.96	-77.00	250.34
-9.50	904.27	-26.50	466.62	-43.50	342.30	-60.50	283.68	-77.50	249.54
-10.00	877.25	-27.00	461.00	-44.00	339.98	-61.00	282.42	-78.00	248.75
-10.50	852.63	-27.50	455.57	-44.50	337.71	-61.50	281.18	-78.50	247.97
-11.00	828.52	-28.00	450.30	-45.00	335.49	-62.00	279.96	-79.00	247.20
-11.50	805.71	-28.50	445.20	-45.50	333.31	-62.50	278.75	-79.50	246.43
-12.00	784.20	-29.00	440.25	-46.00	331.17	-63.00	277.57	-80.00	245.68
-12.50	763.02	-29.50	435.46	-46.50	329.08	-63.50	276.40	-80.50	244.94
-13.00	744.06	-30.00	430.80	-47.00	327.03	-64.00	275.25	-81.00	244.20
-13.50	726.77	-30.50	426.28	-47.50	325.02	-64.50	274.12	-81.50	243.46

Programme VARTHERM 4(2): print-out for 1275°C peak temperature.

TIME	TEMP	RAT	U	V	F	T	Q	M	T	S	G
3.00	120.00	0.792	Q1=	5528.000	T0=	120	QM=	650.00	TS=	3.5	G= 10.00
2.50	120.00	-14.50		664.67	-31.00		411.20		-48.00	516.33	-65.00
2.00	120.00	-14.50		651.26	-31.50		407.74		-48.50	516.48	-65.50
1.50	120.00	-15.00		638.51	-32.00		405.40		-49.00	514.67	-66.00
1.00	120.00	-15.50		626.50	-32.50		399.65		-49.50	512.90	-66.50
0.50	120.00	-16.00		614.64	-33.00		396.00		-50.00	511.15	-67.00
0.00	120.00	-16.50		605.48	-33.50		392.45		-50.50	509.44	-67.50
-0.50	142.50	-17.00		592.80	-34.00		388.98		-51.00	507.76	-68.00
-1.00	276.03	-17.50		582.57	-34.50		385.60		-51.50	506.10	-68.50
-1.50	512.11	-18.00		572.77	-35.00		382.31		-52.00	504.48	-69.00
-2.00	713.13	-18.50		565.36	-35.50		379.10		-52.50	502.88	-69.50
-2.50	836.39	-19.00		554.32	-36.00		375.96		-53.00	501.31	-70.00
-3.00	900.98	-19.50		545.64	-36.50		372.90		-53.50	499.77	-70.50
-3.50	927.40	-20.00		537.30	-37.00		369.91		-54.00	498.25	-71.00
-4.00	931.61	-20.50		529.27	-37.50		366.99		-54.50	496.76	-71.50
-4.50	923.48	-21.00		521.54	-38.00		364.14		-55.00	495.30	-72.00
-5.00	912.45	-21.50		514.09	-38.50		361.35		-55.50	493.86	-72.50
-5.50	905.42	-22.00		506.91	-39.00		358.63		-56.00	492.44	-73.00
-6.00	902.56	-22.50		499.99	-39.50		355.97		-56.50	491.04	-73.50
-6.50	901.01	-23.00		493.31	-40.00		353.36		-57.00	489.67	-74.00
-7.00	897.37	-23.50		486.85	-40.50		350.81		-57.50	488.32	-74.50
-7.50	891.55	-24.00		480.61	-41.00		348.32		-58.00	486.99	-75.00
-8.00	881.54	-24.50		474.58	-41.50		345.88		-58.50	485.68	-75.50
-8.50	858.17	-25.00		468.75	-42.00		343.49		-59.00	484.39	-76.00
-9.00	852.17	-25.50		463.10	-42.50		341.15		-59.50	483.12	-76.50
-9.50	854.41	-26.00		457.64	-43.00		338.86		-60.00	481.87	-77.00
-10.00	815.72	-26.50		452.34	-43.50		336.62		-60.50	480.64	-77.50
-10.50	796.75	-27.00		447.21	-44.00		334.42		-61.00	479.43	-78.00
-11.00	777.95	-27.50		442.23	-44.50		332.27		-61.50	478.23	-78.50
-11.50	759.53	-28.00		437.40	-45.00		330.16		-62.00	477.06	-79.00
-12.00	741.97	-28.50		432.71	-45.50		328.09		-62.50	475.90	-79.50
-12.50	725.06	-29.00		428.16	-46.00		326.06		-63.00	474.76	-80.00
-13.00	708.86	-29.50		423.74	-46.50		324.07		-63.50	473.63	-80.50
-13.50	693.42	-30.00		419.44	-47.00		322.12		-64.00	472.52	-81.00
-14.00	678.70	-30.50		415.26	-47.50		320.21		-64.50	471.43	-81.50

