



**THE INFLUENCE OF ALLOYING ADDITIONS ON THE
MICROSTRUCTURE AND MECHANICAL PROPERTIES IN HIGH
CURRENT GTA WELDING OF STRUCTURAL STEEL**

by

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ABSTRACT

The possibility of joining carbon-manganese (C-Mn) steel using the High Current Gas Tungsten Arc Welding (HC-GTAW) process was investigated.

HC-GTAW effectively allows the welding of thick materials with minimal edge preparation. High currents (500 - 1000 Amps) generate large arc forces which promote the formation of a deep cavity in the molten weld pool, within which the tip of the tungsten electrode may be immersed.

HC-GTAW has been used successfully to join non-ferrous metals. However, due to high dilution from the base material (up to 100% for autogenous welding) and solid state transformations during cooling, difficulties arise in achieving Charpy toughness requirements when using this process to join C-Mn steel.

An experimental programme was undertaken to investigate ways to improve the microstructure and subsequent mechanical properties of C-Mn steel welds by introducing a low carbon steel filler wire to the weld pool. This reduced dilution from the base material to approximately 75% and lowered the carbon content of the weld. The influence of the two other major alloying elements, manganese and silicon, was studied. The control of these elements is necessary to optimise the acicular ferrite content and hence the toughness. This was achieved by producing welds with varying levels of silicon (0.08% to 0.47%) and manganese (0.53% to 1.71%) and examining the resultant microstructures and mechanical properties.

Experimental techniques employed to characterise the microstructure included chemical analysis, microscopy (light, scanning and transmission electron), wavelength dispersive spectrometry, and x-ray powder diffraction. The mechanical properties were determined through hardness, all weld metal tensile and Charpy V notch testing.

Results indicated that the addition of silicon had minimal effect on the microstructure and mechanical properties of the welds, but did however, influence molten weld pool fluid motion which in turn affected the depth of penetration. The addition of manganese increased the proportion of acicular ferrite and led to increases in hardness and tensile strength. The improved microstructure did not however translate into improved Charpy toughness. The degradation in toughness was attributed to an increased presence of unstable retained austenite in the high manganese welds, which under impact could transform to brittle martensite.

DECLARATION

This work contains no material which has been accepted for the award of any other degree or diploma in any university or other tertiary institution and, to the best of my knowledge and belief, contains no material previously published or written by another person except where due reference has been made in the text.

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David M. Viano

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ACROYNYS

A-TIG	Active Tungsten Inert Gas
AC	Alternating Current
AS	Australian Standard
ASM	American Society for Metals
ASME	American Society of Mechanical Engineers
ASTM	American Society for Testing and Materials
AWS	American Welding Society
BCC	Body Centred Cubic (structure)
BF	Bright Field
BHP	Broken Hill Proprietary Limited
CDF	Centred Dark Field
C-Mn	Carbon - Manganese (steel)
CRC-MWJ	Cooperative Research Centre for Materials Welding and Joining
CSIRO	Commonwealth Scientific and Industrial Research Organisation
CVN	Charpy V Notch
DC	Direct Current
DCEN	Direct Current Electrode Negative
DCEP	Direct Current Electrode Positive
EDM	Electric Discharge Machining
EDS	Energy Dispersive Spectroscopy
FAC	Ferrite with Aligned Carbide
FCC	Face Centred Cubic (structure)
FESEM	Field Emission Scanning Electron Microscope
FS	Ferrite with Second Phase
FS(NA)	Ferrite with Non-Aligned Second Phase
GTAW	Gas Tungsten Arc Welding
HC-GTAW	High Current Gas Tungsten Arc Welding
HSLA	High Strength Low Alloy (Steel)
HVEM	High Voltage Electron Microscope
ICP-AES	Inductively Coupled Plasma Atomic Emission Spectrometry
IIW	International Institute of Welding

RE	Rare Earth (elements)
ROI	Region of Interest
SAD	Selected Area Diffraction
SEM	Scanning Electron Microscope
TEM	Transmission Electron Microscope
TWI	The Welding Institute
WDX	Wavelength Dispersive X-Ray Spectrometry
WIA	Welding Industries Australia
WRC	Welding Research Council
WTIA	Welding Technology Institute of Australia
XRPD	X Ray Powder Diffraction

NOMENCLATURE

α	Ferrite
ϕ	Angle between the normal to the interface at the particular point and the welding velocity vector
δ	Delta ferrite
θ	Angle at which a XRPD intensity peak was observed
γ	Austenite
λ	Wavelength of the radiation used for XRPD
ρ	Density
μ_0	Permeability in free space
σ	Surface tension
A_1	Area, defined in Figure 6-10
A_2	Area, defined in Figure 6-10
A_3	Area, defined in Figure 6-10
C.E.	Carbon Equivalent
C_o	Original solute content
d	Spacings between the crystalline lattices
d/w	Depth-to-width ratio
D	Dilution

E_w	IIW carbon equivalent
F_{arc}	Arc force (N)
FS	Ferrite with second phase
G	Thermal gradient in the liquid
I	Current (Amps)
k	Equilibrium distribution coefficient
m	Deoxidation parameter
MAC	Martensite, austenite, carbide microphase
P	Pressure
P_{cm}	Ito - Bessyo carbon equivalent
PF(G)	Grain boundary ferrite
PF(I)	Intragranular ferrite
R	Growth rate of the solid at a given point on the moving solid-liquid interface
R_1	Radius of the arc in the cathode region (electrode tip)
R_2	Radius of the arc in the anode region (weld pool)
T	Temperature
UTS	Ultimate tensile strength (MPa)
V_c	Critical welding velocity
V_w	Welding velocity



1. INTRODUCTION

In order to satisfy the ever increasing demand for improvements in the toughness of weld deposits in engineering structures, there is an incentive to improve welding consumables and to design new welding methods [1].

The metal fabrication industry is constantly searching for new methods of joining which can improve joint quality and/or productivity. Many aspects determine the productivity of welding. Some factors which need to be considered include the complexity of joint preparation and fit-up, the number of passes required, the amount (if any) of filler wire needed, the time it takes to complete a joint, and whether or not any post weld finishing operations such as grinding are required.

High current (HC) Gas Tungsten Arc Welding (GTAW), also known as “high density”, “buried arc”, “immersed arc” and “sub-surface” GTAW, is a variation of conventional GTAW [2]. HC-GTAW effectively allows the welding of thick materials with minimal edge preparation, and hence improves productivity. High currents (500 - 1000 Amps) generate large arc forces which promote the formation of a deep cavity in the molten weld pool, within which the tip of the electrode may be located. These conditions effectively localise the heat source nearer to the plate centre, and increase the heat transfer efficiency by reducing radiation loss [2].

HC-GTAW has been used successfully to join non-ferrous metals such as titanium [3] and aluminium [4]. To date the only research on applying HC-GTAW for joining steel has been carried out by Shtrikman *et al* [5] in 1973 and Lukens [6] in 1987.

Shtrikman *et al* [5] reported acceptable mechanical properties in a medium alloy steel weld after a post weld heat treatment.

Lukens [6] conducted a feasibility study into using the HC-GTAW process to join high strength low alloy (HSLA) steel. He was able to successfully join 19 mm HSLA plate with a carbon content of 0.05% autogenously (without the addition of filler wire) and achieve satisfactory mechanical properties.

When welding autogenously, the molten weld pool is composed entirely of the melted base material, which means that the chemical composition of the weld should match the base material. Structural carbon-manganese (C-Mn) steels have a carbon content of approximately 0.15%, so that autogenous welding of a C-Mn steel would result in a joint with unacceptable mechanical properties. A means of overcoming this problem is to introduce a gap between the abutting plates and add a low carbon steel filler wire. This reduces the dilution from the base material from 100% to approximately 75% and lowers the weld carbon content to a tolerable level of approximately 0.10%. Due to the high cooling rates associated with welding and the solid state transformations occurring during the cooling of ferrous metals, the weld metal can have a leaner chemical composition than the base material whilst matching strength and toughness.

The other two major alloying elements in C-Mn steel are manganese and silicon. Manganese is added to improve strength and to some extent assist in deoxidation. Silicon is principally added to assist in deoxidation.

In the weld deposit, manganese lowers the solid state transformation temperature to ferrite which refines the microstructure and improves strength. Silicon influences the weld pool

fluidity and has also been reported to increase strength. Both manganese and silicon act as deoxidisers in the weld pool and therefore influence oxide inclusion composition and subsequent acicular ferrite nucleation.

This present study investigates the influence of varying these two elements, silicon and manganese, on the microstructure and mechanical properties of HC-GTA welds in C-Mn steel.

2. HIGH CURRENT GTAW

2.1 BACKGROUND - GAS TUNGSTEN ARC WELDING

Gas tungsten arc welding (GTAW) is a welding process involving an arc between a non-consumable tungsten electrode and the weld pool. The process is used with a shielding gas which is usually inert. No pressure is involved in making the joint. Filler materials are introduced in some applications [7].

Although it was first promoted for the welding of aluminium, GTAW is suitable for joining most metals and alloys. These include ferrous, copper and nickel alloys, active metals such as magnesium and titanium, and refractory metals, including tungsten and tantalum [7]. Many of these materials are particularly difficult to weld by other means.

The essential components of GTAW are the shielding gas, the tungsten electrode, the power source and the material to be welded. Each is critical in determining the characteristics of the process.

The properties of the shielding gas are important in regulating the properties of the arc, and the transfer of heat and momentum to the weld pool. For example, helium has a much higher ionisation potential than argon, and therefore requires higher arc voltages [8]. This means that, for a given current and arc length, more heat is produced with helium than with argon, and consequently helium is used when maximum heating capacity is required. On the other hand, argon provides better shielding, (it is heavier than air), gives a readily controlled arc, and is less expensive. The gases can be mixed, and others (eg, hydrogen) can be added to give a greater range of properties [7].

The tungsten electrode plays an important role in determining the current density within the arc. Pure tungsten has a relatively high work function (the energy required to remove an electron from a metal), and when used as a cathode the tip melts [8]. Other elements, such as thorium or lanthanum, are often added to lower the work function and prevent melting. Considerable study has also been devoted to the effects of electrode diameter and the shape of the tip [9].

The welding current can be Direct Current (DC), Alternating Current (AC) or pulsed. DC can be connected with the electrode as either positive (the anode) or negative (cathode). Since the electrons in the arc carry most of the kinetic energy, and also consume energy in escaping from the surface, approximately 70% of the heat flows to the anode. For this reason the tungsten electrode is usually connected as the cathode (DC Electrode Negative (DCEN)). However, the cathode spot produces a cleaning action which can be used to rid materials (eg aluminium) of refractory oxide layers. In those cases DC electrode positive (DCEP) or AC, may be used.

The current can also be applied in a DC pulsed mode. One advantage of this is that it allows some decoupling between penetration and heat input [10], which is desirable for control purposes. The reason is that the forces produced by the arc are dependent on both the magnitude of the current and its rate of change with time. Consequently, variation of either pulse frequency or pulse shape can be used to alter the arc forces while maintaining constant average power or heat input.

GTAW has many advantages and some disadvantages when compared to other welding processes. On the positive side, the process produces a high quality weld, is free of spatter, can be used with or without filler wire, and can be used with almost all metals. The equipment is relatively inexpensive and easy to control, making the process ideal for automation. Its limitations include slow welding speed, susceptibility to draughts (as in out-

door work), lower deposition rates than with consumable electrode welding and the need for slightly higher levels of operator skill. Finally, it is normally reserved for relatively thin sections less than 10 mm thick [7, 11].

2.2 BACKGROUND - HIGH CURRENT GAS TUNGSTEN ARC WELDING

The first development of high current GTAW was reported by Goryachev and Zelenin [3] in 1964. They reported on research carried out on titanium alloys welded with a tungsten electrode. They found that deep penetration welds could be achieved with currents greater than 500 Amps, by using a non-consumable electrode connected to an arc length controller. Butt joint weldments in 25.4 mm thick plate were fabricated in four passes (two from each side) without any edge preparation and with only minimal additions of filler wire in the capping runs.

In 1965 Liptak [4] reported that welding 5083 aluminium up to 32 mm thick was achieved in two passes using a square butt joint configuration, no added filler material, DCEN, a thoriated tungsten electrode and helium shielding gas.

Recently the Paton Welding Institute has developed a further variation, Active TIG (A-TIG) welding, which involves GTA welding with a flux containing active ingredients [12]. It is claimed that the A-TIG process can produce single pass full penetration welds in 12 mm C-Mn steel.

2.3 PRINCIPLES OF HC-GTAW

2.3.1 Arc Force

HC-GTAW relies on the generation of high arc forces to form a deep cavity in the weld pool within which the tip of the electrode may be located. However, excessively high arc forces may be detrimental to the quality of a weld [13]. Weld defects such as humped beads and undercutting have also been attributed to high arc forces [14, 15].

Converti [16] studied the arc forces associated with HC-GTAW and developed the following equation (Equation 2-1) which shows that the arc force mainly depends on the current, but the cross-sectional area of the arc also has an influence.

$$F_{\text{arc}} = \frac{\mu_0 I^2 (2 \ln R_2 + 1)}{8\pi R_1} \quad \text{Equation 2-1}$$

where F_{arc} is the arc force (N), μ_0 is the permeability in free space, I is the current (Amps), R_1 is the radius of the arc in the cathode region (electrode tip) and R_2 is the radius of the arc in the anode region (weld pool).

Figure 2-1 shows a schematic diagram of the HC-GTAW process. As indicated in Equation 2-1, arc force is proportional to the current squared. If the current is high enough, the arc forces are sufficient to create a crater in the weld pool. The arc length (the distance between the electrode tip and the weld pool) is proportional to the voltage. With the combination of high current and low voltage it is possible to locate the electrode below the surface of the plate.

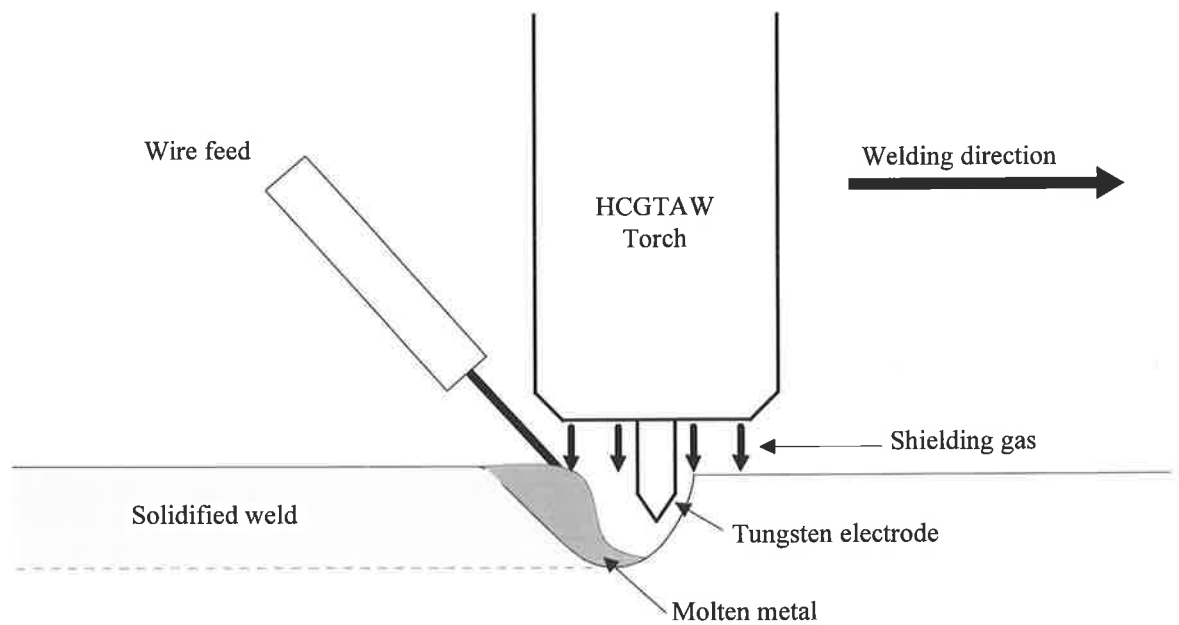


Figure 2-1: Schematic diagram of the HC-GTAW process

To maintain welding stability and consistent depth of penetration, the arc length and therefore arc voltage have to be carefully controlled. Nesterov *et al* [17] studied the effect of varying arc voltage on arc stability. Their results, shown in Figure 2-2, indicate that for any given electrode and material combination, there is an allowable arc voltage operating range which narrows as current increases. If the arc voltage is above this band, the weld pool motion becomes turbulent resulting in an uneven surface profile [17], sometimes known as humping [15]. Other researchers have also reported the formation of large internal voids caused by arc instability at high currents [18, 19]. Attempting to use an arc voltage below this operating band increases the risk of short circuiting the electrode with the front edge of the weld pool.

The reason for restriction of the maximum arc voltage is not clearly understood. Nesterov *et al* [17] postulated that with an increase in arc length (ie, increase in arc voltage) with other

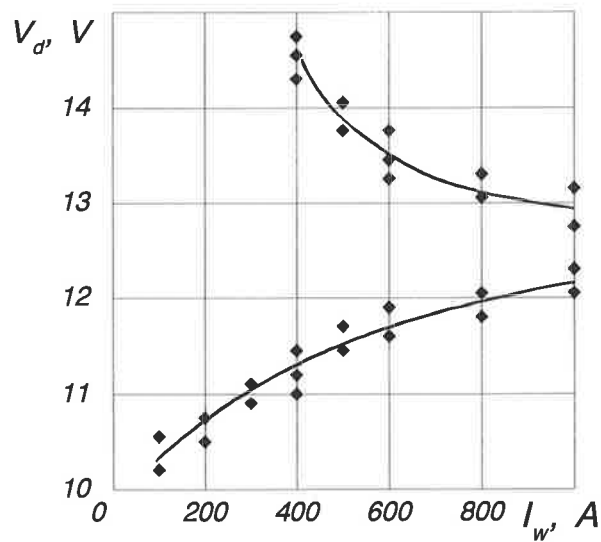


Figure 2-2: The allowable voltage operating range for a given electrode and material combination.

From Nesterov et al [17]

conditions remaining constant, the rigidity of the arc decreases. This loss of rigidity may cause periodic wandering of the arc which would lead to turbulent liquid metal flow and hence, an irregular outer surface.

2.3.2 Arc Length Control

Arc length can be maintained by a variety of methods, of which voltage control is the most common. Voltage control of arc length involves using a servomechanism which controls the electrode position depending on changes in the arc voltage. This process assumes that arc voltage is directly proportional to arc length and that it is independent of welding current. However, voltage tends to change slightly, for a given arc length, as current rises [4].

Arc voltage controllers monitor the voltage of the entire system, which is the combination of 3 separate voltage drops. These are:

1. the drop across the arc,
2. the drop across the tungsten electrode, and
3. the drop in the welding cables.

At start-up the voltage drop across the electrode and across the welding cables will rise due to the heat generated by the welding current. Consequently the controller reduces the arc length to maintain the sum of all 3 voltage drops constant [4]. Another problem associated with using this method is that the depth of penetration varies whenever the electrode is moved up or down. As a result, side wall defects may occur in two sided welding (due to a lack of penetration) [20].

Goryachev and Zelenin [3] briefly mentioned that arc length can be maintained by using a wire feed mechanism dependent on arc voltage. This has not been reported elsewhere.

Simonik and Vereshchagin [20] developed a control system which varied the current to maintain constant arc length and stability, thus allowing the electrode to remain at a predetermined depth. Consequently, “inertia-less” control could be carried out. To demonstrate the effectiveness of this system, welding was carried out on 10 mm thick plate with a gap varying from 0 to 4 mm. The current decreased as the gap widened to compensate for the smaller amount of metal to be melted while the electrode remained at the same depth. The current automatically increased again when the gap closed to zero width.

2.3.3 Influence of the Electrode Tip Immersion Depth

Adonyi *et al* [13] characterised the effect of electrode tip position on arc force using grooved, water-cooled, copper anodes. To simulate the geometric containment experienced by the arc, 12.7 mm rectangular grooves of varying depths were machined into the copper anodes. The

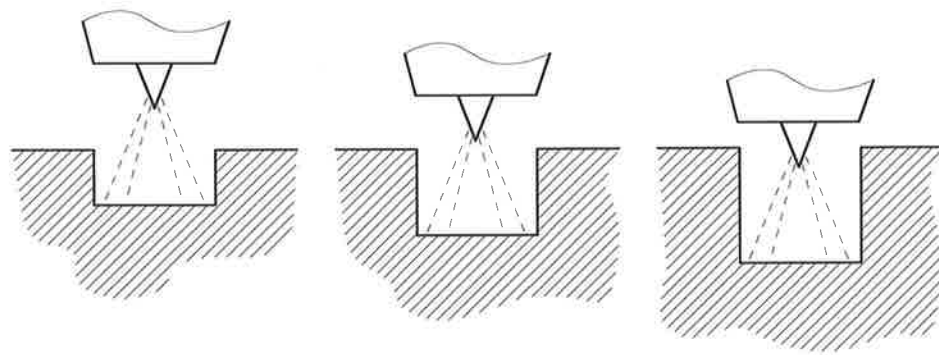


Figure 2-3: Variable groove depth geometries for subsurface GTA simulation experiments.

From Adonyi et al [13].

electrode position was adjusted to maintain a constant arc length, as illustrated in Figure 2-3. It was found that for a given arc length and current, arc force increased as the electrode immersion depth increased.

In another experiment, Adonyi *et al* [13] investigated the effect of reducing the groove width from 32 to 12.7 mm, maintaining a constant depth of 6 mm (for the non-consumable anode) as shown in Figure 2-4. The arc length used was 6 mm, which effectively meant zero electrode tip-to-work distance. The results indicated that the arc force reduces as the groove width increases.

The results of experiments on grooved, water-cooled, copper anodes clearly suggest an “arc containment” effect, such that an arc produced within a cavity surrounded by solid metal exerts a greater force than an arc produced on a flat open surface using the same parameters [13].

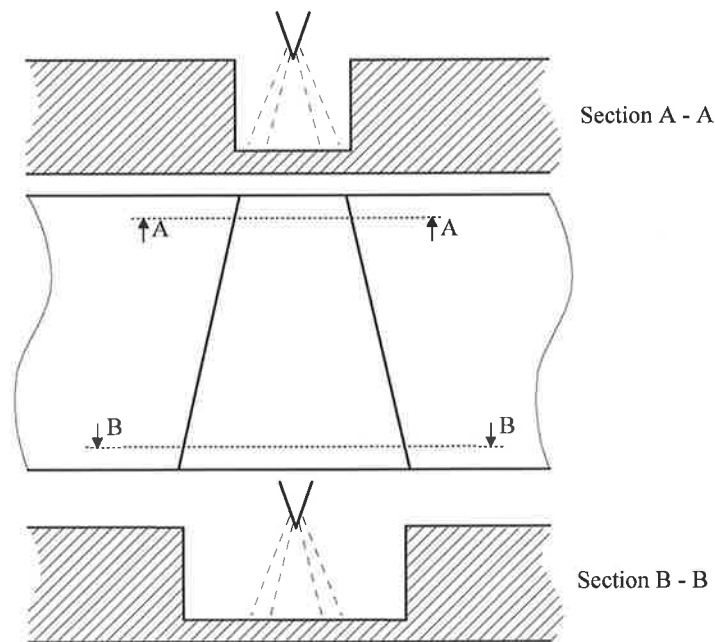


Figure 2-4: Variable groove width geometry for subsurface GTA simulation experiments.

From Adonyi et al [13].

2.4 WELD POOL FLUID FLOW

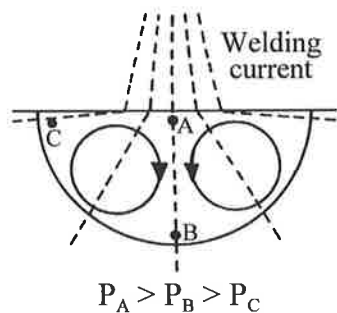
The size and the shape of the weld pool is influenced by the simultaneous occurrence of several important physical processes. These include the amount of heat transferred from the heat source to the work piece, the fluid flow in the weld pool, and the accompanying convective heat transfer [21]. Adjusting welding parameters such as current and voltage can alter the amount of heat transferred. The direction of the fluid flow and the accompanying convective heat transfer are influenced by a number of forces.