Programme VARTHERM 4(2): print-out for 930°C peak temperature.

TIMETEMPERATURE	0.931	G1=	3528.000	T0=	120	QM=	650.00	TS=	3.5	G=	10.00	
3.00	120.00	-14.00	619.59	-31.00		400.04		-48.00		513.35		-65.00
2.50	120.00	-14.50	608.81	-31.50		396.41		-48.50		511.60		-55.50
2.00	120.00	-15.00	598.41	-32.00		392.87		-49.00		509.88		-66.00
1.50	120.00	-15.50	588.38	-32.50		389.41		-49.50		508.20		-66.50
1.00	120.00	-16.00	578.72	-33.00		386.05		-50.00		506.54		-67.00
0.50	120.00	-16.50	569.42	-33.50		382.76		-50.50		504.91		-67.50
0.00	120.00	-17.00	560.44	-34.00		379.56		-51.00		503.31		-68.00
-0.50	120.00	-17.50	551.79	-34.50		376.43		-51.50		501.74		-68.50
-1.00	120.00	-18.00	543.45	-35.00		373.37		-52.00		500.19		-69.00
-1.50	120.00	-18.50	535.40	-35.50		370.39		-52.50		498.67		-69.50
-2.00	120.00	-19.00	527.64	-36.00		367.47		-53.00		497.18		-70.00
-2.50	120.00	-19.50	520.14	-36.50		364.62		-53.50		495.71		-70.50
-3.00	120.00	-20.00	512.90	-37.00		361.84		-54.00		494.26		-71.00
-3.50	120.00	-20.50	505.91	-37.50		359.12		-54.50		492.84		-71.50
-4.00	120.00	-21.00	499.15	-38.00		356.45		-55.00		491.44		-72.00
-4.50	120.00	-21.50	492.61	-38.50		353.85		-55.50		490.06		-72.50
-5.00	120.00	-22.00	486.28	-39.00		351.30		-56.00		488.71		-73.00
-5.50	120.00	-22.50	480.16	-39.50		348.81		-56.50		487.37		-73.50
-6.00	120.00	-23.00	474.24	-40.00		346.37		-57.00		486.06		-74.00
-6.50	120.00	-23.50	468.50	-40.50		343.98		-57.50		484.77		-74.50
-7.00	120.00	-24.00	462.93	-41.00		341.64		-58.00		483.50		-75.00
-7.50	120.00	-24.50	457.54	-41.50		339.35		-58.50		482.24		-75.50
-8.00	120.00	-25.00	452.31	-42.00		337.10		-59.00		481.01		-76.00
-8.50	120.00	-25.50	447.24	-42.50		334.91		-59.50		479.79		-76.50
-9.00	120.00	-26.00	442.31	-43.00		332.75		-60.00		478.60		-77.00
-9.50	120.00	-26.50	437.53	-43.50		330.64		-60.50		477.42		-77.50
-10.00	120.00	-27.00	432.89	-44.00		328.57		-61.00		476.26		-78.00
-10.50	120.00	-27.50	428.37	-44.50		326.53		-61.50		475.11		-78.50
-11.00	120.00	-28.00	423.98	-45.00		324.54		-62.00		473.98		-79.00
-11.50	120.00	-28.50	419.72	-45.50		322.59		-62.50		472.87		-79.50
-12.00	120.00	-29.00	415.57	-46.00		320.67		-63.00		471.77		-80.00
-12.50	120.00	-29.50	411.53	-46.50		318.79		-63.50		470.69		-80.50
-13.00	120.00	-30.00	407.60	-47.00		316.94		-64.00		469.63		-81.00
-13.50	120.00	-30.50	403.77	-47.50		315.13		-64.50		468.58		-81.50

Programme VARTHERM 4(2): print-out for 765°C peak temperature.