In a keynote address, Matsunawa [22] outlined the four forces which can induce flow in a molten weld pool. They are:

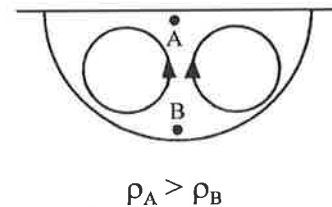
1. electromagnetic (Lorentz) force;
2. gravitational or buoyancy force;
3. surface tension (Marangoni flow); and
4. Aerodynamic drag force.

Lorentz and buoyancy forces are body forces which act inside the weld pool, while surface tension and aerodynamic drag are shear stresses which react only on the pool surface.

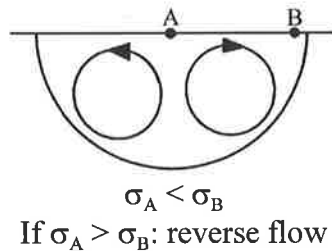
Figure 2-5 shows a schematic diagram of the flow direction by each motive force [22].



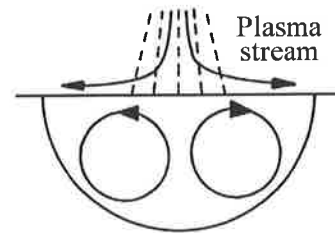
(a) Electromagnetic force



(b) Buoyancy force



(c) Surface tension



(d) Aerodynamic drag force

Figure 2-5: Flow directions induced by 4 possible motive forces in a weld pool.

From Matsunawa [22].

2.4.1 Electromagnetic (Lorentz) Force

When a current flows through a conductor the resultant magnetic field generates a compressive force on it. If the current is divergent (as in GTAW) then a pressure gradient will be established along the conduction path. In cases where the conductor is a fluid (molten weld pool) this will result in acceleration of the material in the direction of magnetic pressure [23]. Current entering the weld pool from the arc plasma is strongly divergent, and can be expected to generate a flow radially inward across the surface of the pool (see Figure 2-5(a)).

2.4.2 Buoyancy Forces

The density of a liquid generally reduces as the temperature increases. The temperature gradient in weld pool therefore results in a density gradient. This density gradient induces a small buoyancy force on the weld pool as indicated in Figure 2-5(b). Walsh and Nunes [24] reported that buoyancy forces are one to two orders of magnitude lower than either Lorentz or Marangoni forces and are therefore unlikely to influence weld pool flow.

2.4.3 Surface Tension (Marangoni Flow)

A model by Heiple and Roper [25] suggest that weld penetration is related to the direction of the surface tension driven flow (Marangoni flow) in the liquid metal pool. Surface tension driven flow arises in systems which contain a surface tension gradient, the direction of flow being from a region of low surface tension to one of high surface tension. A temperature gradient always exists across a weld pool surface and the accompanying surface tension gradient will generate fluid flow in a direction which depends on the sign of surface tension coefficient ($d\sigma/dT$) [26].

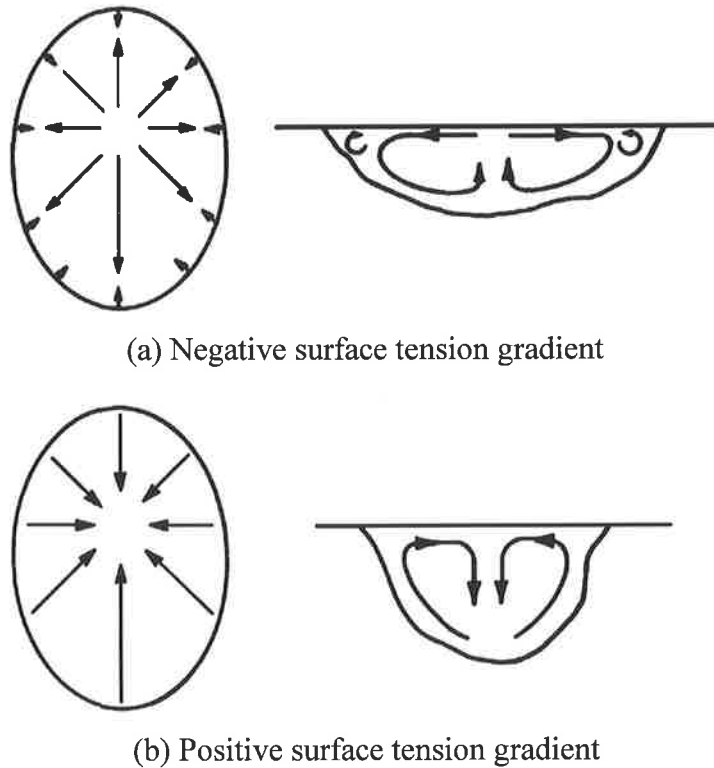


Figure 2-6: Surface and subsurface fluid flow in a weld pool.

From Heiple and Roper [25].

Pure metals exhibit a negative surface tension coefficient, i.e., the surface tension decreases as the temperature increases. Since the temperature is highest under the arc, the surface tension is highest at the edge of the weld pool. This gradient produces flow from the centre of the weld pool outward, producing efficient heat transfer to the sides of the weld and a wide shallow weld pool as shown in Figure 2-6(a). Alternatively, the addition of trace amounts of surface active elements such as sulphur and oxygen produce a positive surface tension coefficient. This produces a strong inward flow on the weld pool surface and results in a narrow deep weld pool as shown in Figure 2-6(b) [27].

The model by Heiple and Roper was designed for conventional GTAW and assumes a flat weld pool surface. Due to the higher arc forces associated with HC-GTAW the weld pool

surface would be deformed. Choo *et al* [28] modelled weld pool circulation of deformed weld pools and found that the shape of the weld pool surface markedly affected the circulation.

2.4.4 Shear Stress by the Plasma Stream

The current density in a welding arc is usually much higher in the vicinity of the electrode than at the base material. A plasma stream is generated by the axial component of the electromagnetic force. This plasma stream interacts with the weld pool in various ways. These include: depression of the weld pool by a momentum change at jet impingement, friction forces on the liquid surface, and convective heat transfer from the hot plasma to the base material [22].

2.5 WELDING EQUIPMENT

Improvements to the GTA process, such as heavier duty electrode holders and improved tungsten electrodes, have occurred over time. These advances have been evolutionary in nature and have led to significant increases in maximum operating current [4]. Standard GTA power supplies are suitable for high current GTAW and two compatible power supplies can be connected in parallel to increase welding current capability [13].

Weld pools produced by HC-GTAW are larger than conventional GTAW, and hence larger, water-cooled gas shrouds are required [4]. A trailing gas cover is also needed for some alloys, such as titanium, to prevent oxidation of the cooling weld pool [5].

CSIRO Manufacturing Science and Technology developed a HC-GTAW torch, shown in Figure 2-7, capable of operating at currents up to 1000 Amps at 100% duty cycle (continuous

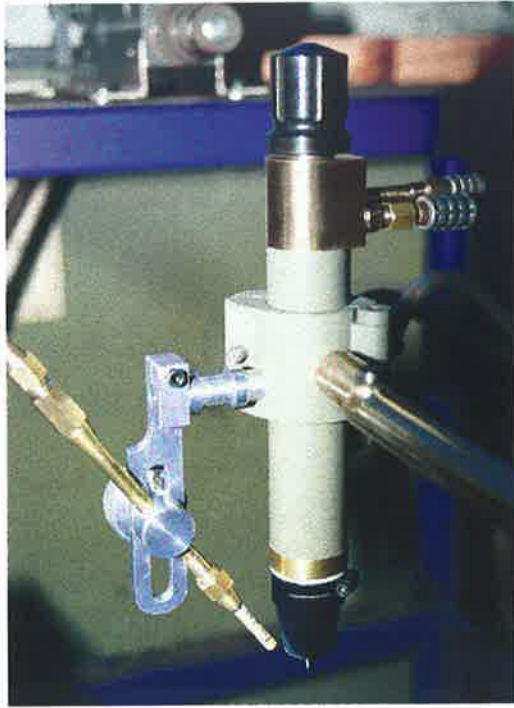


Figure 2-7: HC-GTAW torch developed by CSIRO.

operation) [29]. Special features of this design include a water cooled barrel surrounding the electrode and a water cooled gas shroud.

2.6 MATERIALS USED

A variety of materials have been welded using HC-GTAW. Initial work was carried out on titanium and aluminium alloys [3, 4, 30 - 32]. Since then stainless steels [5, 13, 33], low alloy steels [6] and magnesium alloys [17, 34] have also been welded using this process.

2.7 APPLICATIONS OF HC-GTAW

To date most HC-GTAW has been carried out in the downhand position. However, in 1971 Rager [35] demonstrated HC-GTAW of 2014 aluminium in both the vertical and horizontal positions. Lukens [6] commented that this process could also be used for web-to-flange fillet welding.

Two major advantages of this system are that the edge preparation is minimised and the need for filler wire additions is reduced or eliminated [4]. According to Liptak [4], the elimination of filler wire in welding alloys such as aluminium leads to cost savings of 25 - 35%, which offsets the disadvantage of the reduced welding speeds of between 100 and 300 mm/min compared with conventional GTAW [6, 13, 14]. HC-GTAW also makes it possible to reduce the number of passes required [5].

The principal drawbacks of HC-GTAW are the narrow parameter operating range and welding process control difficulties [2].

3. MICROSTRUCTURE DEVELOPMENT

3.1 INTRODUCTION

To be able to understand the development of microstructure which controls mechanical properties one needs to understand what happens during the solidification and subsequent phase transformations of the weld deposit and the influence of welding parameters and weld deposit composition.

3.2 SOLIDIFICATION

3.2.1 Introduction

Melting and solidification are two of the most important reactions involved in fusion welding [36]. Solidification behaviour controls the size and shape of grains, the extent of segregation, and the distribution of inclusions and defects such as porosity and hot cracks [37]. Since the properties and integrity of the weld metal depend on the solidification behaviour and the resulting microstructural characterisations, understanding weld pool solidification behaviour is essential [37].

Chalmers and co-workers [38] are credited with developing modern solidification principles. Pioneering research in the area of weld solidification, however, was carried out by Savage and co-workers [39]. During the usual solidification process the transformation of liquid phase to a solid normally occur by a process of nucleation and growth. In the fusion welding process nucleation is not significant since a solid-liquid interface is always present and epitaxial growth may occur from the unmelted base material [39].

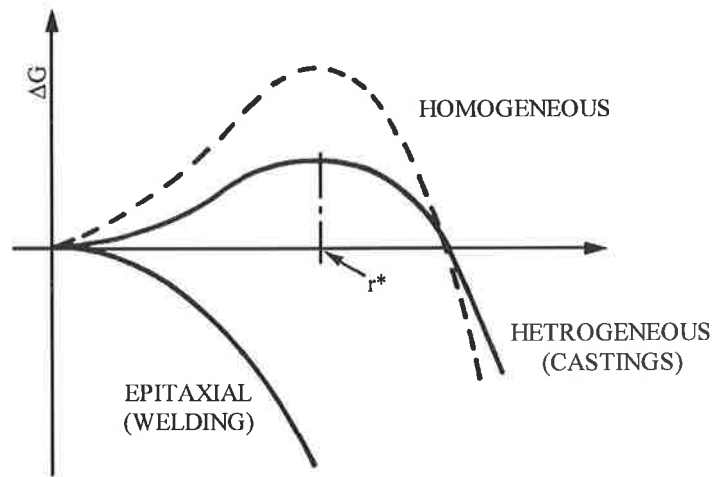


Figure 3-1: Comparison of free energy of formation vs particle radius for homogeneous, heterogeneous, and epitaxial nucleation.

From David and Vitek [37].

The weld metal almost perfectly wets partially-melted grains in the Heat Affected Zone (HAZ) of the parent plate and growth takes place from these grains into the weld pool after the passage of the welding heat source [40].

3.2.2 Nucleation

Welding involves a special case of heterogeneous nucleation. Under autogenous welding conditions, the molten metal composition is identical to the base material and initial growth is often epitaxial in nature. The nucleation barrier vanishes and nucleation occurs spontaneously below the liquidus temperature as indicated in Figure 3-1 [37, 40].

In non-autogenous welding, in which a filler metal is added, epitaxial growth may still occur.

3.2.3 Growth

3.2.3.1 Shape of the Weld Pool

On average, the growth of the solid occurs parallel to the maximum temperature gradient, which is perpendicular to the advancing solid-liquid interface. Figure 3-2 shows a schematic diagram of moving weld pool (taken from Savage [36]). For the shape of the weld pool to remain constant, the growth rate of the solid must be given by (Equation 3-1):

$$R = V_w \cos \phi \quad \text{Equation 3-1}$$

where R is the growth rate of the solid at a given point on the moving solid-liquid interface, V_w is the welding velocity and ϕ is the angle between the normal to the interface at the particular point and the welding velocity vector.

Therefore the growth velocity would vary from $R = 0$ at the sides of the weld pool (Points A and C) where $\phi = 90^\circ$, to a maximum of $R = V_w$ at the centreline of the weld (Point E).

Since the rate of liberation of latent heat is directly proportional to the rate of growth of the solid, the rate of heat liberation is maximised at the weld centreline. Unfortunately, the minimum temperature gradient, and therefore the minimum capability to conduct the latent heat to the cooler surroundings is found at this location. Consequently, when welding velocity reaches a critical value, V_c , such that the rate of growth parallel to the welding direction can no longer be sustained, the weld pool will assume a teardrop shape [36].

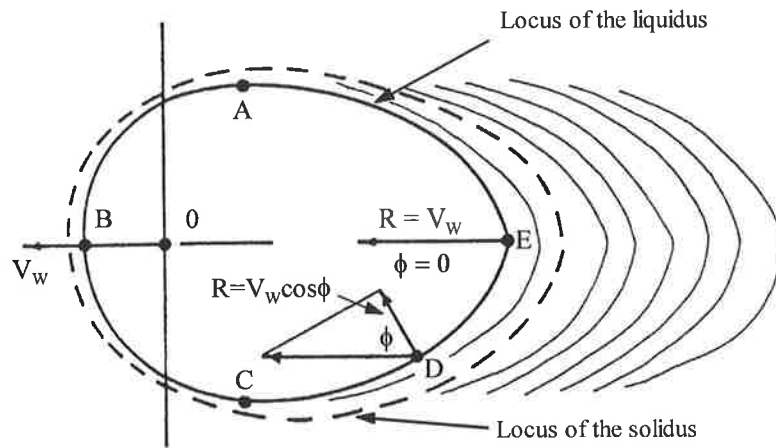


Figure 3-2: Schematic of a moving weld pool.

From Savage [36].

3.2.3.2 Epitaxial Growth

The base metal, being of similar composition to the molten metal, makes an ideal substrate upon which growth of a solid phase can occur. Thus to extend the solid phase, atoms from the liquid merely surrender their latent heat of fusion to the surroundings and occupy positions corresponding to lattice sites in an extension of the crystal structure of the existing solid [36]. In the presence of such an ideal substrate, the supercooling necessary to sustain growth rate in the solid is probably well under one degree Celsius [36].

Since growth during weld pool solidification begins from grains in the parent plate, the width of the weld metal grain at the fusion boundary will be determined by the width of that grain in the HAZ acting as the growth substrate. [36, 39, 41, 42].

3.2.3.3 Competitive Growth

In general, growth occurs most rapidly along the easy growth direction (the $\langle 100 \rangle$ for body-centred cubic materials) which has the largest component of the temperature gradient [43]. Thus, it has been found that the longitudinal axis of cells and cellular dendrites tends to correspond to a favourably oriented $\langle 100 \rangle$ crystallographic direction [44]. If several randomly oriented grains exist at the solid-liquid interface, the grain with a $\langle 100 \rangle$ direction most nearly parallel to the maximum temperature gradient enjoys a competitive advantage over neighbouring, less favourably oriented grains [43]. In this case, such a grain would grow more rapidly and would tend to crowd out the less favourably oriented grains unless a change in the direction of the maximum temperature gradient occurred. In the presence of a maximum temperature gradient whose direction changes, no single grain enjoys a competitive advantage. This interrelationship between the easy growth direction and the maximum temperature gradient causes what is known as competitive growth [43]. An illustration of the effect of competitive growth on the shape of grains in the fusion zone is shown in Figure 3-3.

In an elliptical weld pool, Figure 3-3(a), the direction of the maximum temperature gradient changes continuously from the fusion line to the centre of the weld and the grains tend to curve in the direction of welding as they follow the moving solid-liquid interface [43]. The competitive advantage shifts from grain to grain as the maximum temperature gradient changes direction [36]. Also, each grain tends to maintain contact with moving solid-liquid interface until it is pinched off in a random fashion by neighbouring grains [36].

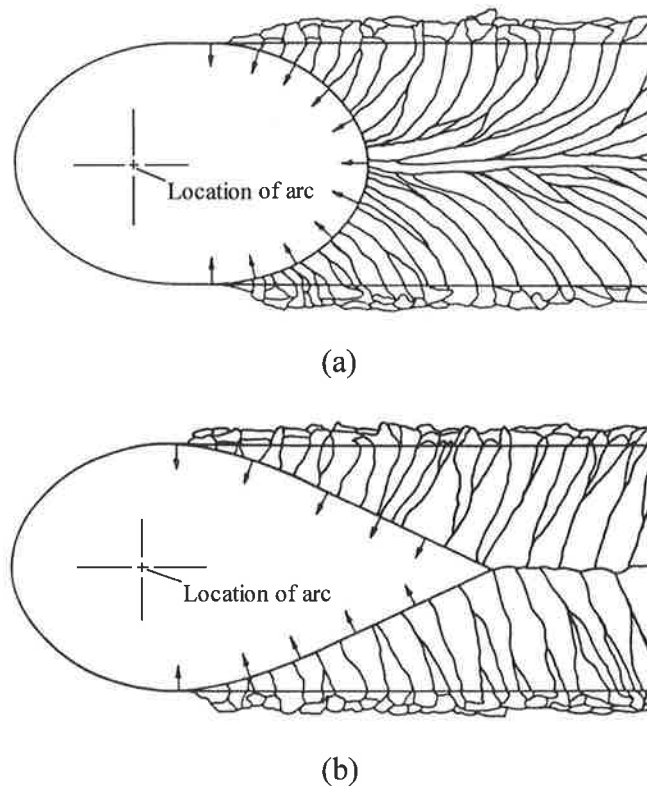


Figure 3-3: Schematic diagram showing competitive growth in (a) an elliptical-shaped weld pool, and (b) a teardrop-shaped weld pool.

From Savage [36].

With a teardrop-shaped weld pool, Figure 3-3(b), the direction of the maximum temperature gradient remains constant over most of the trailing edge of the weld pool [43]. Thus, favourably orientated grains will crowd out less favourably orientated neighbours and grow until impingement from favourably orientated grains from the opposite side of the weld pool. This leads to a high proportion of preferred orientation in the weld [36].

3.2.3.4 Constitutional Supercooling

In an alloy, the redistribution of solutes at the advancing solid-liquid interface can establish a condition known as constitutional supercooling which promotes dendritic solidification even in the presence of a positive temperature gradient at the solid-liquid interface [36].

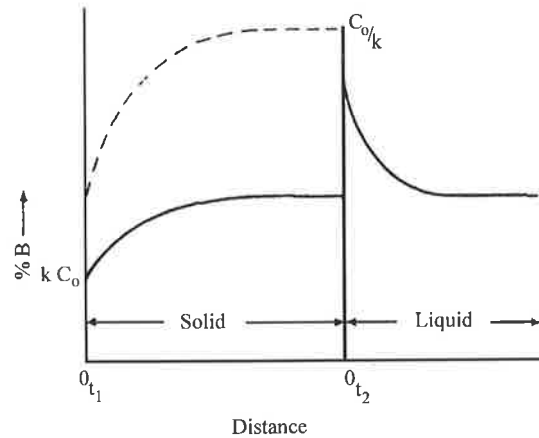


Figure 3-4: Schematic diagram showing solute distribution at initial-transient stage of solidification.

From Savage [36].

Figure 3-4 illustrates the distribution of solute at the beginning (0_{t_1}) and end (0_{t_2}) of the initial transient stage of solidification [36]. Once the initial-transient period is over, the motion of the solid-liquid interface will continue and the solid will have a nominal composition of C_0 . Since very little mechanical mixing of the liquid occurs in the immediate vicinity of the advancing interface, the rejected solute must be redistributed in the liquid by diffusion [39]. The freezing process is so rapid that diffusional process cannot effectively remove the excess solute near the interface [39]. Hence, the concentration gradient in the liquid will diffuse ahead of the interface without changing shape. As a result, the effective liquidus temperature varies with distance from the interface as shown in Figure 3-5.

The way this concentration gradient in the liquid ahead of the solid-liquid interface causes constitutional supercooling is also explained in Figure 3-5 [36]. At the lower left of Figure

3-5, the effective liquidus temperature and the actual temperature gradient are plotted against distance from the solid-liquid interface. The area where the actual temperature is below the effective liquidus temperature is known as the region of constitutional supercooling. If a chance protuberance were to form at the solid-liquid interface, it would grow into a region of undercooled liquid and could grow dendritically [36].

Conditions which give rise constitutional supercooling are: [40]

- low temperature gradients in the liquid
- fast growth rates

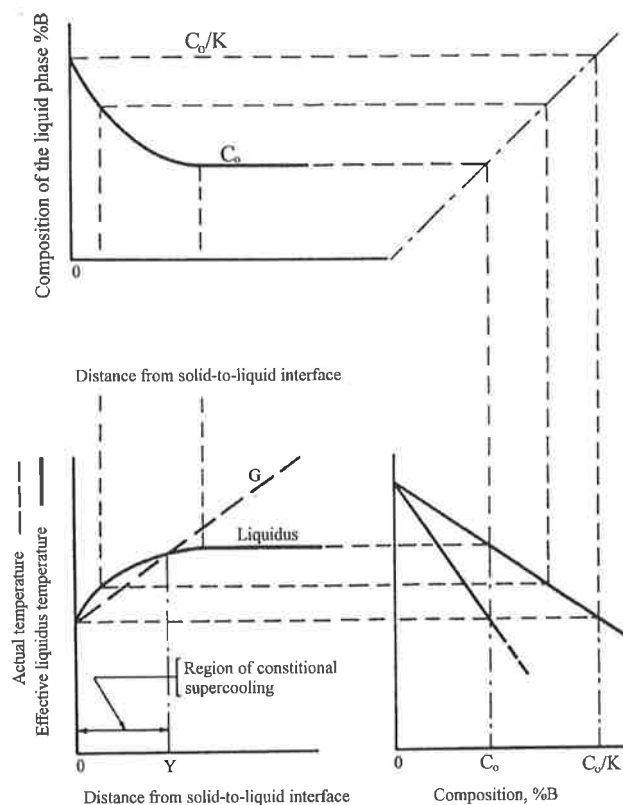


Figure 3-5: Schematic explanation of the origin of constitutional supercooling.

From Savage [36].

- steep liquidus lines
- high alloy content (greater than $\sim 0.2\%$)
- extreme values of the distribution coefficient k_0

3.2.3.5 Solidification Mode

The solute concentration, growth rate (R), and thermal gradient in the liquid (G) all influence the solidification mode as indicated in Figure 3-6. Low values of $G/(R)^{1/2}$ indicate an increased tendency for constitutional supercooling and favours the dendritic mode of solidification [39].

Savage *et al* [43] found that the solidification mode was dependent on both welding current and travel speed. As the current increased for a given travel speed, the solidification mode changed from cellular to cellular-dendritic. The same change in solidification mode was noted as the travel speed was decreased at constant welding current.