TIME/TEMPERATURE/°F/°C	1.075	Q1=	3526.000	10=	120	QM=	650.00	TS=	3.5	G=	10.00
5.00	120.90	-16.00	569.92	-51.00	387.13	-48.00	507.50	-48.00		507.50	
2.50	120.00	-14.50	561.84	-51.50	383.86	-48.50	505.86	-48.50		505.86	
2.00	120.00	-15.00	555.94	-52.00	380.66	-49.00	504.25	-49.00		504.25	
1.50	120.00	-15.50	546.21	-52.50	377.54	-49.50	502.67	-49.50		502.67	
1.00	120.10	-16.00	538.68	-53.00	374.49	-50.00	501.12	-50.00		501.12	
0.50	120.10	-16.50	531.54	-53.50	371.52	-50.50	299.59	-50.50		299.59	
0.00	122.00	-17.00	524.19	-54.00	368.60	-51.00	298.08	-51.00		298.08	
-0.50	137.68	-17.50	517.24	-54.50	365.76	-51.50	296.60	-51.50		296.60	
-1.00	185.93	-18.00	510.48	-55.00	362.97	-52.00	295.15	-52.00		295.15	
-1.50	264.55	-18.50	503.90	-55.50	360.25	-52.50	293.72	-52.50		293.72	
-2.00	349.74	-19.00	497.51	-56.00	357.59	-53.00	292.31	-53.00		292.31	
-2.50	424.68	-19.50	491.30	-56.50	354.98	-53.50	290.92	-53.50		290.92	
-3.00	483.42	-20.00	485.27	-57.00	352.43	-54.00	289.56	-54.00		289.56	
-3.50	526.50	-20.50	479.40	-57.50	349.94	-54.50	288.22	-54.50		288.22	
-4.00	556.66	-21.00	473.70	-58.00	347.49	-55.00	286.90	-55.00		286.90	
-4.50	572.57	-21.50	468.17	-58.50	345.10	-55.50	285.59	-55.50		285.59	
-5.00	593.51	-22.00	462.78	-59.00	342.75	-56.00	284.31	-56.00		284.31	
-5.50	607.67	-22.50	457.55	-59.50	340.46	-56.50	283.05	-56.50		283.05	
-6.00	620.92	-23.00	452.46	-60.00	338.20	-57.00	281.81	-57.00		281.81	
-6.50	632.52	-23.50	447.51	-60.50	336.00	-57.50	280.59	-57.50		280.59	
-7.00	644.29	-24.00	442.70	-61.00	333.85	-58.00	279.38	-58.00		279.38	
-7.50	647.77	-24.50	438.02	-61.50	331.71	-58.50	278.19	-58.50		278.19	
-8.00	650.39	-25.00	433.46	-62.00	329.63	-59.00	277.02	-59.00		277.02	
-8.50	649.93	-25.50	429.02	-62.50	327.59	-59.50	275.87	-59.50		275.87	
-9.00	646.91	-26.00	424.71	-63.00	325.59	-60.00	274.73	-60.00		274.73	
-9.50	643.91	-26.50	420.50	-63.50	323.63	-60.50	273.61	-60.50		273.61	
-10.00	635.50	-27.00	416.40	-64.00	321.70	-61.00	272.51	-61.00		272.51	
-10.50	628.15	-27.50	412.41	-64.50	319.81	-61.50	271.42	-61.50		271.42	
-11.00	620.20	-28.00	408.52	-65.00	317.96	-62.00	270.35	-62.00		270.35	
-11.50	611.91	-28.50	404.73	-65.50	316.13	-62.50	269.29	-62.50		269.29	
-12.00	603.46	-29.00	401.03	-66.00	314.35	-63.00	268.25	-63.00		268.25	
-12.50	594.96	-29.50	397.43	-66.50	312.59	-63.50	267.22	-63.50		267.22	
-13.00	586.51	-30.00	393.91	-67.00	310.86	-64.00	266.20	-64.00		266.20	
-13.50	578.15	-30.50	390.48	-67.50	309.17	-64.50	265.20	-64.50		265.20	

Programme VARTHERM 4(2): print-out for 650°C peak temperature.

Y[F]=	0.792	Q1=	3528.000	T0=	120	QM=	650.00	TS=	3.5	N=	2.00	G=	10.00	
K0=	0.04200	K02=	0.00200	M=	0.0005140	BS=	551	A=	0.0000000659	B=-0.0000510073	MF=	188		
3.00	120.00	-14.00	664.67	-31.00	388.47	-48.00	293.27	-65.00	247.88					
2.50	120.00	-14.50	651.28	-31.50	384.20	-48.50	291.48	-65.50	246.91					
2.00	120.00	-15.00	638.51	-32.00	380.07	-49.00	289.72	-66.00	245.94					
1.50	120.00	-15.50	626.30	-32.50	376.06	-49.50	288.01	-66.50	244.99					
1.00	120.01	-16.00	614.64	-33.00	372.18	-50.00	286.32	-67.00	244.06					
0.50	120.88	-16.50	603.48	-33.50	368.41	-50.50	284.67	-67.50	243.14					
0.00	142.59	-17.00	592.80	-34.00	364.75	-51.00	283.06	-68.00	242.23					
-0.50	276.03	-17.50	582.57	-34.50	361.20	-51.50	281.47	-68.50	241.34					
-1.00	512.11	-18.00	572.77	-35.00	357.74	-52.00	279.92	-69.00	240.46					
-1.50	713.13	-18.50	563.36	-35.50	354.39	-52.50	278.39	-69.50	239.59					
-2.00	836.89	-19.00	554.32	-36.00	351.13	-53.00	276.89	-70.00	238.73					
-2.50	900.93	-19.50	545.64	-36.50	347.96	-53.50	275.42	-70.50	237.89					
-3.00	927.40	-20.00	536.19	-37.00	344.87	-54.00	273.98	-71.00	237.06					
-3.50	931.61	-20.50	526.26	-37.50	341.87	-54.50	272.57	-71.50	236.24					
-4.00	923.48	-21.00	516.60	-38.00	338.95	-55.00	271.18	-72.00	235.43					
-4.50	912.45	-21.50	507.36	-38.50	336.10	-55.50	269.81	-72.50	234.63					
-5.00	905.42	-22.00	498.53	-39.00	333.32	-56.00	268.47	-73.00	233.85					
-5.50	902.59	-22.50	490.10	-39.50	330.62	-56.50	267.16	-73.50	233.07					
-6.00	901.01	-23.00	482.03	-40.00	327.98	-57.00	265.86	-74.00	232.31					
-6.50	897.87	-23.50	474.31	-40.50	325.41	-57.50	264.59	-74.50	231.55					
-7.00	891.55	-24.00	466.92	-41.00	322.90	-58.00	263.35	-75.00	230.81					
-7.50	881.54	-24.50	459.82	-41.50	320.45	-58.50	262.12	-75.50	230.07					
-8.00	868.17	-25.00	453.02	-42.00	318.06	-59.00	260.91	-76.00	229.35					
-8.50	852.17	-25.50	446.47	-42.50	315.73	-59.50	259.73	-76.50	228.63					
-9.00	834.41	-26.00	440.19	-43.00	313.45	-60.00	258.56	-77.00	227.92					
-9.50	815.72	-26.50	434.15	-43.50	311.22	-60.50	257.41	-77.50	227.23					
-10.00	796.75	-27.00	428.31	-44.00	309.05	-61.00	256.28	-78.00	226.54					
-10.50	777.95	-27.50	422.69	-44.50	306.92	-61.50	255.17	-78.50	225.86					
-11.00	759.63	-28.00	417.28	-45.00	304.84	-62.00	254.08	-79.00	225.19					
-11.50	741.97	-28.50	412.06	-45.50	302.80	-62.50	253.01	-79.50	224.52					
-12.00	725.04	-29.00	407.01	-46.00	300.81	-63.00	251.95	-80.00	223.87					
-12.50	708.86	-29.50	402.14	-46.50	298.86	-63.50	250.91	-80.50	223.22					
-13.00	693.42	-30.00	397.43	-47.00	296.96	-64.00	249.89	-81.00	222.59					
-13.50	678.70	-30.50	392.88	-47.50	295.09	-64.50	248.88	-81.50	221.95					