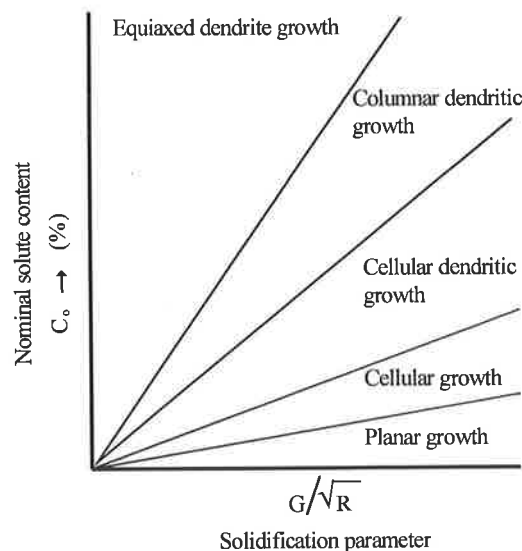


Figure 3-6: Effect of growth rate and temperature gradient on solidification mode.

From Savage [36].

At the side of the pool steep thermal gradients together with comparatively low solidification rates favour cellular growth while at the weld centre line, high solidification rates promote a transition to dendritic growth modes. As a result, a range of growth substructures can be observed in solidified welds [40].

3.2.4 Segregation

3.2.4.1 Microsegregation

A significant problem arising out of the solidification of weld metal is the microsegregation, which is a result of the redistribution of solute accompanying solidification [39, 43]. The level of purity in welding is such that microsegregation always occurs on solidification. Its extent is determined largely by the shape of the solidus and liquidus of the constitutional diagram for the segregating components [39].

For solutes which exhibit distribution coefficients less than 1.0, the molten weld metal becomes enriched in that solute ahead of the solid-liquid interface (as described in section 3.2.3.4). As solidification proceeds, this enriched liquid freezes in grain boundaries and cell boundaries [43].

Microsegregation is remarkably persistent and may only be reduced by prolonged high temperature treatments [39].

3.2.4.2 Macrosegregation

Macrosegregation occurs either by solute accumulation in the grain boundary grooves or by the impingement of two interfaces growing with a growth component normal to each

other. If a columnar growth predominates, the level of macrosegregation at the weld centre line can be very marked [40].

3.2.4.3 Solute Banding

Solute banding is a segregation pattern commonly observed in fusion welds. It occurs from changes in the growth rate during solidification resulting in periodic regions of solute enrichment or depletion [40].

Variations in the growth rate is generally attributed to a variety of factors including:

- periodicity in the heat flow associated with extraction of the latent heat of fusion [45, 46, 47];
- periodicity in heat input caused by slight fluctuations from the welding power source [45, 48]; or
- natural oscillation of the weld pool resulting from surface tension and weld pool surface temperature fluctuations [49].

Garland and Davies [50] reported that a rapid increase in the interface velocity will result in concentration of solute in the solidifying material. A decrease in the interface velocity will lead to a depletion of solute in the solidifying material. Figure 3-7 shows the effect of changes in the growth rate on localised solute distribution in the solid for an increase in growth rate (R_1), and a decrease in growth rate (R_2).

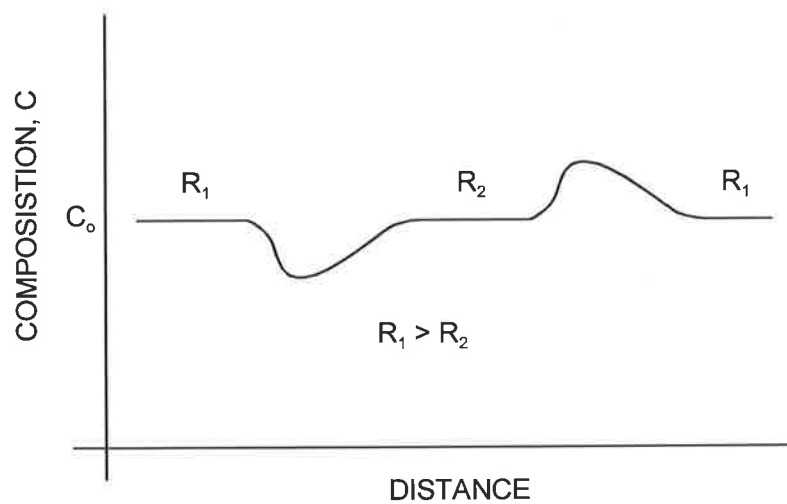


Figure 3-7: Effect of changes in growth rate R on localised solute distribution in the solid.

From Brooks and Mahin [51].

3.2.5 Relationship between Solidification Structure and Weld Properties

Since the mode of weld pool solidification controls the size and pattern of solidified grain development as well as the nature and extent of the associated solute segregation in the weld bead, a significant correlation would be expected between solidification pattern and the properties of the weld deposit [40].

Under particular welding conditions and with particular consumables, low and high carbon martensite and bainitic carbides form in groups of islands along prior solidification boundaries. This produces a marked decrease in as-welded impact toughness [40].

3.3 DELTA FERRITE TO AUSTENITE TRANSFORMATION

Generally C-Mn steel weld deposits solidify as delta ferrite which subsequently transforms to austenite. However if the carbon level or substitutional alloy content or cooling rate is

sufficiently high, then austenite would be the first solid to form and columnar austenite grains would grow directly from the melt [52].

As with delta ferrite, growth of the austenite will be parallel to the thermal gradient. However, under the influence of a moving heat source the thermal gradient is different from that which existed for delta ferrite solidification. Thus austenite grain boundaries are neither coincident nor parallel to delta ferrite grain boundaries and can cross over each other [48, 52].

Figure 3-8 shows a schematic diagram illustrating the location of inclusions in the microstructure of a weld which solidified as (a) δ ferrite and (b) austenite (γ), assuming that the inclusions locate themselves preferentially at the cell boundaries during solidification. Figure 3-8 (a) also shows the austenite grain boundaries diverging from the prior δ ferrite boundaries caused by the moving heat source [48, 52].

Transformation to austenite during cooling destroys evidence of the position of the delta ferrite grain boundaries. By using a solute-sensitive etch the original solidification boundaries can be revealed indirectly due to the segregation of elements with low partition coefficients such as manganese and nickel [52].

The columnar austenite grain size must to some extent correlate with the grain size in the parent plate at the fusion boundary, since solidification occurs by epitaxial growth of those grains [40]. Crystallographic texture from the parent plate, and the plane of that plate on which the weld is deposited, must influence the final austenite grain structure. Clear differences in the austenite grain structure were found between three welds deposited on

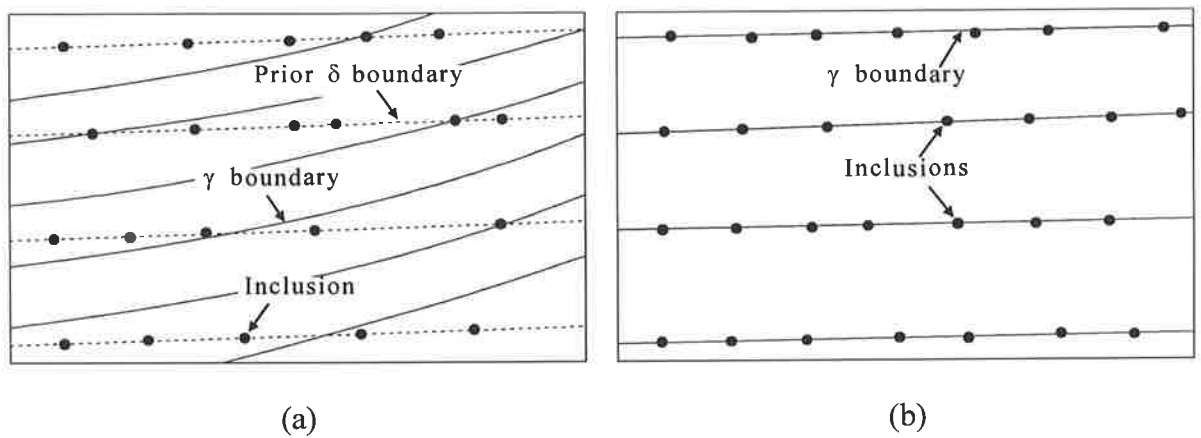


Figure 3-8: Schematic diagram showing the location of inclusions for a weld which solidifies as (a) δ ferrite and (b) austenite (γ).

From Sugden and Bhadeshia [52].

mutually perpendicular faces of the same sample, in a recent experiment designed to illustrate the influence of texture on the grain size [53, 54].

It has been common practice to define the austenite grain size from the transverse section of a weld, the size being measured not at random, but by aligning the test lines normal to the larger dimension of the grain sections [55].

3.4 AUSTENITE TO ALPHA FERRITE

The austenite to alpha ferrite transformation behaviour is affected by non-metallic inclusions, alloying additions and prior austenite grain size [56]. During austenite decomposition a large variety of microstructures can develop, depending on cooling rates and weld metal chemical composition. A variety of terminology is used by researchers to identify micro-constituents. This has led to confusion when comparing research results. The International Institute of Welding (IIW) has published guidelines designed to quantify volume fractions of micro-constituents by point counting [57]. Whilst this has allowed

weld deposits to be quantified using a standardised method, it is necessary to correlate the IIW terminology with conventional nomenclature. For example Widmanstätten ferrite (conventional) is designated ferrite with second phase or FS (IIW).

Generally solid state transformations in metals occur heterogeneously by nucleation at high energy sites such as grain corners, grain boundaries, inclusions, dislocations, and vacancy clusters. In addition, in steel weld deposits, fine distributions of non-metallic inclusions provide favourable sites for heterogeneous nucleation [58].

3.4.1 Allotriomorphic Ferrite (Grain Boundary Ferrite)

On cooling below the A_{c3} temperature, allotriomorphic ferrite will nucleate initially at austenite grain corners and boundaries, since these sites generally provide the lowest energy barrier to nucleation. The ferrite subsequently grows in a planar manner [58]. In low alloy steel welds, the boundaries rapidly become decorated with virtually continuous layers of ferrite, so that subsequent transformation simply involves the reconstructive thickening of these layers [53].

3.4.2 Widmanstätten Ferrite

With an increasing degree of undercooling, the redistribution of carbon at the allotriomorphic ferrite interface becomes insufficient to maintain a planar growth mode. Further growth of ferrite can only take place by lateral movement of ledges along a low energy interface [59]. Once nucleated, this structure known as Widmanstätten ferrite grows very rapidly under the prevailing conditions as a result of an efficient redistribution of carbon to the sides of the advancing interface [58]. The calculated growth rates are found to be so high for typical weld deposits, that the formation of Widmanstätten ferrite is

usually complete within a fraction of a second [53]. The formation of Widmanstätten ferrite can occur at relatively low driving forces [60, 61]. Widmanstätten ferrite is also known as ferrite side plates, and ferrite with second phase.

3.4.3 Acicular Ferrite

Acicular ferrite is a phase most commonly observed as austenite transforms during cooling of low alloy steel weld deposits. It is of considerable commercial importance because it provides a relatively tough and strong microstructure [53].

Simultaneously with or immediately after the formation of Widmanstätten ferrite at the austenite grain boundaries, acicular ferrite may start to form intragranularly at inclusions [58]. Subsequent plates may nucleate autocatalytically, so that a one to one correspondence between the number of active inclusions and the number of acicular ferrite plates is not expected [62]. Ito and Nakanishi [63] first reported the existence of acicular ferrite. A microstructure consisting of acicular ferrite is superior in toughness and strength to other transformation products due to its fine interlocking laths which offer excellent resistance to crack propagation [56].

Acicular ferrite nucleates in the transformation temperature range between allotriomorphic ferrite and upper bainite [62, 64 - 66]. It is believed to be formed by a ledge mechanism and each lath is typically 1 - 3 μ m wide with an aspect ratio between 2:1 and 4:1 [62]. Acicular ferrite laths normally grow until impingement occurs on neighbouring grains thus the final grain size will largely be determined by the average spacing between inclusions [58]. Bhadeshia and Svensson [67, 68] reported that the coarsening of acicular ferrite *per se* is unlikely to lead to any drastic changes in the strength of the weld deposit. The

amount of acicular ferrite decreases with the austenite grain size, as grain boundary nucleated phases such as allotriomorphic ferrite become more dominant [69].

Recently, Bhadeshia and co-workers [70 - 75] have provided evidence that acicular ferrite is essentially intragranularly nucleated bainite. Acicular ferrite does not normally grow in sheaves like bainite because of hard impingement between plates nucleating independently at adjacent sites (inclusions).

Babu and Bhadeshia [70] concluded that the formation of acicular ferrite can be enhanced by reducing the number density of austenite grain boundary nucleation sites. This could be done by decorating the austenite grain surfaces with thin layers of allotriomorphic ferrite. However, the allotriomorphic ferrite must be rendered inert to prevent the formation of Widmanstätten ferrite and bainite by having sufficient carbon to partition in the austenite at the ferrite/austenite interface.

3.4.4 Microphases

In low carbon alloy welds microphases usually represent only a small percentage of the micro-constituents. Microphases exist from carbon diffusing into the remaining austenite during transformation to ferrite. The remaining austenite which can be enriched with up to 7 at% carbon may transform into a large variety of microphases such as pearlite, bainite and martensite or it can remain as inter-lath islands of retained austenite within an acicular ferrite structure [76 - 78].

During solidification, manganese segregates to the remaining melt which is thus enriched at the cell boundaries and at the solid/liquid interface. Especially when the weld is heavily alloyed with manganese, extensive formation of microphases takes place in areas where the

manganese content is high. In high manganese welds the microphases appear as bands, reflecting the solidification pattern, while in low manganese welds, the microphases are more evenly distributed [79].

3.5 THE ROLE OF INCLUSIONS IN C-MN STEEL WELDS

As previously mentioned in section 3.4.3, acicular ferrite is nucleated intragranularly from inclusions within the austenite grains. It is therefore imperative to understand how inclusions are formed and which inclusions favour the nucleation of acicular ferrite.

It has been demonstrated theoretically that inclusions are less effective in nucleating ferrite when compared with austenite grain surfaces [62, 80]. Experiments confirm this since formation first begins at the austenite grain boundaries. Furthermore, larger inclusions are expected to be more effective since the curvature of the inclusion/nucleus interface will then be reduced; this is again generally consistent with experimental observations [53, 80].

3.5.1 The Formation of Inclusions

Deoxidation is generally accomplished by adding elements with higher affinity for oxygen than iron to the weld pool. During the solidification process, the excess oxygen is rejected from solution and combines with the alloying elements present such as manganese, silicon, aluminium, titanium, etc., to form inclusions [81].

For C-Mn steel weld metals, the analytical oxygen content normally falls within the range calculated for equilibrium with silicon and manganese at 1800 to 1900°C [82]. The deoxidation products are trapped in the weld as it cools [83]. Abson [84] mentioned that higher melting point inclusions would probably crystallise as angular shapes while the

lower melting point inclusions were likely to be more nearly spherical. Recently however, Fox and Brothers [85] stated that higher melting point inclusions which form from the liquid are spherical.

In welding processes, more than one deoxidiser is added to the weld pool and the deoxidation sequence is complicated and most of the inclusions occur in a combined form, with multiple oxides [86]. The core usually consists of a mixture of Mn, Si, and Al, oxides. A number of researchers have reported sulphide shells surrounding the inclusions [87 - 90], while others have reported titanium compounds (TiO and TiN) at the outer edge of inclusions [91 - 95].

Olson *et al* [96] reported that segregation during solidification can play an important role in inclusion formation. As mentioned previously in section 3.2.4.1, elements which have a distribution coefficients less than 1.0 will segregate to interdendritic regions. In particular elements such as oxygen and nitrogen which have very low distribution coefficients will partition to the liquid. Other elements such as manganese, silicon, titanium and sulphur will also segregate. The enrichment of these elements in the interdendritic regions may play an important role in determining the type of inclusions formed.

Ahlblom *et al* [97] used a particle analysing scanning electron microscope to characterise the types of inclusions found in steel weld deposits. These were classified into six groups:

1. MnO and MnO.SiO₂
2. Al₂O₃ and MnO + Al₂O₃ and MnO.SiO₂ + Al₂O₃
3. MnS + SiO₂
4. MnS + Al₂O₃ and MnS + SiO₂ + Al₂O₃

5. MnS
6. Others

Liu and Olson [81] used an Energy Dispersive Spectroscopy (EDS) system to determine the chemical composition of the inclusions found in the weld deposit. They found that the chemical nature of the inclusions is very complex and very large variations in manganese and silicon content were found. Using stoichiometric calculations they were able to determine the chemical formulas for over 50 types of inclusions. The 12 major types of inclusions they found in decreasing order of occurrence frequency were:

1. $\text{MnO-SiO}_2\text{-FeO-Al}_2\text{O}_3$
2. $\text{SiO}_2\text{-FeO-Al}_2\text{O}_3$
3. $\text{MnS-MnO-SiO}_2\text{-FeO-Al}_2\text{O}_3$
4. $\text{MnO-SiO}_2\text{-FeO-Al}_2\text{O}_3\text{-Al}_2\text{O}_3\text{-TiO}_2$
5. $\text{MnS-SiO}_2\text{-FeO}$
6. $\text{MnS-FeO-SiO}_2\text{-Al}_2\text{O}_3$
7. $\text{TiO}_2\text{-ZrO}_2\text{-Al}_2\text{O}_3\text{-MnO-P}_2\text{O}_5$
8. $\text{SiO}_2\text{-FeO}$
9. $\text{MnS-MnO-SiO}_2\text{-FeO}$
10. $\text{MnO-SiO}_2\text{-Al}_2\text{O}_3$
11. $\text{SiO}_2\text{-Al}_2\text{O}_3$
12. $\text{TiO}_2\text{-ZrO}_2\text{-Al}_2\text{O}_3\text{-SiO}_2\text{-MnO-P}_2\text{O}_5$

Kluken and Grong [91] modelled deoxidisation reactions in the weld metal and used this model to predict inclusion chemical compositions. They also proposed a mechanism of sequential oxidation. This sequential oxidation had originally been assumed by Bailey and

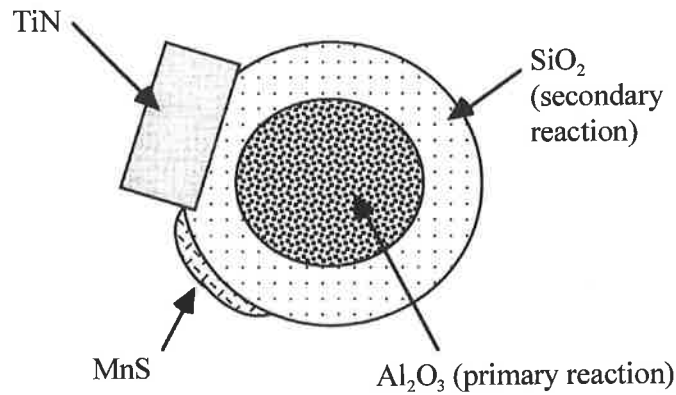


Figure 3-9: Schematic diagram of the layered structure of inclusions.

From Babu *et al* [99].

Pargarter [98]. It was assumed that the sequence of reactions occur in the following order: Al_2O_3 , Ti_2O_3 , SiO_2 , MnO , TiN , and MnS .

Babu *et al* [99] created a schematic diagram reproduced in Figure 3-9 showing a simplified model for inclusion formation based on results from Klucken and Grong [91] and Thewlis [100]. The core of the inclusion is usually made up of Al_2O_3 or Ti_2O_3 with the SiO_2 surface partially covered by MnS and TiN . The precipitation of MnS may occur in the final stages of solidification as the interdendritic liquid becomes enriched in sulphur, and TiN growth can occur in the solid state on oxide inclusions [99].

3.5.1.1 Aluminium and Titanium Oxides

Bhadeshia [53] stated that most welds contain aluminium which in general is a stronger oxide-former than titanium. Consequently, Al_2O_3 (alumina) forms first in the melt followed by TiO_2 (titania) which often grows as a thin coating on the alumina particles. Burkhardt *et al* [101] also found that Al_2O_3 and TiO_2 inclusions form first and act as nuclei

for further inclusion growth, leading to inclusion chemistries where the inclusion centre is quite different from the inclusion periphery.

Frost *et al* [102] predicted the sequence of deoxidation reactions based on the assumptions that the weld pool comes to equilibrium with the strongest deoxidant, and that the concentrations of oxygen deoxidants will be increased by microsegregation during solidification to a point where the solubility product is exceeded. Their calculations predicted that alumina inclusions will precipitate before and after solidification and that Ti_3O_5 and Ti_2O_3 inclusions will precipitate about midway through solidification.

Liao and Liu [86] reporting on information from Kiessling and Lange [103] stated that if titanium was the only deoxidiser present a number of oxides are formed depending on the oxygen content. At increasing oxygen content the following sequence of phases is expected: TiO , Ti_2O_3 , Ti_3O_5 , TiO_2 , Fe.TiO_2 , and Fe_2TiO_4 .

Fox and Brothers [85] examined inclusions using a high voltage electron microscope (HVEM) at 1MeV which was powerful enough to penetrate the inclusions and generate diffraction patterns. They identified MnO.TiO_2 (hexagonal Pyrophanite) with some evidence of $\text{MnO.Al}_2\text{O}_3$ (galaxite). From this discovery and knowing the melting points of the inclusions they proposed that galaxite first starts to solidify at 1800°C from the liquid, followed eventually by hexagonal Pyrophanite at 1360°C as the remaining liquid solidifies.

Thewlis [95] found that at constant oxygen and sulphur levels the inclusion composition changed from MnO.SiO_2 to galaxite ($\text{Al}_2\text{O}_3.\text{MnO}$) and finally Al_2O_3 as the aluminium content increased. Also, as the aluminium content increased and the inclusion type

changed there was a reduction in the number of inclusions. Increasing the titanium content resulted in extensive TiO or TiN in the surface layers of the inclusions.

Data from Thewlis [95] showed that welds containing Al_2O_3 inclusions had the lowest acicular ferrite content suggesting that Al_2O_3 inclusions raised the free energy barrier for ferrite nucleation. However, Grong *et al* [104] had showed that there was an orientation relationship between Al_2O_3 and ferrite and Mills *et al* [93] had showed that there was a low lattice misfit between Al_2O_3 and ferrite, indicating that Al_2O_3 should be an effective nucleant for acicular ferrite. In light of this knowledge, Thewlis [95] investigated further and found that the welds containing Al_2O_3 inclusions and low acicular ferrite also had high levels of soluble aluminium. Small concentrations of dissolved aluminium have been reported to promote Widmanstätten ferrite [105 - 107] which is discussed further in section 4.5.

3.5.1.2 MnO.SiO_2

Evans [108] showed that welds containing MnO.SiO_2 inclusions had low levels of acicular ferrite. However, the acicular ferrite content increased as the titanium content increased. Thewlis [95] found that welds containing MnO.SiO_2 inclusions with isolated areas of TiO or TiN on the surface was not effective in nucleating acicular ferrite whereas welds containing MnO.SiO_2 inclusions with a continuous layer of TiO or TiN were effective. Thewlis [95] indicated that it is likely that MnO.SiO_2 inclusions have a high free energy barrier to ferrite nucleation, thereby retarding acicular ferrite formation.

3.5.1.3 Manganese Sulphides

Manganese sulphides (MnS) are a common impurity in steels [53]. As mentioned in section 3.5.1 MnS shells have been found surrounding inclusions by various researchers [87 - 90, 109].

There appear to be conflicting views in the literature as to the effectiveness of MnS for nucleating acicular ferrite. Yamamoto *et al* [90] claimed that manganese sulphides, which grew on titanium oxides long after solidification, were responsible for the nucleation of acicular ferrite. However, a year later Chijiwa *et al* [110] (co-authors in Yamamoto *et al* [90]), found that a reduction in sulphur concentration tends to decrease the amount of allotriomorphic ferrite and promote the formation of acicular ferrite. Evans [111] and Devillers *et al* [87] have both reported that manganese sulphide shells inhibit acicular ferrite formation

Bhadeshia [53] reported that manganese sulphides precipitate during solid state transformation over a temperature range estimated to be 1050 - 1400°C in the solute enriched interdendritic regions of the solidification microstructure. These regions are rich in manganese and hence have a relatively low tendency to transform to ferrite. Therefore, the ability of any manganese sulphides to act as heterogeneous nucleation sites for ferrite is then reduced by the locally large concentration of austenite stabilising elements in the interdendritic regions.

3.5.2 Equations Used to Predict Inclusion Formation

Various empirical equations and ratios have been proposed in the literature to estimate the proportion or type of inclusions formed.

Widgery [112] reported that the weld inclusion volume fraction can be estimated as $5.5(\%O + \%S)$. A similar equation was put forward by Saggese *et al* [113] who estimated the weld inclusion volume fraction to be $4.9(\%O + \%S)$.

Kluken and Grong [91] proposed that the volume fraction of inclusions could be estimated by $10^{-2}[5.0(\%O) + 5.4(\%S - 0.003)]$.

Grong and Matlock [58] introduced a deoxidation parameter, $m = [\%Al]/[\%O]^2$. They found that good mechanical properties and the highest number of inclusions equated with a deoxidisation parameter close to 28.

Nakanishi and Komizo [114] report on a critical oxygen/titanium ratio of 1:1.

Saggese *et al* [113] showed graphically that inclusions containing more than 2% titanium (or TiO) were effective in nucleating acicular ferrite. From this information Abson [84] estimated that the critical ratio of titanium to (oxygen + sulphur) to give 2% titanium in the inclusions is 1:15. This ratio was confirmed by Evans [108]. Harrison [115] reviewed the data from Saggese *et al* [113] and of Bonnet and Charpentier [116] and proposed a Ti:O ratio of 1:10.

Various aluminium to oxygen (Al/O) ratios have been proposed for the formation of galaxite ($MnO \cdot Al_2O_3$) and subsequent nucleation of acicular ferrite. Grong and Matlock [58] found that as the Al/O ratio increased from 0.15 to 0.62, the proportion of acicular ferrite increased from approximately 40% to 70%. Harrison [115] found that an Al/O ratio of 0.84 appropriate for the formation of galaxite.

3.5.3 Mechanisms for the Nucleation of Acicular Ferrite from Inclusions

A variety of mechanisms have been proposed for the way in which inclusions aid the nucleation of acicular ferrite. These include the provision of inert substrates for heterogeneous nucleation [117], the formation of tessellated stresses [87], the punching out of dislocations, differential thermal contraction between the inclusions and austenite, and crystallographic matching, and chemical effects on the hardenability of the surrounding steel [118 - 120]. It is possible that more than one mechanism is operating in a particular weld.

Ricks *et al* [62] developed a model taking into consideration the free energy barrier for ferrite nucleation and the radius of inclusions. The model showed that inclusions with a diameter greater than 0.2 μm would be effective in nucleating acicular ferrite. Harrison [121] found that larger inclusions facilitated acicular ferrite nucleation better than smaller inclusions.

Devillers and co-workers [87, 122] claimed that tessellated stresses, generated around inclusions as a consequence of differences in thermal expansion coefficient, nucleate acicular ferrite. Inclusions with larger differences in thermal expansion coefficients to austenite are more favourable to nucleate acicular ferrite. Grong and Matlock [58] plotting data from Brooksbank and Andrews [123] showed that Al_2O_3 inclusions had a large difference in thermal expansion compared with austenite which imposed strain on the austenite/particle boundary. This strain was postulated to generate dislocations which in turn decreased the energy barrier to nucleate ferrite. Devillers *et al* [87] also noted that MnS inclusions have a thermal expansion coefficient similar to austenite which they claim

explains why MnS inclusions or inclusions with MnS shells do not nucleate acicular ferrite.

Liu and Olson [80] evaluated the effects of inclusion/matrix differential thermal contraction on weld metal phase transformation by using a “ball in elastic hole” model from Hirth and Lothe [124]. It was assumed that the inclusion (the ball) was a rigid sphere, incompressible and non-deformable, and all the strain is taken by the austenite (the elastic hole) during cooling. The total strain energy in the matrix was estimated by calculating the work done on the hole surface. The calculated strain energy values associated with the presence of inclusions was $\sim 0.5 \text{ J/cm}^3$. When compared with the austenite to ferrite transformation free energy change ($\sim 100 \text{ J/cm}^3$) there were several orders of magnitude difference, indicating that the strain energy term is probably insignificant in the overall reaction.

A number of researchers have found inclusions which have a low planar disregistry with ferrite provide a good crystalline substrate with good lattice matching and encourage epitaxial growth of ferrite [84]. Inclusions found to have a low planar disregistry include TiO [93, 94, 111, 125] and galaxite ($\text{MnO} \cdot \text{Al}_2\text{O}_3$) [93, 94]. Other researchers have discounted this mechanism as being unlikely [62, 92].

Gregg and Bhadeshia used pressure bond experiments [118, 119] and molten steel inoculation experiments [120] to study the chemical interaction between inclusions and surrounding weld metal. They found that Ti_2O_3 inclusions act as a sink for manganese and hence causes a reduction in manganese levels in the surrounding steel. Manganese is known to retard the formation of ferrite from austenite (see section 4.4), and so depletion in the steel has the opposite effect.

In their molten steel inoculation experiments, Gregg and Bhadeshia [120] implied that at elevated temperatures TiO_2 releases oxygen which is diffused into the steel and reacts with carbon. The reaction with carbon causes local carbon depletion around the inclusion which stabilises ferrite over austenite, thus accelerating transformation locally and producing nucleation.

4. THE EFFECT OF ALLOYING ELEMENTS

4.1 THE ROLE OF OXYGEN AND NITROGEN

As discussed in section 3.5, oxygen combines with aluminium, titanium, silicon, and manganese, as well as some rare earth (RE) elements to form non-metallic inclusions. It has also been discussed in section 3.4.3 that these inclusions nucleate acicular ferrite. For this reason oxygen is a vital alloying addition to C-Mn steel weld deposits and its role in nucleating acicular ferrite has been investigated by many researchers (for example: [126 - 129]).

Considerable effort has been directed towards understanding the effect oxide inclusions have on mechanical properties, in particular Charpy toughness. With upper shelf toughness, the fracture mode is micro-void coalescence. Increasing oxygen content also increases the volume fraction of inclusions which assist crack propagation through micro-void coalescence. For this reason increasing the oxygen content is detrimental to upper shelf Charpy toughness [130 - 132].

Tweed and Knott [133] observed that at low temperatures cleavage fracture is initiated at large inclusions. However, cleavage resistance also depends on microstructure because toughness does not continue to improve as the oxygen content decreases [63, 134]. According to Ito and Nakanishi [63] acicular ferrite may fail to form in weld deposits with less than 200 ppm oxygen. Since acicular ferrite has higher toughness than other micro-constituents, a weld oxygen content greater than 200 ppm is normally aimed for.

Numerous studies have found that optimum toughness is achieved for oxygen contents between 250 and 350 ppm [34, 63, 135, 136].

Nitrogen is not usually a deliberate alloying addition in most low alloy steel weld deposits. It is picked up from the environment and from impurities in the consumables used in the welding process. Where the weld is diluted by the parent plate, the composition of the plate must also influence the nitrogen concentration of the weld [53]. Variations in arc length during welding can easily lead to substantial variations in the weld nitrogen content. Nitrogen is known to have a potent detrimental effect on the toughness of the weld. The mechanism of embrittlement is believed to be associated with strain ageing and solid-solution hardening effects, both of which increase the yield stress of the weld, and consequently cause a decrease in toughness [8, 137].

Mori *et al* [125] indicated that nitrogen has no detectable influence on the acicular ferrite content of welds, whereas Ito and Nakanishi [63] indicated alterations in microstructure as a function of the nitrogen concentration.

Hori *et al* [138] and Lau *et al* [139] studied titanium/boron/nitrogen phenomena in submerged arc welds. They found that nitrogen is not expected to influence the development of microstructure in the absence of boron additions. Boron was added with the intention of improving hardenability and enhancing the opportunity for the austenite to transform to acicular ferrite. Titanium is usually added in order to protect the boron from oxidation during the transfer across the arc.

4.2 CARBON

The effect of carbon is mainly to limit the width of the coarse-grained allotriomorphic ferrite, formed at the prior austenite grain boundaries, and in influencing the rate of Widmanstätten ferrite formation. During the transformation from austenite to ferrite, the carbon atoms diffuse into the remaining austenite and the growth (or thickening) rate of the allotriomorphic ferrite is controlled by the diffusion rate of carbon in austenite. A higher carbon content gives a slower growth rate of the ferrite at the prior austenite grain boundaries [79].

Increasing carbon content leads to lower contents of both allotriomorphic ferrite and Widmanstätten ferrite, giving room for increasing contents of fine grained acicular ferrite. However, it is not known whether the actual growth rate of acicular ferrite is influenced by the carbon content [79].

Carbon has a marked strengthening effect in as-deposited metal, greater than that of most other elements. Dorsch and Stout [140] performed welds on high purity iron and introduced increasing levels of carbon. Their results showed that carbon dramatically increased the weld hardness.

Evans [141] reported that carbon reduced the prior austenite grain width and had a beneficial effect on the acicular ferrite content. However, it also increased the volume fraction of grain boundary carbide films which offset any improvements made in the acicular ferrite content or grain size. Abson [142] found that an increase in carbon was also accompanied by an increase in the proportion of microphases including retained austenite.

Svensson and Gretoft [79] produced welds at four different carbon levels ranging from 0.03 to 0.12% . They found that for a given manganese level, increasing the carbon content increased the yield strength. Also, at higher manganese levels (1.6 - 1.8%), welds containing intermediate levels of carbon (0.06 - 0.09%) had the best Charpy toughness results.

After reviewing various studies into the effect of carbon Abson and Pargeter [143] concluded that for satisfactory Charpy toughness it is advisable to maintain carbon levels at least as low as 0.10% and possibly lower.

4.3 SILICON

Various researchers have reported that silicon strengthens the weld. John *et al* [144] claimed that silicon was the principal strengthening element in a series of manual metal arc welding trials that they conducted. Evans [145] found that the magnitude of the strengthening effect of silicon decreased as the silicon content increased.

High levels of silicon are considered detrimental to weld toughness [143]. However, a continuous lowering of the silicon content does not necessarily give a continuous improvement in toughness. The optimum weld metal silicon content usually lies in the range 0.3 - 0.5%. Ito and Nakanishi [63] reported that the optimum silicon content was between 0.15 and 0.25%. Hannerz [146, 147] found that a weld silicon content in the range 0.25 to 0.35% gave the best Charpy toughness results. Widgery [148] reported good mechanical properties from welds containing 0.58% silicon. Generally the oxygen content decreases as silicon content increases [63, 142, 145, 149]. Abson [142] reported a slight

deterioration in Charpy toughness values in spite of a slight increase in acicular ferrite content as the weld silicon content increased.

Hill and Levine [150] found that in high heat input welding (3.5 kJ/mm), increasing the silicon content from 0.24% to 0.51% increased the acicular ferrite content which corresponded with an improvement in Charpy toughness values.

A few researchers have noticed some refinement in the microstructure as the silicon content increases [142, 149]. Abson [142] found that adding silicon lowered the transformation temperature which would account for the observed refinement in microstructure.

Evans [145] varied the silicon content in manual metal arc welds from 0.2 to 0.9% at four different manganese levels (0.6, 1.0, 1.4, and 1.8% manganese). He found that at low manganese levels the proportion of acicular ferrite increased as silicon increased, but at higher manganese levels the effect was less pronounced. The Charpy toughness decreased as the silicon content increased with the upper shelf being depressed and the transition curves being displaced to higher temperatures. Evans also reports that even though the amount of acicular ferrite did not change the acicularity increased with concurrent increase in the amount of dark etching microphase.

Contrary to Abson [142], Evans did not observe any refinement of the acicular ferrite as the silicon content increased. He did comment however, that the acicular ferrite aspect ratio increased and there was more retained austenite as the silicon content increased.

4.4 MANGANESE

The major role of manganese in the weld metal transformation kinetics is to restrict the growth of grain boundary ferrite allotriomorphs (and, indirectly Widmanstätten ferrite) until the degree of undercooling reaches the point where intragranular nucleation of acicular ferrite is energetically feasible [79, 151].

The addition of manganese significantly lowers the austenite to ferrite transition temperature [121, 152 - 154]. This generally leads to a refinement in the acicular ferrite lath size.

Evans [155] conducted a systematic study into the effect of manganese on weld microstructure. His results showed that manganese increased the proportion of acicular ferrite at the expense of allotriomorphic ferrite and that the acicular ferrite lath width decreased from 3.3 μm to 1.05 μm as the manganese content increased from 0.66 to 1.82%. Similar findings were reported by Taylor [156]. Widgery [157] reported that increasing manganese refined the weld microstructure. Choi and Hill [158] found that at 0.8% manganese the predominant microstructure was Widmanstätten ferrite, at 1.4% manganese the microstructure was mainly acicular ferrite, and at 2.2% manganese the microstructure was mainly bainite. Grong *et al* [151] found that the acicular ferrite content increased with the manganese level and peaked at about 2.2% manganese. Above this value the increased hardenability resulted in the formation of bainite and martensite.

Evans [155] also found that increasing manganese content decreased the upper shelf of Charpy transition curves. In contrast, manganese had a beneficial influence on the lower shelf due to the effect on cleavage resistance. The improved lower shelf Charpy toughness

properties at higher manganese levels were probably a result of the refined microstructure. In the transition region of the Charpy impact curves Evans [155] found that manganese had an optimum influence at 1.5%, despite the progressive improvement in microstructure throughout. Grong *et al* [151] showed that increased additions of manganese (up to a certain level) have a beneficial effect on notch toughness by both enhancing the upper shelf energy and displacing the transition to lower temperatures. At lower testing temperatures Grong *et al* [151] observed the maximum absorbed energy at 1.6% manganese.

Bosward and John [159] studied Charpy toughness in welds in which the manganese content varied from 0.44% to 1.6% and found that the toughness increased progressively with increasing manganese. Dorsch and Stout [140] found that with high heat input welding the Charpy toughness changed little up to 1.6% manganese and then dropped rapidly in welds containing more manganese.

Evans [155] also reported that the yield strength of weld deposits increased with manganese levels and that the amount of retained austenite also increased.

4.5 ALUMINIUM

As mentioned in section 4.1, aluminium is a strong deoxidiser and therefore combines with oxygen to form non-metallic inclusions. For this reason aluminium is a very important micro-alloying element.

It was noted by Thewlis [95], Evans [160], and Abson [161] that increasing the aluminium content changed the composition of the inclusions found in the weld deposit. At high aluminium levels Al_2O_3 inclusions formed. When the aluminium content increased to levels equivalent or greater than stoichiometry with oxygen and Al_2O_3 inclusions formed,

the proportion of Widmanstätten ferrite increased markedly at the expense of acicular ferrite. Terashima and Hart [162], Grong *et al* [106], and Abson [105], have also shown that an increase in soluble aluminium content leads to an increase in the Widmanstätten ferrite content. Work by Mabuchi and Nako [163, 164] showed that soluble aluminium retarded ferrite formation but promoted bainite formation.

Brownlee *et al* [107] found that increasing the aluminium concentration above 250 ppm caused the primary intergranular transformation product to change from acicular ferrite to ferrite with aligned carbide (FAC).

Liao and Liu [86] found that below 550 ppm aluminium the volume fraction of acicular ferrite and bainite increased with aluminium while the volume fraction of allotriomorphic ferrite decreased. Above this point (550 ppm aluminium) the volume fraction of allotriomorphic ferrite and bainite increased at the expense of acicular ferrite.

Thewlis [95] suggested that large substitutional atoms of aluminium may build up in the austenite at the austenite/ferrite interface and act as kinetic barrier restricting carbon distribution which would retard ferrite formation. This would reduce the transformation temperature and promote bainite formation.

Evans [160] studied the influence of aluminium on C-Mn steel weld deposits in the range <5 - 600 ppm. The lowest Charpy impact transition temperature was obtained for the aluminium free weld (<5 ppm aluminium). It was also observed that Charpy toughness properties degraded as the aluminium content increased with a small recovery in toughness encountered at an aluminium content of 350 ppm which coincided with an Al/O ratio of 0.8 (see section 3.5.2).

In another study, Evans [165] looked at the effect of aluminium on the microstructure and mechanical properties C-Mn weld deposits containing titanium and boron at varying levels of nitrogen. The results showed that aluminium contents up to 170 ppm led to the Charpy toughness being degraded by nitrogen, whereas amounts in excess of this responded beneficially to the addition of nitrogen.

Thewlis [100, 166] reported that in welds containing titanium and boron, optimum Charpy toughness and acicular ferrite content was achieved at an aluminium content of 0.028%. Poor toughness at higher aluminium levels was attributed to increased proportions of allotriomorphic ferrite and Widmanstätten ferrite.

Terashima and Hart [167] found that the tolerance to high weld metal aluminium contents was dependent upon the oxygen content. Welds with an oxygen content of ~ 200 ppm were able to only tolerate ~ 240 ppm aluminium, whereas welds containing increased levels of oxygen (~ 300 ppm) could accept up to ~ 400 ppm aluminium.

4.6 TITANIUM

The introduction of a very small amount of titanium can totally modify the weld microstructure. Evans [168] reports a ten-fold increase in the volume fraction of acicular ferrite on increasing titanium up to 30 ppm.

Titanium combines with oxygen and nitrogen to form TiO , TiO_2 , and TiN compounds. TiO , $\text{MnO} \cdot \text{TiO}_2$, and TiN compounds have been found on the surface of other inclusions such as $\text{MnO} \cdot \text{SiO}_2$ [95] and Galaxite [85, 95]. Thewlis [95] has attributed the nucleation of acicular ferrite to the titanium oxide coatings found on some inclusions.

The other role titanium plays in weld metal transformation is to protect boron from nitrogen and oxygen and to reduce the level of free nitrogen in the weld. Ohkita *et al* [169] developed a model where during solidification titanium reacted with oxygen to form titanium oxides leaving boron to react with nitrogen to form BN. During cooling in the austenite range, they claim that titanium further protects boron by preferentially reacting with nitrogen to form TiN, and free boron segregates to the austenite grain boundaries where it helps suppress ferrite formation (see section 4.8).

Johnson *et al* [170] studied the effect of titanium on as deposited thermally cycled weld metal microstructure. They reported that the welds with the best toughness contained 30 and 240 ppm titanium while welds containing 0 and 100 ppm titanium had low toughness.

Oh *et al* [171] studied the effects of increasing the weld titanium content up to 728 ppm at five different boron levels (6 to 91 ppm). They found increasing the titanium level refined the microstructure and also found that the amount of titanium required to maximise the volume fraction of acicular ferrite drops as the boron content increases. They also reported that excessive amounts of boron and titanium are detrimental to weld metal toughness by increasing the volume fraction of upper bainite. Thus, boron and titanium contents must be controlled to produce the desired microstructure.

Liao and Liu [86] reported a threshold concentration of titanium (approximately 500 ppm) above which the volume fraction of bainite increased at the expense of acicular ferrite.

Sanchez-Osio *et al* [172] found that titanium increased the recovery of manganese and silicon from electrode coatings. They found an optimum titanium content of 300 ppm (and 15 ppm boron) which maximised the volume fraction of acicular ferrite and minimised the

weld hardness. With increasing amounts of titanium and boron, above 350 ppm and 30 ppm respectively, they found that weld hardenability becomes excessive and the formation of martensite and microconstituents is promoted.

Evans [108] found that even though the proportion of acicular ferrite in the weld metal remained high as the titanium content increased, toughness peaked at a titanium content of 30 ppm. Bonnet and Charpentier [116] found that at least 45 ppm titanium is required to produce a large volume fraction of acicular ferrite. Finally, titanium is only effective provided there is sufficient alloying with manganese [173].

4.7 SULPHUR AND PHOSPHORUS

Evans [111] conducted a systematic study of the influence of sulphur and phosphorus on weld metal microstructure and mechanical properties. Sulphur levels were varied between 0.007 and 0.046% with the phosphorus level constant at 0.007%. The results showed that sulphur increased the volume fraction of Widmanstätten ferrite at the expense of acicular ferrite which in turn affected the Charpy toughness results. The upper shelf absorbed energies were reduced and the ductile to brittle transition temperature increased as the sulphur content increased. Examination of the inclusions revealed that MnO.SiO_2 was found in the centre and the surface layer varied with the sulphur content. At low sulphur levels titanium compounds, presumably TiO , were found in the outer layer. However, as the sulphur content increased the surface layer was found to be MnS . Fox and Brothers [85] and Thewlis [95] have since shown that inclusions with titanium compounds at the surface nucleate acicular ferrite. The change in the inclusion surface composition reduces the number of nucleation sites for acicular ferrite and as a result greater amounts of Widmanstätten ferrite are formed.

In the same study [111], Evans also investigated the influence of phosphorus. Phosphorus levels were varied between 0.007 and 0.040% with sulphur at 0.005%. The results showed that there was only a marginal effect on microstructure as the phosphorus content increased. Similarly there was little effect on Charpy toughness values and only a slight increase in hardness and tensile properties.

From this detailed study it is evident that sulphur is detrimental to weld microstructure and mechanical properties due to its tendency to form manganese sulphide layers around inclusions. Phosphorus, on the other hand, acts principally as a solid solution strengthener and can be tolerated in higher concentrations. Evans [111] found little benefit in reducing sulphur or phosphorus levels below 0.005%.

4.8 BORON

With balanced additions of boron and titanium the formation of grain boundary ferrite can nearly be eliminated [135, 174 - 176]. Free boron is known to segregate extensively to the prior austenite grain boundaries where it reduces the grain boundary energy [172, 177]. By reducing the grain boundary energy the surface free energy ratio is also lowered. This in turn increases the energy barrier to nucleation [178].

Thewlis [95] studied the influence of soluble boron on weld microstructure and found that boron decreased the austenite to ferrite transformation temperature. The effect of boron on the formation of allotriomorphic ferrite was unclear because of other variables. However, Thewlis concluded that soluble boron acts to reduce allotriomorphic ferrite formation and promote acicular ferrite.

5. EXPERIMENTAL

5.1 BASE MATERIAL SELECTION

The parent material chosen for experimental welding was 12 mm AS 3678-250 (grade 250) structural steel [179] supplied by BHP. Grade 250 steel has a minimum specified tensile strength of 260 MPa and is commonly used in the metal fabrication industry. The chemical composition limits for grade 250 along with the actual base material chemical analysis is shown in Table 5-1. Chemical analysis of the grade 250 plate used for the welding showed that all the elements and the carbon equivalent were within the limits specified by AS 3678 [179].

The microstructure of grade 250 is shown in Figure 5-1. The microstructure consists of grains of ferrite (light etching) and pearlite (dark etching).

	C %	Mn %	Si %	S %	P %	Ni %	Cr %	Cu %	Ti %	Al %	C.E. (2)
Min	*	*	*	*	*	*	*	*	*	*	
Max	0.22	1.70	0.55	0.040	0.040	(1)	(1)	(1)	0.040	0.100	0.45
Actual	0.16	0.68	0.01	0.012	0.019	0.03	0.02	<.01	<.01	0.05	0.28

* = no specified limit

NOTES

(1) The following elements may be present to the limits stated, subject to a total content of 1.00%.