YFFJ=	0.551	61=	5528.000	T0=	120	QM=	650.00	IS=	3.5	N=	2.00	G=	10.00	MF=	188
KU=	0.04200	KU2=	0.06200	M=	0.0005140	RS=	551	A=	0.0000000659	B=	-0.0000510073				
5.00	120.00	-14.00	619.59	-31.00	377.09	-48.00	288.81	-65.00	245.52						
2.50	120.00	-14.50	608.81	-31.50	373.21	-48.50	287.12	-65.50	244.58						
2.00	120.00	-15.00	598.47	-32.00	369.45	-49.00	285.46	-66.00	243.66						
1.50	120.00	-15.50	588.58	-32.50	365.79	-49.50	283.83	-66.50	242.74						
1.00	120.00	-16.00	578.72	-33.00	362.24	-50.00	282.24	-67.00	241.85						
0.50	120.00	-16.50	569.42	-33.50	358.79	-50.50	280.67	-67.50	240.96						
0.00	120.00	-17.00	560.44	-34.00	355.44	-51.00	279.14	-68.00	240.09						
-0.50	125.78	-17.50	551.79	-34.50	352.17	-51.50	277.63	-68.50	239.22						
-1.00	289.18	-18.00	543.45	-35.00	349.00	-52.00	276.15	-69.00	238.37						
-1.50	429.47	-18.50	535.97	-35.50	345.91	-52.50	274.70	-69.50	237.54						
-2.00	567.28	-19.00	528.55	-36.00	342.90	-53.00	273.28	-70.00	236.71						
-2.50	650.58	-19.50	515.06	-36.50	339.97	-53.50	271.88	-70.50	235.90						
-3.00	683.50	-20.00	506.17	-37.00	337.11	-54.00	270.51	-71.00	235.10						
-3.50	714.81	-20.50	497.66	-37.50	334.53	-54.50	269.16	-71.50	234.30						
-4.00	750.69	-21.00	489.49	-38.00	331.62	-55.00	267.83	-72.00	233.52						
-4.50	788.47	-21.50	481.66	-38.50	328.97	-55.50	266.53	-72.50	232.75						
-5.00	844.51	-22.00	474.15	-39.00	326.39	-56.00	265.25	-73.00	231.99						
-5.50	951.10	-22.50	466.92	-39.50	323.88	-56.50	264.00	-73.50	231.24						
-6.00	1057.94	-23.00	459.98	-40.00	321.42	-57.00	262.76	-74.00	230.50						
-6.50	1203.58	-23.50	453.51	-40.50	319.02	-57.50	261.55	-74.50	229.77						
-7.00	1385.50	-24.00	446.88	-41.00	316.67	-58.00	260.35	-75.00	229.05						
-7.50	1604.70	-24.50	440.69	-41.50	314.38	-58.50	259.18	-75.50	228.34						
-8.00	1860.15	-25.00	434.73	-42.00	312.15	-59.00	258.02	-76.00	227.64						
-8.50	2152.45	-25.50	428.98	-42.50	309.96	-59.50	256.89	-76.50	226.95						
-9.00	2482.46	-26.00	423.43	-43.00	307.82	-60.00	255.77	-77.00	226.26						
-9.50	2850.89	-26.50	418.07	-43.50	305.73	-60.50	254.67	-77.50	225.59						
-10.00	3268.58	-27.00	412.90	-44.00	303.69	-61.00	253.59	-78.00	224.92						
-10.50	3735.42	-27.50	407.89	-44.50	301.69	-61.50	252.53	-78.50	224.26						
-11.00	4262.55	-28.00	403.06	-45.00	299.73	-62.00	251.48	-79.00	223.61						
-11.50	4849.59	-28.50	398.58	-45.50	297.81	-62.50	250.45	-79.50	222.97						
-12.00	5496.70	-29.00	395.85	-46.00	295.94	-63.00	249.43	-80.00	222.34						
-12.50	6204.54	-29.50	389.46	-46.50	294.10	-63.50	248.43	-80.50	221.71						
-13.00	6982.50	-30.00	385.21	-47.00	292.30	-64.00	247.45	-81.00	221.09						
-13.50	7830.78	-30.50	381.09	-47.50	290.54	-64.50	246.48	-81.50	220.48						

Programme VARTHERM 5: print-out for 765°C peak temperature.