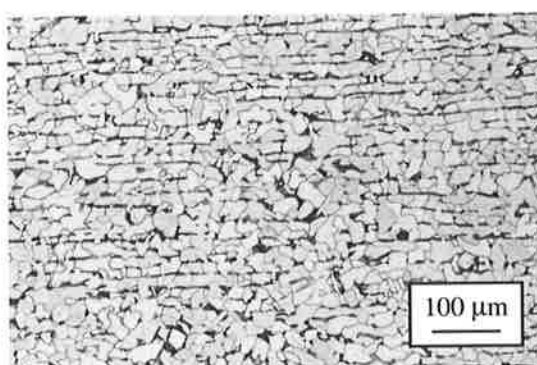
- (a) Copper 0.40%
- (b) Nickel 0.50%
- (c) Chromium 0.30%
- (d) Molybdenum 0.10%

(2) Carbon equivalent (C.E.) is calculated from the following equation:

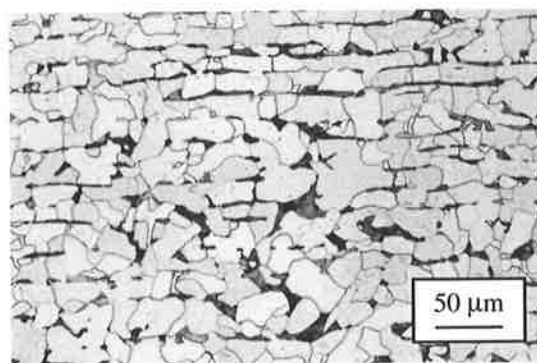
$$\text{C.E.} = \text{C} + \frac{\text{Mn}}{6} + \frac{\text{Cr} + \text{Mo} + \text{V}}{5} + \frac{\text{Ni} + \text{Cu}}{15}$$

Table 5-1: Chemical composition limits and analysis of Grade 250 steel.

From AS 3678 [179]



(a) low magnification



(b) higher magnification

Figure 5-1: Micrographs of grade 250 steel base material (etched in 2% nital).

5.2 METAL CORE WIRE

Metal core wire is a welding wire that consists of solid metal outer casing and a metal powder filling. The chemical composition of the wire can be varied by altering the composition of the metal powder.

A cross section through a metal core filler wire is shown in Figure 5-2. The dark area in the centre of the wire is the metal powder. The steel casing overlaps to prevent moisture penetrating through to the powder. The wire consists of 80% metal casing and 20% metal powder by weight.

5.2.1 Metal Core Wire Manufacture

A schematic diagram of the metal core wire manufacturing process is shown in Figure 5-3. Metal core wire is manufactured by first forming a 14 mm low carbon metal strip into a “U” shaped channel and filling it with metal powder. The channel then passes through a series of closing rollers and reducing rollers to reach a diameter of 2.5 mm.

Following rolling the wire is drawn through a series of dies until it has reached a final diameter of 1 mm. After drawing the wire is wiped using a cloth and baked at 350°C in air for 1 hour to remove drawing lubricant.

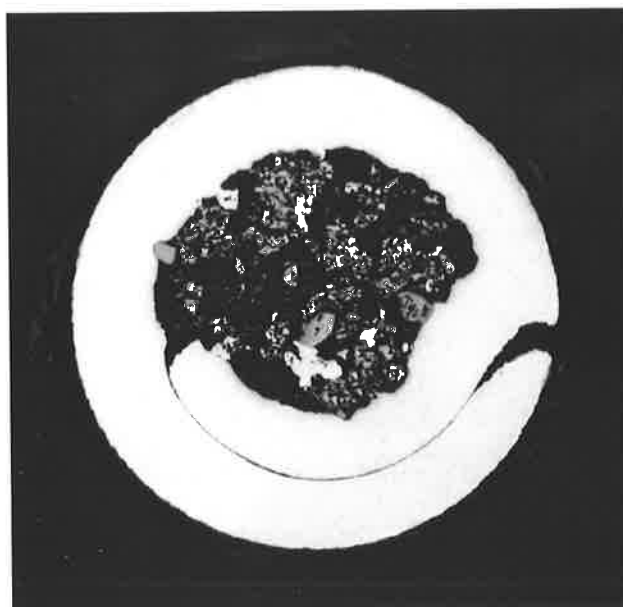


Figure 5-2: Cross section of a metal core filler wire.

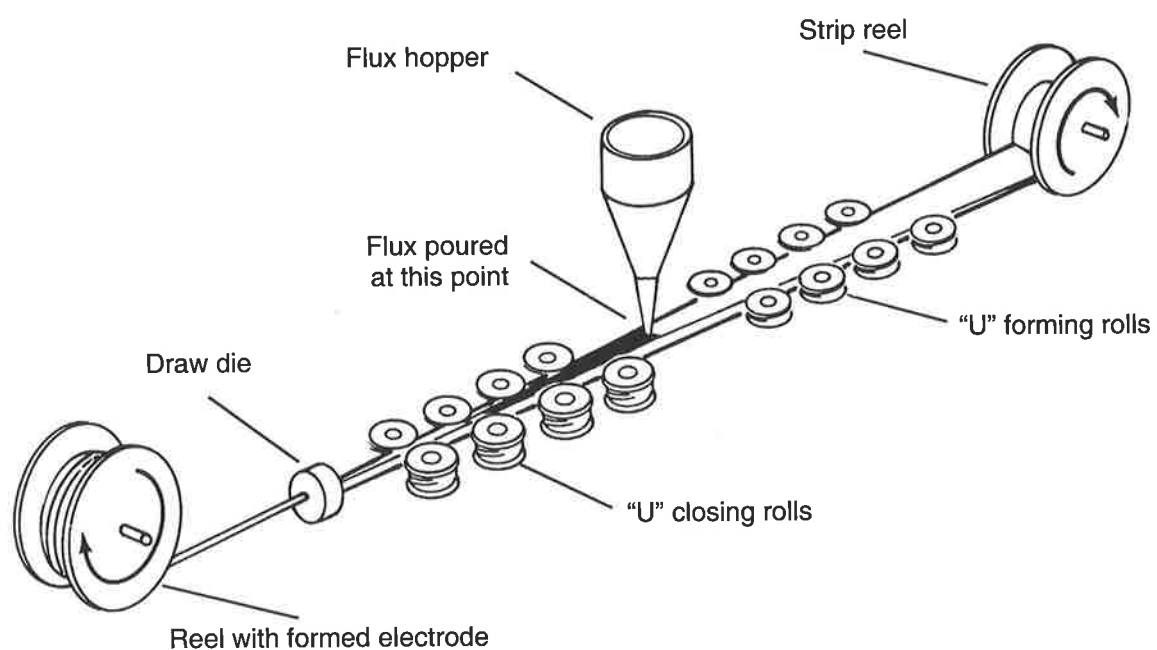


Figure 5-3: Schematic diagram of the metal core wire manufacturing process.

From Jackson [180].

5.3 METAL CORE WIRE COMPOSITION

Experiments at CSIRO found a metal core wire which showed promising results. This wire was used as the basis for the experimental wires. To determine a metal core wire chemical composition which would lead to a desired weld chemical composition a mass balance was conducted on a test plate welded with this wire using welding conditions similar to those to be used in the experiment. Details of the mass balance for silicon and manganese are contained in Appendix A.

Two series of wires were produced. Series 1 wires had different levels of silicon while maintaining a constant manganese content, and series 2 wires had different levels of manganese while maintaining a constant silicon content.

Table 5-2 lists the chemical compositions of the metal core wire ingredients. By altering the metal powder ratios, it was possible to change the metal core wire chemical composition. In series 1 wires the percentage of manganese powder was held constant and the ratios of iron powder and ferrosilicon powder were varied to produce wires of varying silicon contents. In series 2 wires the percentage of ferrosilicon powder was held constant and the ratios of iron powder and manganese powder were varied to produce wires of varying manganese contents.

Each series consisted of four wires. One wire from series 1 (wire 1-2) was designed to have the same silicon content as the wires in series 2. This allowed the results from wire 1-2 to be used in both series of wires.

	C wt%	Mn wt%	Si wt%	Al wt%	Fe
Iron powder (Fe)	0.01	0.02	0.07	-	rem
Manganese powder (Mn)	-	99.5	-	-	
Ferrosilicon powder (Fe/Si)	0.02	-	46.0	1.0	rem
Steel casing	0.05	0.22	0.01	0.015	rem

Table 5-2: Experimental wire ingredients and their chemical composition.

The calculations used to determine the metal powder ratios required to give the desired weld chemical compositions is contained in Appendix A. Table 5-3 lists the metal powder ratios calculated to give the required weld silicon contents for series 1 metal core wires. Table 5-4 lists the metal powder ratios calculated to give the required weld manganese content for series 2 metal core wires (note the inclusion of wire 1-2).

Wire	Predicted Weld	Predicted Wire	Metal powder ratios		
Number	Silicon Content (%)	Silicon content (%)	Fe/Si (χ)	Fe ($0.75 - \chi$)	Mn
1-1	Minimum	0.0	0.00	0.75	0.25
1-2	.25	0.89	0.07	0.68	0.25
1-3	.30	1.07	0.10	0.65	0.25
1-4	.40	1.42	0.15	0.60	0.25

Table 5-3: Metal powder ratios for series 1 metal core wires (varying silicon).

Wire	Predicted Weld	Predicted Wire	Metal powder ratios		
Number	Manganese Content (%)	Manganese content (%)	Fe/Si	Fe (0.93 - χ)	Mn (χ)
2-1	Minimum	0.17	0.07	0.93	0.00
2-2	0.90	2.44	0.07	0.81	0.12
2-3	1.10	3.62	0.07	0.75	0.18
1-2	1.30	4.80	0.07	0.68	0.25
2-4	1.50	6.00	0.07	0.62	0.31

Table 5-4: Metal powder ratios for series 2 metal core wires (varying manganese).

5.4 TEST PLATE PREPARATION

The 12 mm steel plate was cut using oxy-acetylene equipment into plates measuring 200 mm x 800 mm. One long edge was milled to remove material thermally damaged during the cutting process and to produce a square edge. A 2.5 mm gap was placed between adjoining plates and was maintained by lengths of 2.5 mm diameter solid wire at the base. Small tabs were welded on each end to restrain the joint during welding and to act as run-on and run-off plates. The joint configuration is detailed in Figure 5-4.

Immediately prior to welding the joint was grit blasted and the top surface was dressed with an angle grinder to remove any surface oxide.

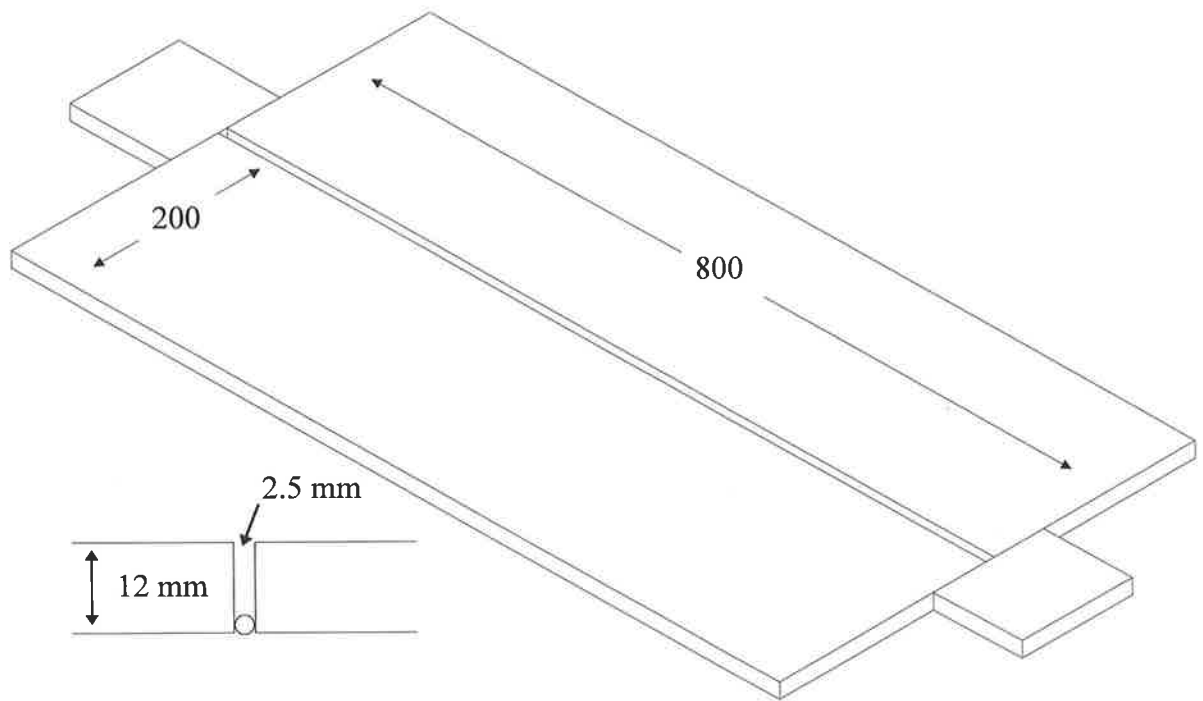


Figure 5-4: Test Plate set up and joint configuration prior to welding.

5.5 WELDING PROCEDURE

Figure 5-5 shows a schematic of the welding process. The HC-GTAW torch (described earlier in section 2.4) was mounted on a beam and carriage mechanism which controlled the horizontal travel. The arc length controller (Jetline model ALC 201) monitored the voltage and adjusted the vertical position of the torch. As mentioned in section 2.3, the vertical position of the electrode is crucial for maintaining process stability. The power supply used was a Strata STMP 650 inverter which had a maximum current of 650 Amps.

The entry position of the filler wire was crucial. If the wire is fed in too far away from the electrode it would stub in the solidified weld pool, if it was too close the wire would arc against the electrode. To overcome this problem an adjustable wire feeder was constructed

which allowed the wire feed position to be adjusted during welding. The wire was fed at a constant rate during welding. The return lead was attached to the test plate via a bolt which was welded to the run-on plate.

Welding was performed in the downhand position from one side only using the settings listed in Table 5-5. Due to the high arc forces and low voltage the arc length controller placed the electrode tip approximately 1.5 mm below the plate surface. The arc was started

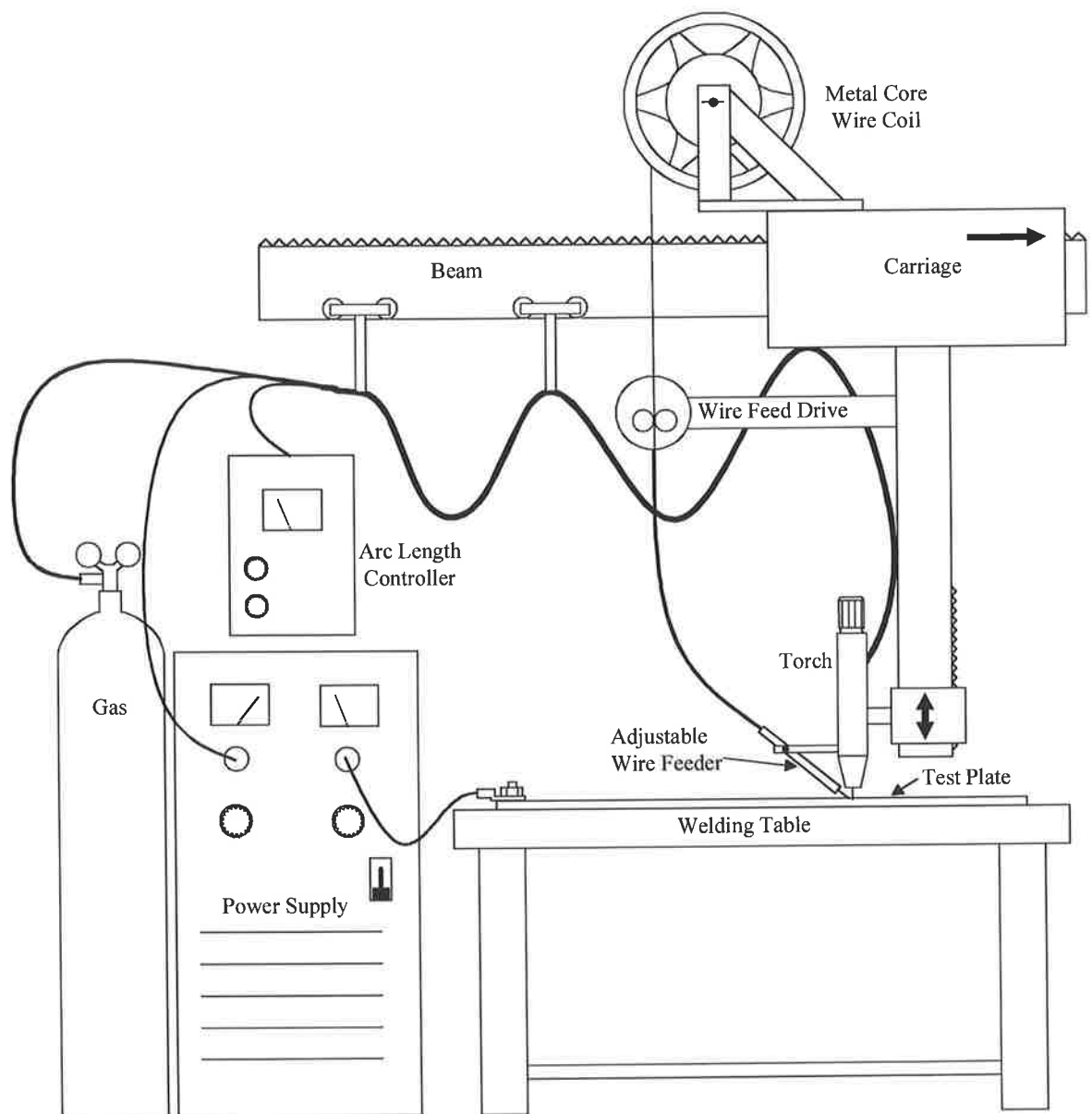


Figure 5-5: Schematic diagram of the experimental welding system.

Electrode	3.2 mm Tungsten with 1.5% La ₂ O ₃
Electrode Angle	45°
Shielding Gas	Welding Grade Argon - 25 l/min
Current	530 Amps
Voltage	11.5 V
Travel Speed	150 mm/min
Wire Feed Rate	2.5 m/min

Table 5-5: Welding parameters.

on the run-on plate using a high voltage with low current. As the arc stabilised the current was raised and the voltage lowered. Once optimum settings were reached metal cored wire was introduced to the rear of the weld pool.

Figure 5-6 shows a captured video image of welding in progress. The image was recorded by using a pulsed nitrogen laser which illuminated the weld area, synchronised with a variable shutter speed video camera.

In most cases the single sided welding procedure did not fully penetrate the 12 mm base material and a groove remained on the reverse side. As a result of this groove it was not possible to perform radiography to check for internal defects.



Figure 5-6: Laser strobe image of HC-GTA welding in progress.

6. TESTING PROCEDURES

6.1 MECHANICAL TESTING

6.1.1 Hardness Testing

Diamond pyramid hardness testing was performed on the weld deposit using a Vickers square base pyramid diamond with face angles of 136° . Hardness testing was conducted in accordance with Australian Standard AS 1817 [181] using a 5 kg load. Specimens for hardness testing were prepared in a similar manner to that for metallographic examination. The surface was polished to a $1\mu\text{m}$ diamond finish and lightly etched in 2% nital to reveal the weld zone. Prior to testing the welded specimens the hardness testing machine was calibrated using a 195 HV5 standard hardness test block.

Microhardness testing was carried out identify phases in the weld microstructure. The machine used was a Leco M400 H2 microhardness tester fitted with a Vickers diamond pyramid indenter and the load applied was 25 grams.

6.1.2 Charpy Testing

Charpy V Notch (CVN) testing consists of measuring the energy absorbed in breaking a pre-notched weld sample at a specific temperature using a pendulum mounted striker. CVN testing is generally carried out over a range of temperatures to determine the ductile to brittle transition temperature. However, due to a limited amount of weld material a single test temperature was chosen. Three sub-size CVN samples with the dimensions 5 x 10 x 55 mm, were prepared from each test plate. A 2 mm V notch with a 45° angle was

machined through the centre of the weld using a purpose built broach. Figure 6-1 shows details of the Charpy sample dimensions and Figure 6-2 shows a schematic diagram detailing the extraction of the Charpy samples from the welded test plate. Testing was carried out in accordance with Australian Standard AS 1544.2 [182] and specimens were tested at 0°C.

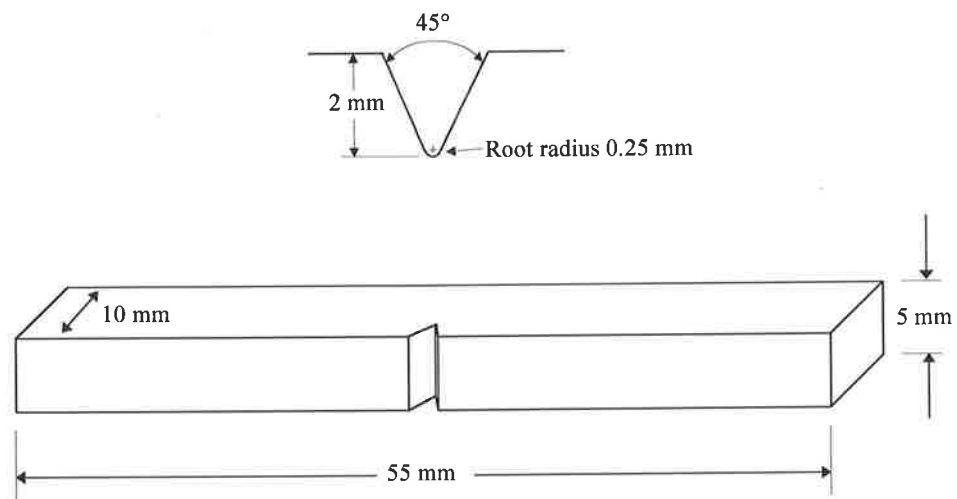


Figure 6-1: Charpy V-notch sample dimensions and notch detail.

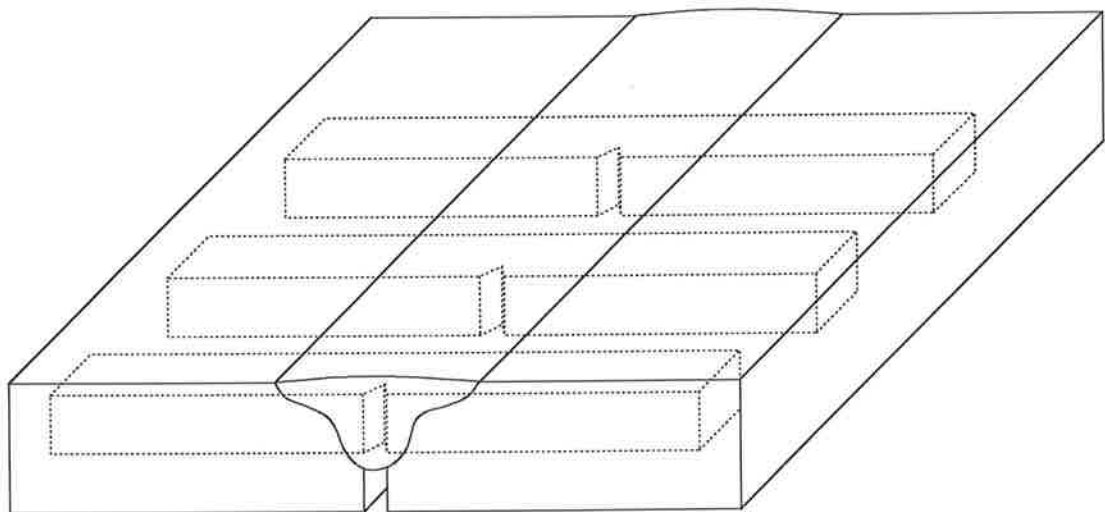


Figure 6-2: Extraction of CVN samples from the welded joint.

Following Charpy testing, the fracture surfaces of selected broken samples were examined by scanning electron microscopy and light microscopy. A Cambridge Stereoscan 250 SEM was used to examine the fracture surface and determine the failure mode (cleavage or microvoid coalescence). After SEM the broken Charpy samples were sectioned as indicated in Figure 6-3, mounted in bakelite, polished, and etched for light microscopy.

6.1.3 Tensile Testing

All weld metal tensile testing was carried out to determine the influence of alloy additions on the strength of the weld deposit. Due to the small size of the weld deposit the tensile sample dimensions were based on Hounsfield No. 12 round test pieces [183]. Figure 6-4

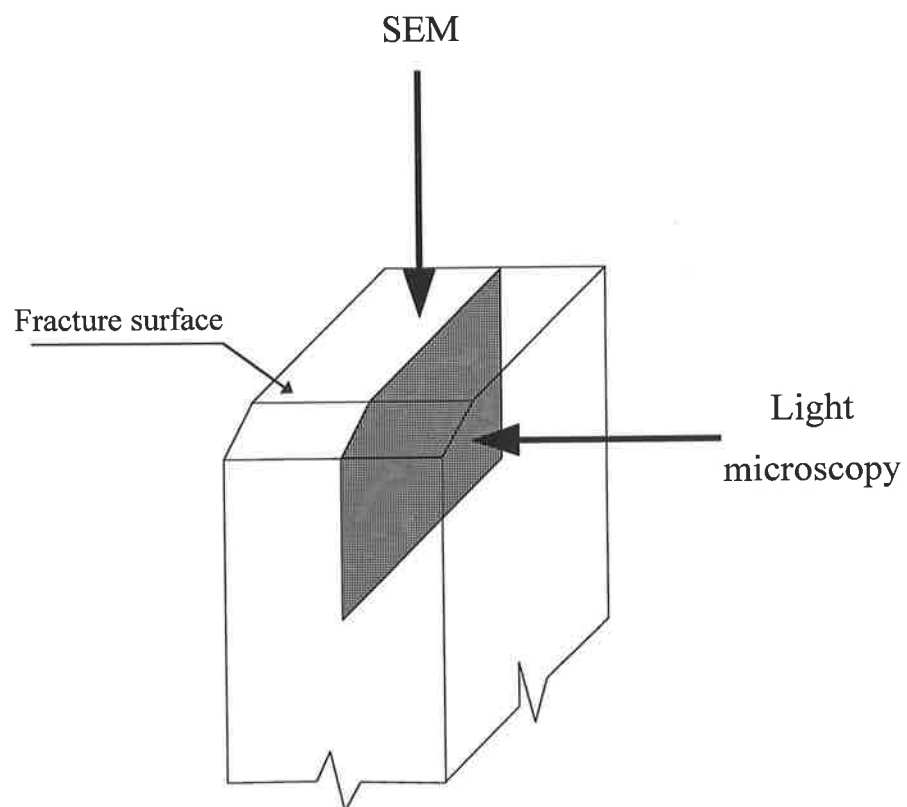


Figure 6-3: Schematic diagram showing the examination procedure for broken Charpy samples.

shows the dimensions of the tensile samples used and Figure 6-5 details how the tensile samples were extracted from the test plates.

Tensile testing was carried out in an MTS model 810 servo hydraulic universal testing machine using a 12.5 kN load cell. An adaptor bar was machined to allow the use of Hounsfield chucks in the MTS grips. The tensile testing was monitored by creating a data file which logged load and displacement. Due to the small size of the samples it was not

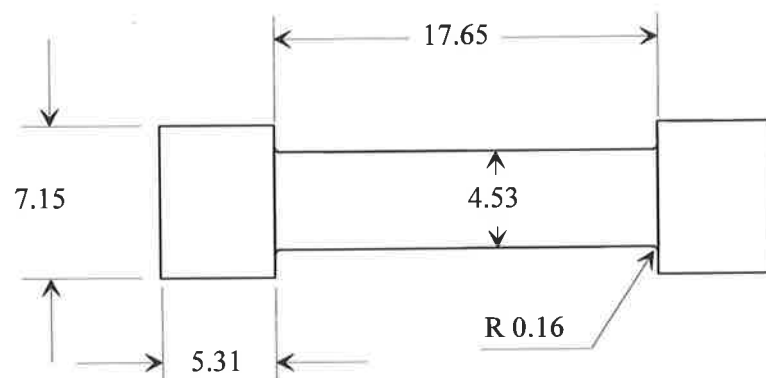


Figure 6-4: Tensile sample dimensions.

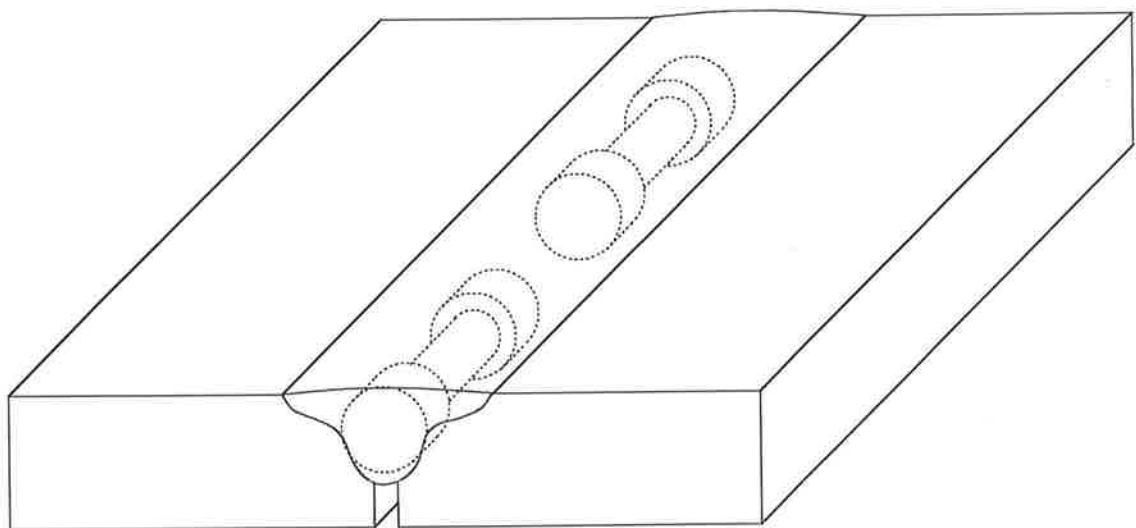


Figure 6-5: Extraction of tensile samples from the welded joint.

possible to attach an extensometer to the gauge length for measuring strain. However, an estimate of the strain was determined by dividing the displacement by the gauge length.

Following tensile testing the fracture surface of selected broken samples was examined in a Cambridge Stereoscan 250 SEM.

6.2 ANALYTICAL PROCEDURES

6.2.1 Chemical Analysis

The chemical analysis of the metal core wires was determined by the Inductively Coupled Plasma Atomic Emission Spectroscopy (ICP-AES) method at an independent laboratory. ICP-AES was chosen for its ability to accurately analyse multiple elements at the parts per million level. Oxygen and Nitrogen content was determined using a LECO TC136 analyser. A diagram explaining the extraction of samples for chemical analysis is shown in Figure 6-6.

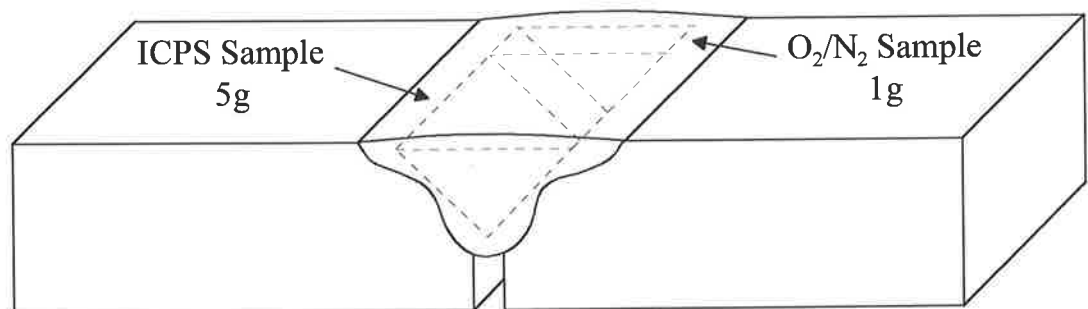


Figure 6-6: Schematic diagram showing the extraction of samples for chemical analysis.

6.2.2 X Ray Powder Diffraction

X Ray Powder Diffraction (XRPD) is a powerful tool used for determining the chemical composition of inclusions found in the weld deposit. The basic powder experiment uses a powder sample placed in a collimated monochromatic beam.

Samples were prepared for XRPD by cutting a triangular wedge of weld deposit approximately 65 mm long from the test plate as shown in Figure 6-7. Each section of weld metal was then suspended in a solution of 10% hydrochloric acid and methanol with a constant current density of 1.5 mA/cm^2 passed through it to dissolve the metal matrix as shown in Figure 6-8. After 7 days the solution is filtered through a $0.22\mu\text{m}$ filter and the filtrate is submitted to XRPD examination.

A schematic diagram of the apparatus for performing XRPD is shown in Figure 6-9. A thin layer of powder is spread on the sample holder and placed in the diffractometer. An x-ray beam of a known wavelength is directed towards the sample. The sample rotates

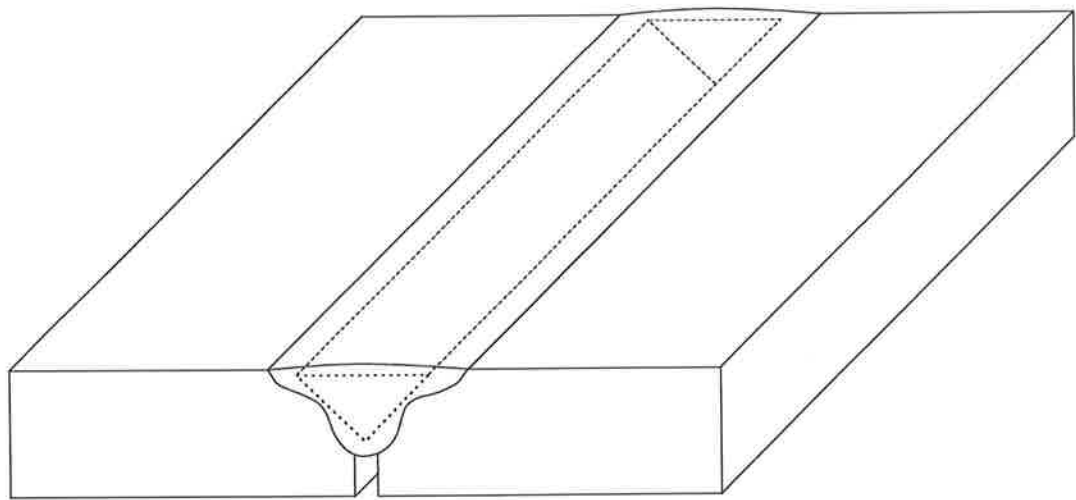


Figure 6-7: Schematic diagram showing XRD sample extraction.

anticlockwise at a constant rate and the detector moves in the same direction but at twice the rate such that when the sample is at an angle of θ , the detector is at an angle of 2θ . The detector measures the intensity of x-rays at each angle and relays the data to a computer and/or chart recorder to build a diffraction pattern. Each compound has a unique diffraction pattern. The spacings between the crystalline lattices (d) can be calculated using the equation:

$$d = \frac{\lambda}{2 \sin \theta} \quad \text{Equation 6-1}$$

where λ is the wavelength of the radiation used and θ is the angle at which a intensity peak was observed.

The compound can be identified by matching the calculated d spacings with known d spacings of compounds recorded in a data base of diffraction standards [184].

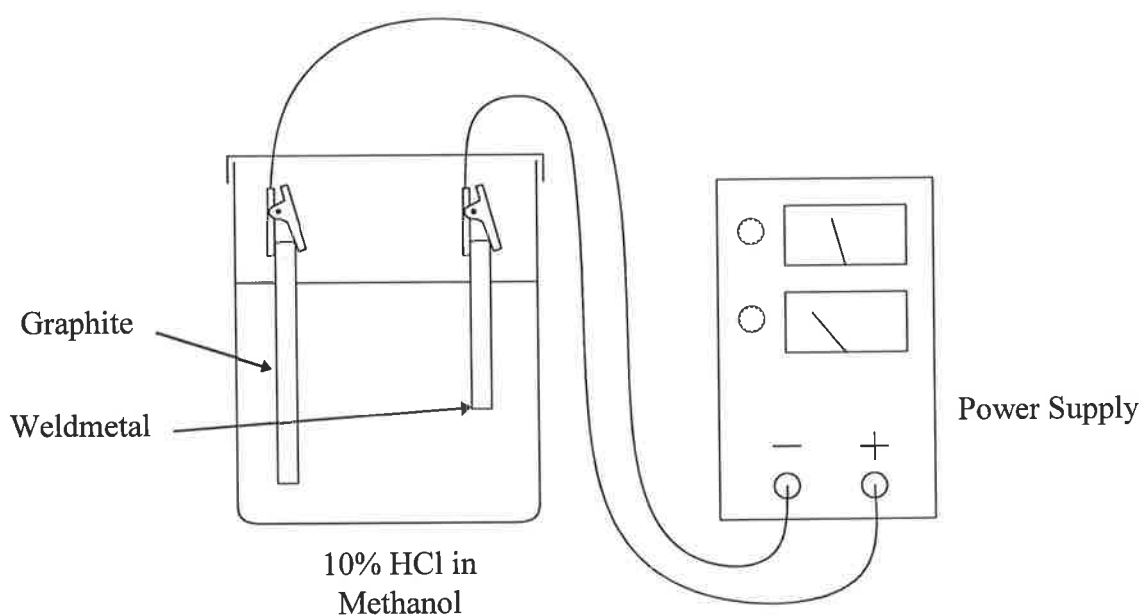


Figure 6-8: Apparatus for dissolving weldmetal for XRPD analysis.

XRPD was carried out using a Phillips PW1130 diffractometer mounted in the vertical position. Cobalt radiation ($K\alpha$, $\lambda = 1.790 \text{ \AA}$) was used and diffraction intensities were measured between the angle $2\theta = 20^\circ$ and $2\theta = 90^\circ$. The scan rate used was $1^\circ/\text{min}$ and a 1° divergent slit was used.

Diffacted intensities were recorded on a linear chart recorder and logged into a computer simultaneously. Computer logging was achieved via a voltage signal from the diffractometer passed through an analog/digital converter. A simple data logging program was utilised to store the voltage signals. The sampling rate was set at 1Hz which effectively meant 60 data points were recorded for each degree of rotation.

The diffraction patterns were then analysed to locate any intensity peaks and the results were compared with powder diffraction standards.

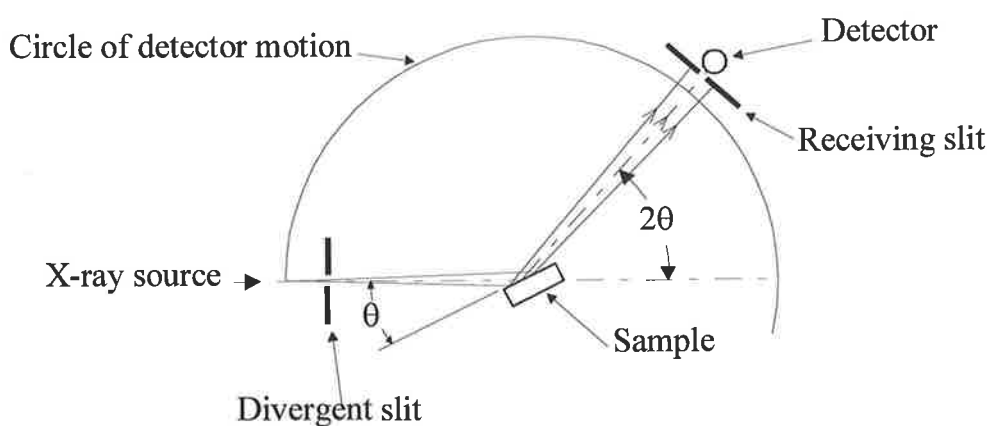


Figure 6-9: Schematic diagram of an XRPD apparatus.

6.2.3 Materials Characterisation

6.2.3.1 Macroscopic Examination

A section of the welded joint was extracted from each test plate to examine the weld profile and to calculate the weld dilution. Sections, 10 mm wide were cut from the test plate using a water-cooled friction saw and ground with wet SiC papers. Following grinding the samples were cleaned with ethanol and polished using a 6 μ m diamond cloth. The samples were subsequently etched with 5% nital (5% nitric acid, 95% ethanol) to reveal the weld zone and the heat affected zone (HAZ). The etched samples were then photographed using a camera fitted with a macro lens under vertical illumination.

6.2.3.2 Dilution Measurements

Dilution is the degree to which the base material contributes to the resultant weld deposit. It is normally expressed as the percentage of melted base material in the total weld metal [185]. After photography, the macro sections were placed on a shadowgraph which projected a magnified image of the weld profile onto a glass screen. The weld profile was copied onto tracing paper attached to the glass screen. The traced weld profiles were scanned into a computer and were superimposed over the original joint preparations. Figure 6-10 shows an example of the original joint preparation overlaid with the weld profile. Image analysis software was used to determine the area of melted base material (shaded regions 1 and 3 in Figure 6-10), and the total weld area (regions 1, 2, and 3 in Figure 6-10). Dilution (D) was then calculated using the following equation:

$$D = \frac{A_1 + A_3}{A_1 + A_2 + A_3} * 100\% \quad \text{Equation 6-2}$$

where A_1 , A_2 and A_3 are the areas of regions 1, 2 and 3 respectively in Figure 6-10.

6.2.4 Light Microscopy

Light microscopy was used to characterise the microstructure and identify phases present in the weld deposit.

Samples were prepared for metallographic examination by cutting cross sections of the welded joints using a water cooled friction saw and mounting in bakelite. After mounting the samples were wet ground using four grades of SiC papers. Following grinding the samples were cleaned using ethanol and an ultrasonic cleaner to remove grinding particles and polished with 6 μ m, 3 μ m, and 1 μ m diamond cloths. Between each polishing stage the samples were cleaned to prevent contamination. Following polishing the samples were etched in 2% nital for approximately 10 seconds to reveal the microstructure.

Light microscopy was performed using an Olympus PMG3 photo-microscope at magnifications ranging from x25 to x1000. Areas of interest were photographed using a Kodak DCS 410 digital camera attached to the microscope.

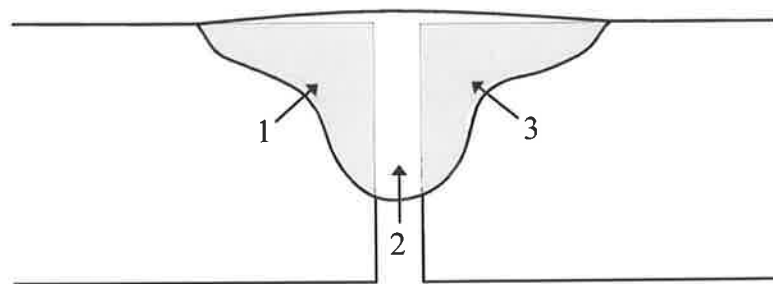


Figure 6-10: Diagram showing the original joint preparation overlaid with the weld profile.

6.2.4.1 Quantitative Microscopy

The microstructure of weld deposits often determines the behaviour of the welded joint. Point counting was carried out to determine the volume fractions of micro constituents in the weld deposit.

Samples for point counting were prepared in a similar manner to that described in section 6.2.4 except the etching time was longer. This was recommended in the guidelines published by the IIW [57] to ensure all grain boundaries were visible. Samples were examined under a Leitz microscope fitted with a Swift point counting device. A total of 1000 points were examined on each sample and the results stored in the point counter. The magnification used was x500 and the point to point distance was 15 μm .

Point counting was conducted in accordance with IIW recommendations [57] with the following deviations:

- Grain boundary ferrite (PF(G)) and Intragranular polygonal ferrite (PF(I)) were combined under the name allotriomorphic ferrite.
- Ferrite with second phase (FS) was called Widmanstätten ferrite. Further subdivision to include non-aligned ferrite (FS(NA)) was not considered necessary since it has been shown that this phase is actually Widmanstätten ferrite which exhibits a different orientation relative to the plane of the polished section [186].

6.2.5 Scanning Electron Microscopy

Scanning electron microscopy (SEM) was used to examine the welds in several ways. Microsegregation in the weld was examined by element mapping regions using a wavelength dispersive x-ray spectrometer. A Field Emission Scanning Electron Microscope (FESEM)

was used to examine the microphases in the weld microstructure at high magnification. A SEM was also used to examine the fracture surfaces from broken Charpy samples and tensile samples.

6.2.5.1 Wavelength Dispersive X-Ray Spectrometry

Wavelength dispersive x-ray spectrometry (WDX) was utilised to perform microanalysis on the weld samples. A wavelength dispersive spectrometer consists of an electromechanical system attached to a SEM that uses diffracting crystals and gas-ionisation detectors to collect and analyse x-rays.

6.2.5.1.1 WDX overview

The following overview has been summarised from the Microspec 600 wavelength dispersive x-ray spectrometry system user manual [187].

WDX relies on a scanning electron microscope to excite x-rays within a specimen, which are then detected, processed, and analysed. A schematic of the WDX system is shown in Figure 6-11. The electron beam from the SEM strikes the sample producing electrons and

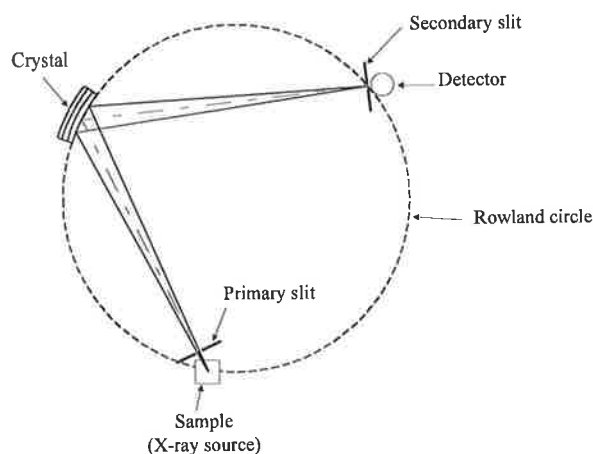


Figure 6-11: Schematic diagram showing the x-ray path for WDX.

x-rays. A portion of the x-rays produced have wavelengths and energies that are characteristic of the elements in the sample.

Figure 6-12 shows the electron beam striking the sample, emitting x-rays and electrons. At the spectrometer-SEM interface there is a magnetic electron trap which prevents electrons entering; x-rays however, are unaffected and pass through. The primary slit acts as an aperture and prevents scattered x-rays entering the spectrometer.

X-rays strike the diffracting crystal (see Figure 6-11) which is situated in the diffracting position. If the WDX is set to a fixed position for analysis of a particular element line, the crystal will diffract only x-rays within a narrow energy band and reject all other x-rays. The x-rays from the element of interest are selectively diffracted towards the detectors; all other x-rays experience destructive interference and are absorbed by the surroundings.

A secondary slit is located in front of the detector (see Figure 6-11) to prevent scattered x-rays entering the detector. The width of the slit may be adjusted to suit various testing conditions

6.2.5.1.2 WDX experimental

Microanalysis elemental mapping was carried out to determine levels of micro-segregation in the weld deposit. The equipment consisted of a Cambridge Stereoscan 250 SEM fitted with a Microspec 600 WDX system. The SEM operated at 20kV and 2000 nA. The sample was tilted at 25° with a working distance of 20 mm.

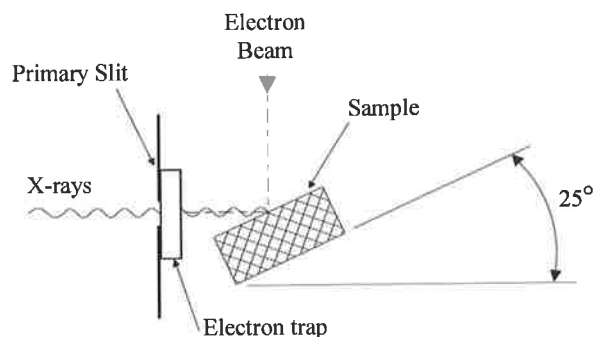


Figure 6-12: Sample configuration for WDX.