Y(F)=	1.075	Q1=	3528.000	T0=	120	QM=	650.00	TS=	3.5	N=	2.00	G=	10.00
K0=	0.04200	K02=	0.06200	M=	0.0005140	RS=	551	A=	0.0000000659	B=-	0.0000510073	MF=	188
3.00	120.00		-14.00	569.92		-51.00	364.17		-48.00		283.61		242.74
2.50	120.00		-14.50	561.84		-31.50	360.71		-48.50		282.03		241.85
2.00	120.00		-15.00	553.94		-32.00	357.36		-49.00		280.48		240.96
1.50	120.00		-15.50	546.21		-32.50	354.09		-49.50		278.95		240.09
1.00	120.00		-16.00	538.80		-33.00	350.90		-50.00		277.46		239.23
0.50	120.10		-16.50	528.95		-33.50	347.80		-50.50		275.99		238.39
0.00	122.06		-17.00	520.25		-34.00	344.78		-51.00		274.55		237.55
-0.50	137.48		-17.50	511.81		-34.50	341.84		-51.50		273.14		236.73
-1.00	165.93		-18.00	503.65		-35.00	338.97		-52.00		271.75		235.92
-1.50	264.39		-18.50	495.76		-35.50	336.17		-52.50		270.38		235.12
-2.00	349.74		-19.00	488.14		-36.00	333.45		-53.00		269.04		234.33
-2.50	424.68		-19.50	480.78		-36.50	330.78		-53.50		267.73		233.55
-3.00	465.42		-20.00	473.67		-37.00	328.19		-54.00		266.43		232.78
-3.50	526.50		-20.50	466.80		-37.50	325.65		-54.50		265.16		232.02
-4.00	556.66		-21.00	460.17		-38.00	323.17		-55.00		263.91		231.27
-4.50	577.57		-21.50	453.76		-38.50	320.76		-55.50		262.68		230.53
-5.00	593.51		-22.00	447.56		-39.00	318.39		-56.00		261.47		229.80
-5.50	607.67		-22.50	441.57		-39.50	316.08		-56.50		260.29		229.08
-6.00	620.92		-23.00	435.78		-40.00	313.83		-57.00		259.12		228.37
-6.50	632.62		-23.50	430.18		-40.50	311.62		-57.50		257.97		227.67
-7.00	641.79		-24.00	424.75		-41.00	309.46		-58.00		256.84		226.98
-7.50	647.77		-24.50	419.51		-41.50	307.35		-58.50		255.73		226.30
-8.00	656.39		-25.00	414.42		-42.00	305.29		-59.00		254.63		225.62
-8.50	649.93		-25.50	409.50		-42.50	303.27		-59.50		253.55		224.96
-9.00	646.91		-26.00	404.73		-43.00	301.29		-60.00		252.49		224.30
-9.50	641.91		-26.50	400.11		-43.50	299.36		-60.50		251.45		223.65
-10.00	635.50		-27.00	395.63		-44.00	297.46		-61.00		250.42		223.01
-10.50	628.15		-27.50	391.28		-44.50	295.60		-61.50		249.41		222.38
-11.00	620.20		-28.00	387.06		-45.00	293.79		-62.00		248.42		221.75
-11.50	611.91		-28.50	382.96		-45.50	292.00		-62.50		247.43		221.13
-12.00	603.46		-29.00	378.98		-46.00	290.26		-63.00		246.47		220.52
-12.50	594.96		-29.50	375.12		-46.50	288.55		-63.50		245.52		219.92
-13.00	586.51		-30.00	371.36		-47.00	286.87		-64.00		244.58		219.32
-13.50	578.15		-30.50	367.71		-47.50	285.23		-64.50		243.65		218.75

Programme VARTHERM 5: print-out for 650°C peak temperature.