In preparation for microanalysis the weld deposits were mounted in bakelite and prepared in a similar manner described previously in section 6.2.4. The samples were examined under a light microscope and a region of interest (ROI) was identified. This ROI was marked using microhardness indents from a 200g load (approximately 40 μm in width). The area marked out by the hardness indents was approximately 300 μm x 300 μm . After marking the samples were briefly polished on a 1 μm diamond cloth to remove the etch only.

The electron beam from the SEM was moved across the ROI in a raster pattern, either 256 x 256 points or 512 x 512 points, dwelling on each point to collect x-ray intensity data. The intensity data was converted to a grayscale number between 0 and 255 and plotted. Lighter areas indicated a higher concentration of the selected element. Due to the shift in electron beam to reach the outer areas of the ROI, a small amount of defocusing occurred which led to a reduced level of x-ray detection. Defocusing was minimised by fully opening the secondary slit to 2.5 mm (see Figure 6-11).

6.2.5.2 Weld Microstructure Examination

To characterise the microphases in the weld deposit metallographic samples which had been lightly etched in 2% nital were examined in a Phillips XL 30 FESEM operating at 10kV. Samples were examined using a secondary electron detector and a backscattered electron detector at a working distance of 12 mm.

6.2.5.3 Fracture Surface Examination

Selected broken Charpy and tensile samples were examined in a Cambridge Stereoscan 250 SEM operating at 10 kV. The fracture surfaces were examined using a secondary electron detector (Charpy samples) and backscatter detector (tensile sample) at a working distance of 20 mm.

6.2.6 Transmission Electron Microscopy

Transmission Electron Microscopy (TEM) was used to identify microphases present between the ferrite laths. TEM was carried out using a Phillips CM200 microscope operating at 200 kV.

6.2.6.1 TEM Foil Preparation

The first stage of TEM foil preparation involved cutting 3 mm diameter cylinders from each weld using electric discharge machining (EDM). Wafers approximately 300 μ m thick, were sliced from the cylinders using a low speed diamond saw. The discs were then wet ground on P800 SiC paper on each side to an overall thickness of 100 μ m. Final thinning of the discs was performed using an electroject thinner, shown schematically in Figure 6-

13. The disc is clamped into polymer sample holder which is placed between two jet nozzles. The disc acts as the anode and the nozzles act as the cathode [188].

The electrolyte consisted of 10% perchloric acid, 20% glycerol and 70% ethanol. A voltage of 70mV was used and the current ranged between 10 and 15 mA depending on the temperature. The electrolyte was cooled to -10°C using liquid nitrogen to slow the reaction rate. Generally the electrolyte temperature rose to -5°C by the conclusion of thinning. The disc was thinned until a perforation appeared. The disc was then removed and checked in an optical microscope .

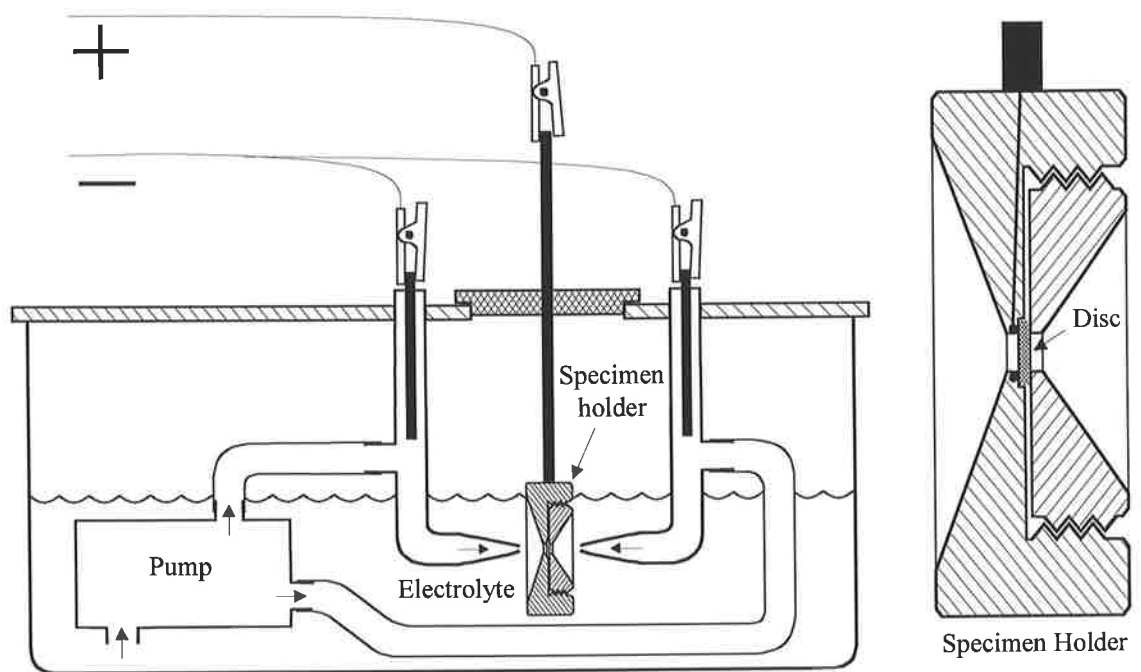


Figure 6-13 Schematic diagram of an electrojet thinner for preparing TEM foils.

7. RESULTS AND DISCUSSION

7.1 CHEMICAL ANALYSIS OF METAL CORE WIRES

The chemical compositions of the experimental metal core wires are listed in Table 7-1. The predicted manganese and silicon contents are shown in italics and underlined. In series 1 where the silicon content was varied, the actual silicon contents of wires 1-2 and 1-3 were lower than predicted, whereas wire 1-4 had a slightly higher silicon content than predicted. The manganese content of series 1 wires varied from 4.60% to 5.02% compared with the predicted value of 4.80%. In series 2, the silicon content, predicted to remain constant at 0.89% ranged from 0.34% to 0.42%. The manganese content increased consistently through series 2 but mostly at lower than predicted values.

All other elements with the exception of aluminium remained relatively constant through both series of metal core wires. The aluminium content was higher in series 1 wires compared with series 2. The ferrosilicon powder ingredient used to add silicon to the metal core wire contained a small amount of aluminium which would explain the higher aluminium levels in series 1. However, no reason can be found to explain the exceptionally high aluminium content found in wire 1-2.

Wire No.	C %	Mn %	Si %	S %	P %	Ni %	Cr %	Mo %	Cu %	V %	Ti %	Al %		
1-1	0.06	4.60	<u>4.80</u>	<.01	<u>0.00</u>	0.012	0.010	0.04	0.01	<.01	0.04	<.01	0.005	
1-2	0.06	5.02	<u>4.80</u>	0.61	<u>0.89</u>	0.012	0.013	0.03	0.02	<.01	0.05	<.01	0.015	
1-3	0.06	4.85	<u>4.80</u>	0.85	<u>1.07</u>	0.012	0.013	0.04	0.03	<.01	0.05	<.01	0.007	
1-4	0.06	4.85	<u>4.80</u>	1.50	<u>1.42</u>	0.012	0.013	0.03	0.02	<.01	0.04	<.01	0.008	
2-1	0.06	0.17	<u>0.17</u>	0.38	<u>0.89</u>	0.010	0.011	0.02	0.03	0.03	0.04	0.02	<.01	<.005
2-2	0.06	1.54	<u>2.44</u>	0.42	<u>0.89</u>	0.010	0.011	0.02	0.02	0.01	0.04	0.01	<.01	<.005
2-3	0.06	2.94	<u>3.62</u>	0.34	<u>0.89</u>	0.011	0.012	0.02	0.03	0.02	0.03	0.01	<.01	<.005
2-4	0.06	5.78	<u>6.00</u>	0.42	<u>0.89</u>	0.011	0.013	0.02	0.03	0.02	0.03	0.01	<.01	<.005

Table 7-1: Chemical analysis of metal core wires.

The predicted metal core wire composition is in *italics and underlined*.

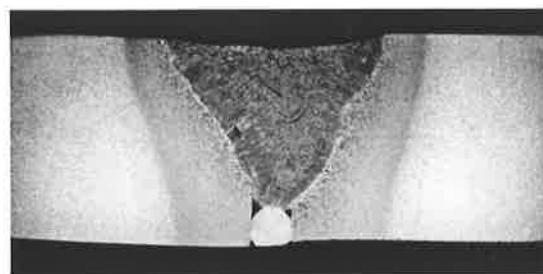
7.2 WELD MACROSTRUCTURE.

Figure 7-1 displays macrographs of the welded joints. Welds 1-4, 2-3, and 2-4 contained some porosity at the base of the weld bead. There was slight joint misalignment in welds 1-1 and 2-1. The 2.5 mm steel wire placed at the base of the joint assisted in maintaining the required gap. However, transverse contraction during the welding process has in some cases deformed the wire spacer and thus reduced the joint gap.

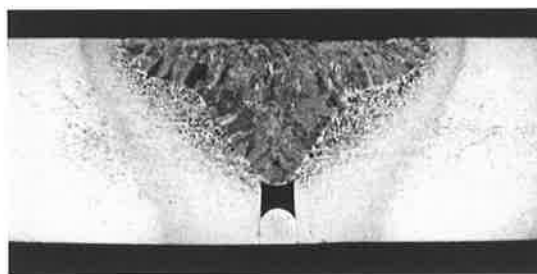
The depth of penetration varied with silicon and manganese contents. As the silicon content increased, the depth of penetration decreased and the width of the weld increased. Conversely, the depth of penetration increased and the width of the weld decreased as the manganese content increased. The weld profiles are discussed further in section 7.3.2.



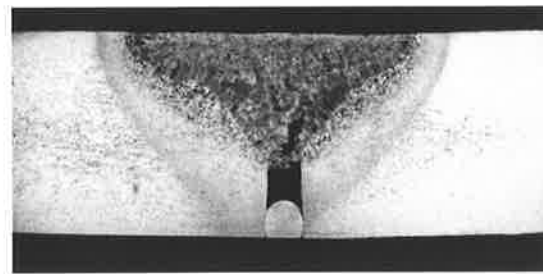
(a) Weld 1-1 0.08% Si 1.38% Mn



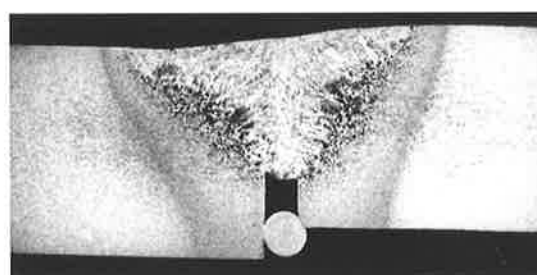
(b) Weld 1-2 0.16% Si 1.54% Mn



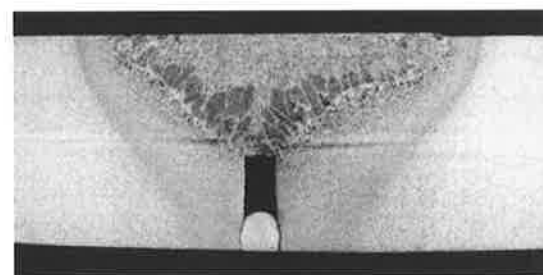
(c) Weld 1-3 0.27% Si 1.38% Mn



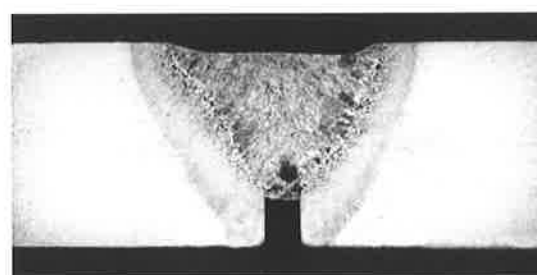
(d) Weld 1-4 0.47% Si 1.61% Mn



(e) Weld 2-1 0.19% Si 0.53% Mn



(f) Weld 2-2 0.20% Si 0.87% Mn



(g) Weld 2-3 0.17% Si 1.15% Mn



(h) Weld 2-4 0.15% Si 1.71% Mn

Figure 7-1: Weld macrographs etched in 5% nital.

Weld No.	C %	Mn %	Si %	S %	P %	Cr %	Ti ppm	Al ppm	O ₂ ppm	N ₂ ppm	C.E.	d/w	Dilution %
1-1	0.12	1.38	0.08	0.011	0.016	0.01	13	264	100	50	0.36	0.78	72.5
1-2	0.12	1.54	0.16	0.011	0.016	0.01	12	273	140	62	0.38	0.75	74.4
1-3	0.12	1.38	0.27	0.011	0.017	0.01	11	301	110	63	0.36	0.46	76.4
1-4	0.12	1.61	0.47	0.012	0.017	0.02	16	253	60	57	0.40	0.45	77.3
2-1	0.11	0.53	0.19	0.011	0.016	0.01	8	250	190	60	0.20	0.47	73.9
2-2	0.11	0.87	0.20	0.010	0.016	0.01	8	262	140	61	0.26	0.36	70.5
2-3	0.12	1.15	0.17	0.011	0.017	0.01	15	284	180	67	0.32	0.66	75.8
2-4	0.11	1.71	0.15	0.011	0.016	0.01	10	244	170	75	0.40	0.63	75.0

All welds: Ni: 0.02%, Cu: 0.01%

Table 7-2: Chemical analysis, dilution and depth to width ratios of experimental weld metals.

7.3 CHEMICAL ANALYSIS OF THE WELD DEPOSITS

Table 7-2 lists the ICP-AES total (soluble and insoluble) chemical analysis of the weld deposits. The carbon content was reasonably consistent for all welds, ranging between 0.11% and 0.12%. The manganese content of series 1 welds ranged from 1.38% to 1.61% and the silicon content increased consistently from 0.08% in weld 1-1 to 0.47% in weld 1-4. In series 2, the manganese content increased from 0.53% in weld 2-1 to 1.71% in weld 2-4. The silicon content of the series 2 welds varied from 0.15% to 0.20%.

Sulphur and phosphorus contents remained constant through both series of welds, only varying by 0.001%. Nickel, chromium, and copper also remained consistent through all

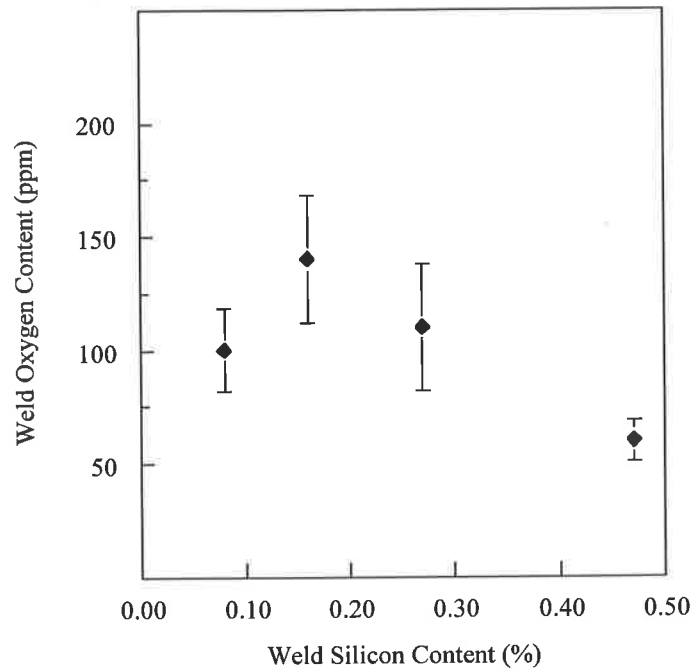


Figure 7-2: Effect of weld silicon content on weld oxygen content.

welds. A very small variation in titanium content was observed for both series. The aluminium content ranged between 244 and 301 ppm.

The oxygen content varied in both series. Since the supposedly fixed element concentration varied (1.38% - 1.61% manganese in series 1 and 0.15% - 0.20% silicon in series 2), there may be some difficulty in isolating the precise effect of each element. In addition, manganese and silicon partition between the matrix oxide inclusions which further complicates the problem. However the experimental results can be used to suggest trends in behaviour.

Figure 7-2 plots oxygen content as a function of weld silicon content. Weld oxygen appears to peak at approximately 0.15% silicon before decreasing with further increases silicon. Evans [145] reported that oxygen decreases as silicon increases, however, Liu and Olson [81] found the opposite effect. Figure 7-3 plots oxygen content as a function of

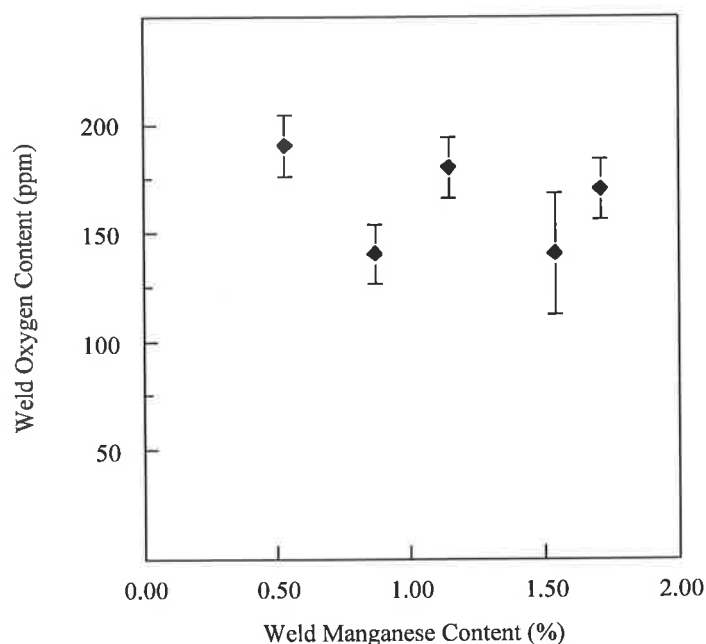


Figure 7-3: Effect of weld manganese content on weld oxygen content.

weld manganese content and shows that manganese appears to have little influence on weld oxygen content. Liu and Olson [81] found that the weld oxygen content decreased with increasing manganese content.

7.3.1 Carbon Equivalent

The carbon equivalent (C.E.) for each weld is also displayed in Table 7-2. In series 1 the carbon equivalent does not vary greatly since silicon was not used in determining C.E. (see Table 5-1). Other carbon equivalents such as the Ito-Bessyo carbon equivalent (Pcm) [189] do consider silicon in their determination, however, in this experimental program the carbon equivalent used in AS 3678 [179] was chosen (see Table 5-1, page 60). In series 2, C.E. increases from 0.20 in weld 2-1 to 0.40 in weld 2-4 highlighting the influence of manganese on hardenability.

7.3.2 Weld Bead Profile

The depth to width ratio (d/w) for all welds are also listed in Table 7-2 and plotted as a function of silicon in Figure 7-4 and manganese in Figure 7-5.

In Figure 7-4 the depth to width ratio drops sharply from 0.75 at 0.16% silicon to 0.45 at 0.27% silicon. Contrary results were reported by Walsh and Savage [190] who found that silicon increased the weld pool depth and d/w . Their explanation was that the presence silicon reduced the melt viscosity. However, Olyer *et al* [191] found that if the silicon content was greater than 0.7% the fluidity of the weld pool became excessive causing the weld pool to widen.

There are numerous possible explanations for the dramatic variation in weld pool shape as the silicon content increases. One possibility is that the surface tension coefficient of the weld pool changed from a positive value to a negative value between 0.16% and 0.27% silicon thereby reversing the Marangoni flow and altering the weld pool shape.

Keene conducted a thorough review of data for the surface tension of iron and its binary alloys [192]. All but two of the 23 references cited by Keene showed that in an iron - silicon binary alloy the surface tension coefficient was negative. According to the model by Heiple and Roper [25] this would induce flow from the centre of the weld pool outward producing a wide shallow weld pool. The sudden change in d/w could indicate that two opposing forces are acting on the weld pool, the Marangoni force inducing outward flow and another force, possibly the Lorentz force, inducing inward flow. At low silicon contents the inward acting force may dominate but as the silicon content increases the outward Marangoni force may take over. Walsh and Savage [190] stated that in the ranges of currents normally employed in thin section GTAW (in their case: 150A) flow velocities

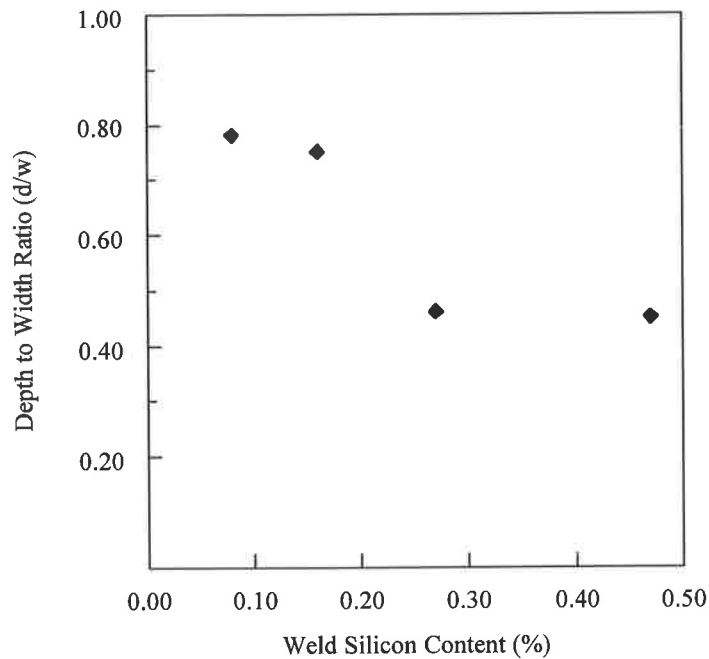


Figure 7-4: Weld depth to width ratio (d/w) as a function of weld silicon content.

caused by surface tension are found to be order of magnitude greater than the Lorentz forces from the arc. However, the Lorentz force is proportional to the current squared, indicating that at high currents the Lorentz force may be of the same order as the surface tension force.

Another explanation could be that the increased presence of silicon deoxidises the weld pool reducing the amount of oxygen available to act on the surface tension gradient.

In contrast to silicon, the addition of manganese increased d/w (Figure 7-5). Since manganese reacts with both the major surface active agents (sulphur and oxygen), one would anticipate that the addition of manganese would impart a negative surface tension gradient and therefore favour the formation of a shallow weld bead. In this case the opposite has occurred indicating that the influence of some alloying elements on weld pool circulation and shape is not fully understood. Metcalfe and Quigley [193] reported a loose

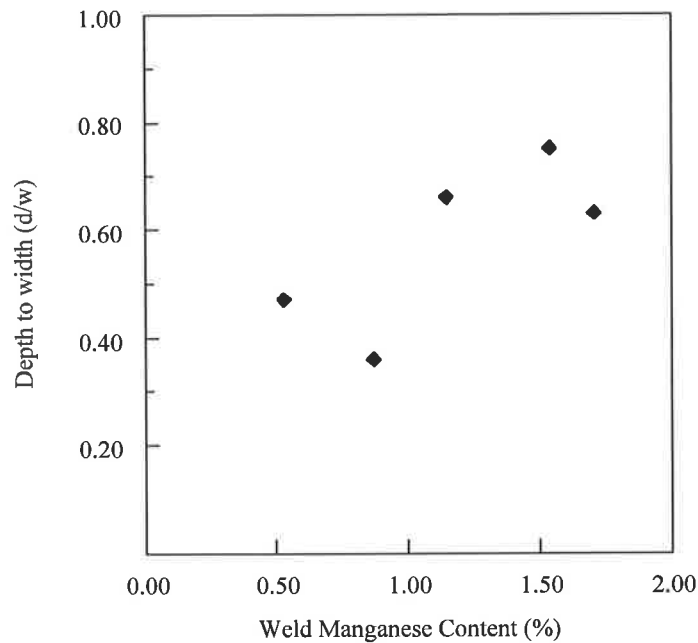


Figure 7-5: Weld depth to width ratio (d/w) as a function of weld manganese content.

correlation showing that the addition of manganese decreased d/w in stainless steel welds. This was contrary to the findings in the present study.

7.3.3 Dilution Measurements

Also listed in Table 7-2 is the measured dilution for each weld. The areas and calculations used to determine the weld dilutions are contained in Appendix B. There appears to be a slight increase in dilution as the silicon content increases. The dilution increased from 72.5% at 0.08% silicon to 77.5% at 0.47% silicon. This increase may be associated with the change in weld bead profile mentioned in section 7.3.2. As the silicon content increased, the weld profile broadened, which increased the volume of base material melted.

There did not appear to be any definite trend in dilution measurements with weld manganese content. In the second series of welds the dilution ranged between 70.5% and 75.8%.

7.4 WELD MICROSTRUCTURE

7.4.1 The Effect of Silicon on Weld Microstructure

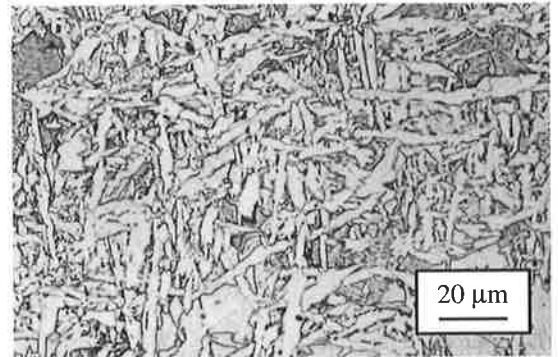
Figure 7-6 contains light micrographs of the microstructure of welds 1-1, 1-2, 1-3, and 1-4 after etching in 2% nital. The microstructure of these four welds with varying silicon content varied little. The microstructure consists mainly of allotriomorphic ferrite, Widmanstätten ferrite, and acicular ferrite. Fine ribbons of allotriomorphic ferrite outline some of the prior austenite grain boundaries. Widmanstätten ferrite was observed in some areas nucleating from allotriomorphic ferrite. Pockets of acicular ferrite were present inside the prior austenite grains.

The results of quantitative microscopy for all welds are listed in Table 7-3. The quantitative microscopy results verified the similarity in microstructure for different silicon contents. Figure 7-7 shows graphically that volume fraction of micro-constituents changed little as the silicon content increased from 0.08% to 0.47%. The volume fraction of acicular ferrite appeared to peak at the expense of Widmanstätten ferrite at 0.27% silicon. The volume fraction of allotriomorphic ferrite remained constant as the silicon content increased.

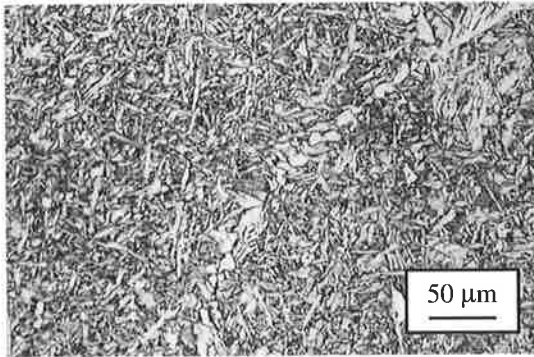
Figure 7-8 shows solute banding observed adjacent to the fusion line in weld 1-2. As indicated in Figure 7-8 (a), the bands run parallel to the fusion line and are the result of periodic increases in the growth rate causing an enrichment of solute. Microhardness testing showed that the banded region had a hardness of 420 HV0.025 indicating a martensitic microstructure. Solute banding was not observed in any other experimental welds.



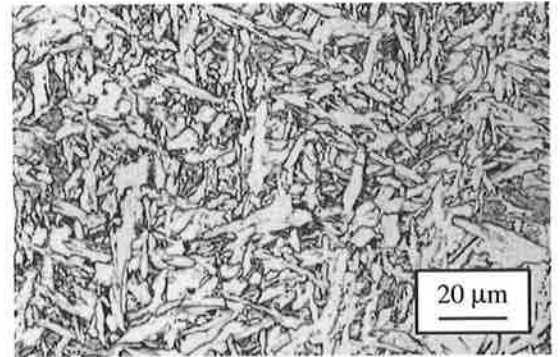
(a) Weld 1-1 0.08% Si 1.38% Mn



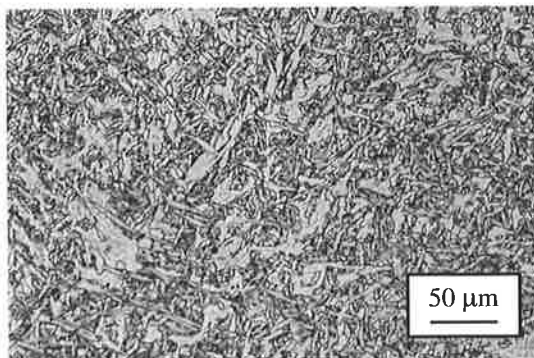
(b) Weld 1-1, higher magnification



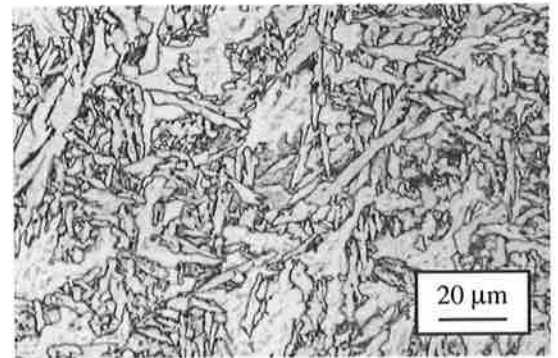
(c) Weld 1-2 0.16% Si 1.54% Mn



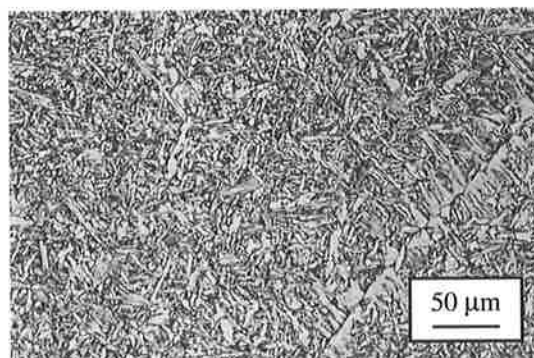
(d) Weld 1-2, higher magnification



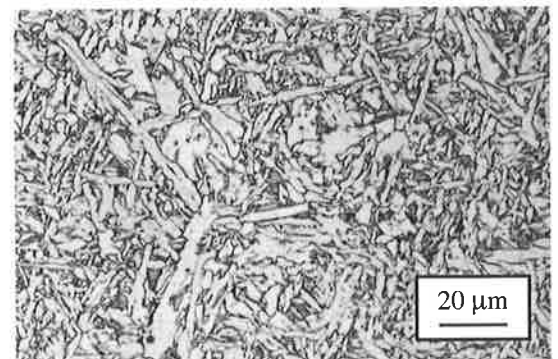
(e) Weld 1-3 0.27% Si 1.38% Mn



(f) Weld 1-3, higher magnification



(g) Weld 1-4 0.47% Si 1.61% Mn



(h) Weld 1-4, higher magnification

Figure 7-6: Light micrographs of series 1 welds etched in 2% nital.

Weld No.	Microstructural Component (%)			
	Allotriomorphic Ferrite	Widmanstätten Ferrite	Acicular Ferrite	MAC
1-1	14.8	30.4	51.4	3.4
1-2	22.9	19.0	55.3	2.8
1-3	15.4	23.4	58.4	2.8
1-4	17.4	28.4	53.0	1.2

Table 7-3: Quantitative microscopy results for series 1 welds.

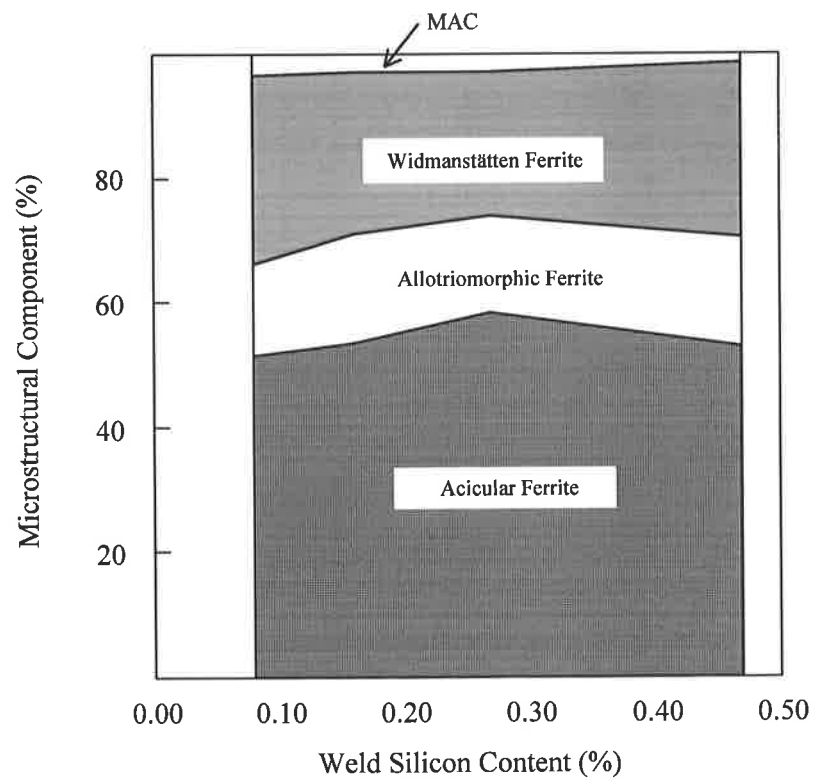
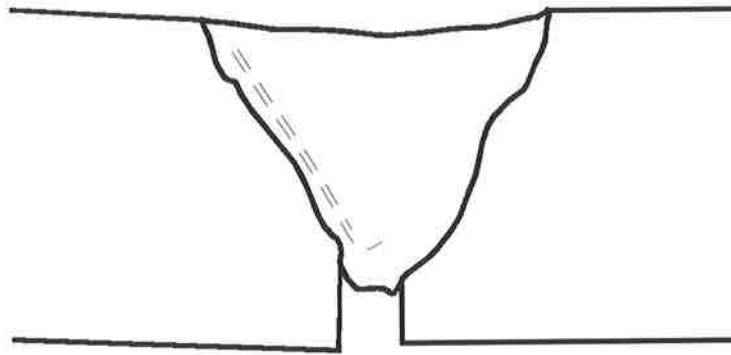
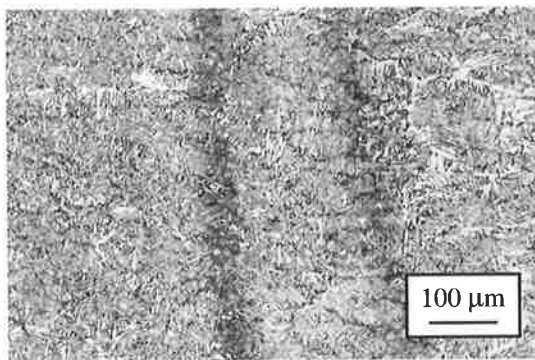


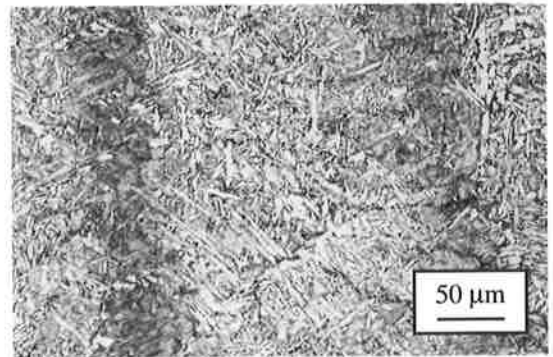
Figure 7-7: Effect of silicon content on the volume fraction of micro-constituents.



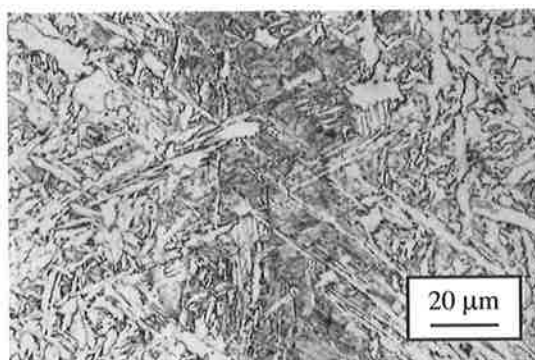
(a) Schematic diagram showing the location of solute banding in Weld 1-2.



(b) Light micrograph of the solute banding region.



(c) Higher magnification of (b).



(d) High magnification view of the left side solute band.



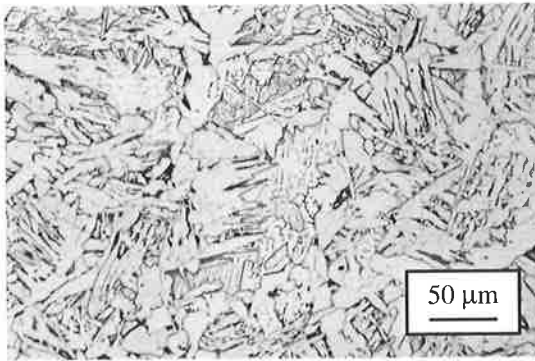
(e) High magnification view of the right side solute band.

Figure 7-8: Light micrographs of the solute banding region observed in Weld 1-2.

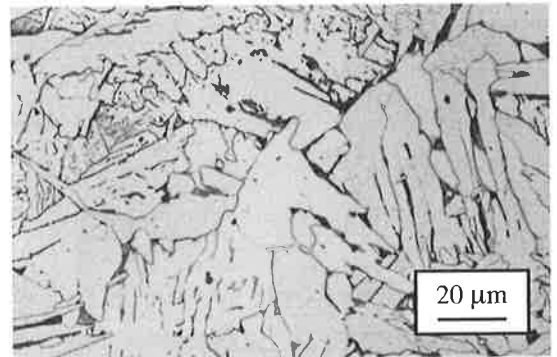
7.4.2 The Effect of Manganese on Weld Microstructure

Figure 7-9 shows light micrographs of the microstructure of welds 2-1, 2-2, 2-3, and 2-4 after etching in 2% nital. The microstructure varied greatly as the manganese content increased. At low manganese, Figure 7-9 (a) and (b) (weld 2-1 containing 0.53% Mn), the microstructure consisted of coarse grains of allotriomorphic ferrite and Widmanstätten ferrite. As the manganese content increased, the microstructure generally became more refined and larger amounts of acicular ferrite became visible.

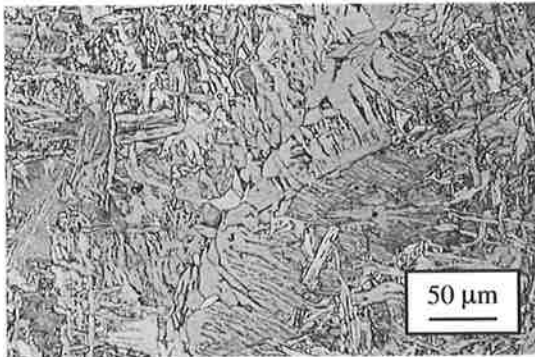
These observations were confirmed by quantitative microscopy. The results of point counting is listed in Table 7-4 and plotted in Figure 7-10. At 0.53% manganese there was only 6.6% acicular ferrite. As the manganese content increased to 1.71%, the acicular ferrite content increased to 62.5%. The increase in acicular ferrite was mainly at the expense of Widmanstätten ferrite and microphase, martensite, austenite, carbide (MAC), and partially at the expense of allotriomorphic ferrite.



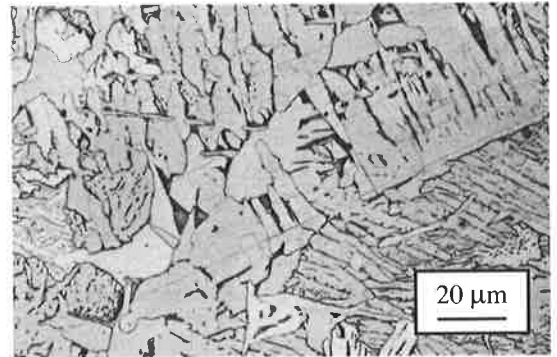
(a) Weld 2-1 0.19% Si 0.53% Mn



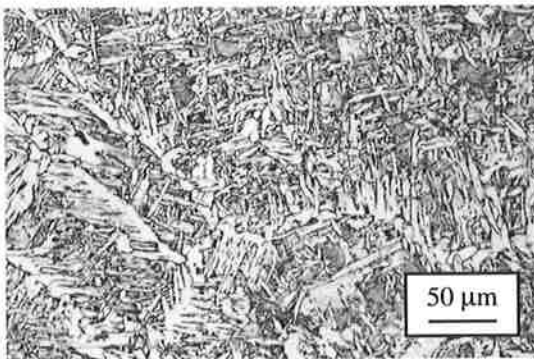
(b) Weld 2-1, higher magnification



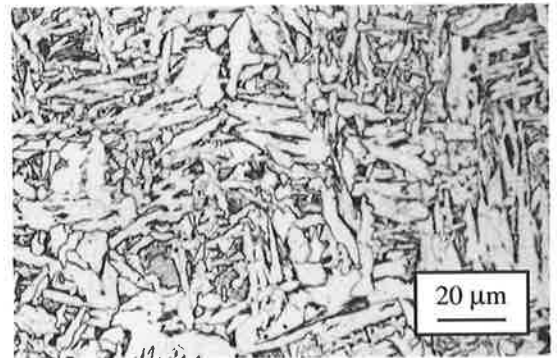
(c) Weld 2-2 0.20% Si 0.87% Mn



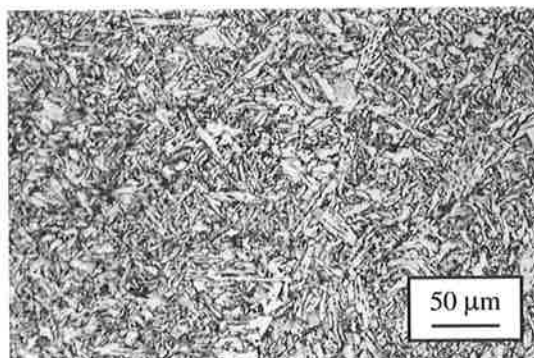
(d) Weld 2-2, higher magnification



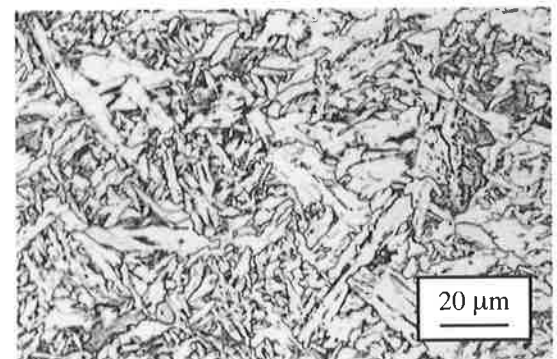
(e) Weld 2-3 0.17% Si 1.15% Mn



(f) Weld 2-3, higher magnification



(g) Weld 2-4 0.15% Si 1.71% Mn



(h) Weld 2-4, higher magnification

Figure 7-9: Light micrographs of series 2 welds etched in 2% nital.

Weld No.	Microstructural Component (%)			
	Allotriomorphic Ferrite	Widmanstätten Ferrite	Acicular Ferrite	MAC
2-1	28.4	56.6	6.6	9.4
2-2	24.4	49.0	18.5	8.1
2-3	17.1	36.6	39.5	6.8
2-4	19.2	17.7	62.5	0.6

Table 7-4: Quantitative microscopy results for series 2 welds.

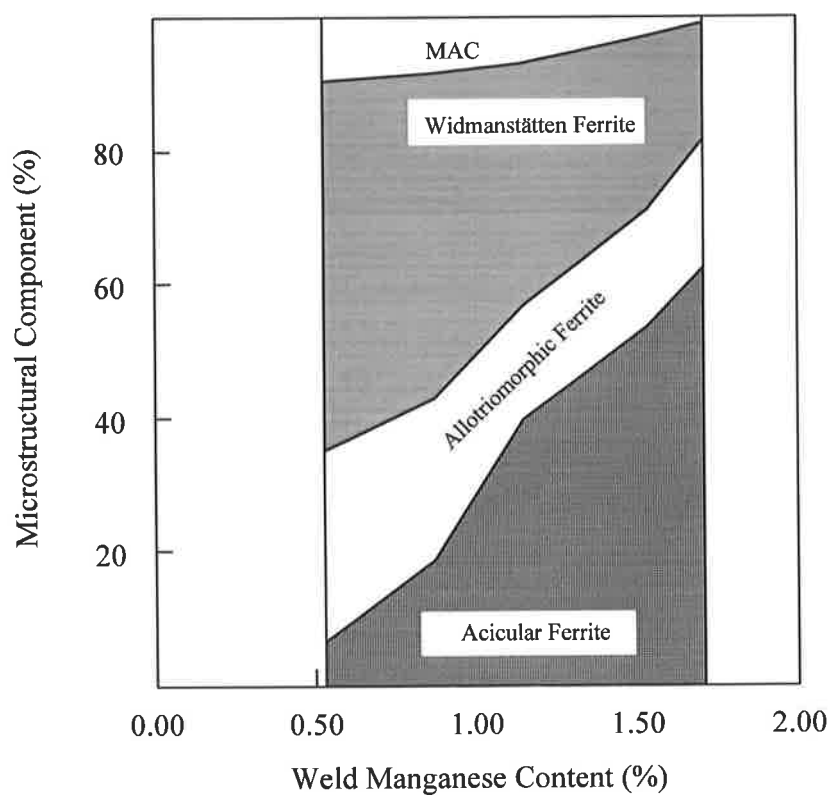


Figure 7-10: Effect of manganese content on the volume fraction of micro-constituents.

7.5 X-RAY POWDER DIFFRACTION

The XRPD traces from inclusions extracted from series 1 and 2 welds are shown in Figure 7-11 and Figure 7-12 respectively. The chart recorder traces are reproduced in full in Appendix C. All welds displayed a strong peak at 31° ($d = 3.35 \text{ \AA}$). This peak is a characteristic d-spacing from graphite which was introduced to the sample residue from the decomposition of carbide during the dissolution process.

In series 1 (Figure 7-11), the XRPD traces are mainly amorphous with only a few small peaks visible. The XRPD traces from series 2 experimental welds (Figure 7-12) show a greater degree of crystallinity with peaks of larger intensity.

Table 7-5 lists the measured d spacings from each XRPD trace and compares them to the d-spacings of some inclusions commonly reported in literature. The most consistent match appears to be $\gamma \text{ Al}_2\text{O}_3$. The two most intense d-spacings for $\gamma \text{ Al}_2\text{O}_3$ are 1.98 and 1.40 $\text{\AA}$. Other compounds which have some matching d-spacings include $\text{MnO} \cdot \text{Al}_2\text{O}_3$, $\text{MnO} \cdot \text{TiO}_2$ and $2\text{MnO} \cdot \text{TiO}_2$. All welds have a d-spacing at 2.59 or 2.60 $\text{\AA}$ which coincides with the most intense d spacings from $2\text{MnO} \cdot \text{TiO}_2$ and MnS .

Other measured d-spacings do not appear to coincide with those of the inclusions listed indicating that there could be other compounds present. These d-spacings include 1.39, 1.90, 2.31, and 2.72 $\text{\AA}$.

In XRPD, d-spacings usually match to the second or third decimal place. However, Horii *et al* [194] found that since oxides present in weld deposits are not nucleated under ideal conditions their d-spacings may not be exact. Based on this reasoning, compounds have been identified even though their d-spacings do not match published data exactly. Due to

the small intensity of peaks in some XRPD traces, accurate measurement of d-spacings was not possible. The correct identification of inclusions is further complicated by having multiple compounds and layered inclusions in a weld deposit.

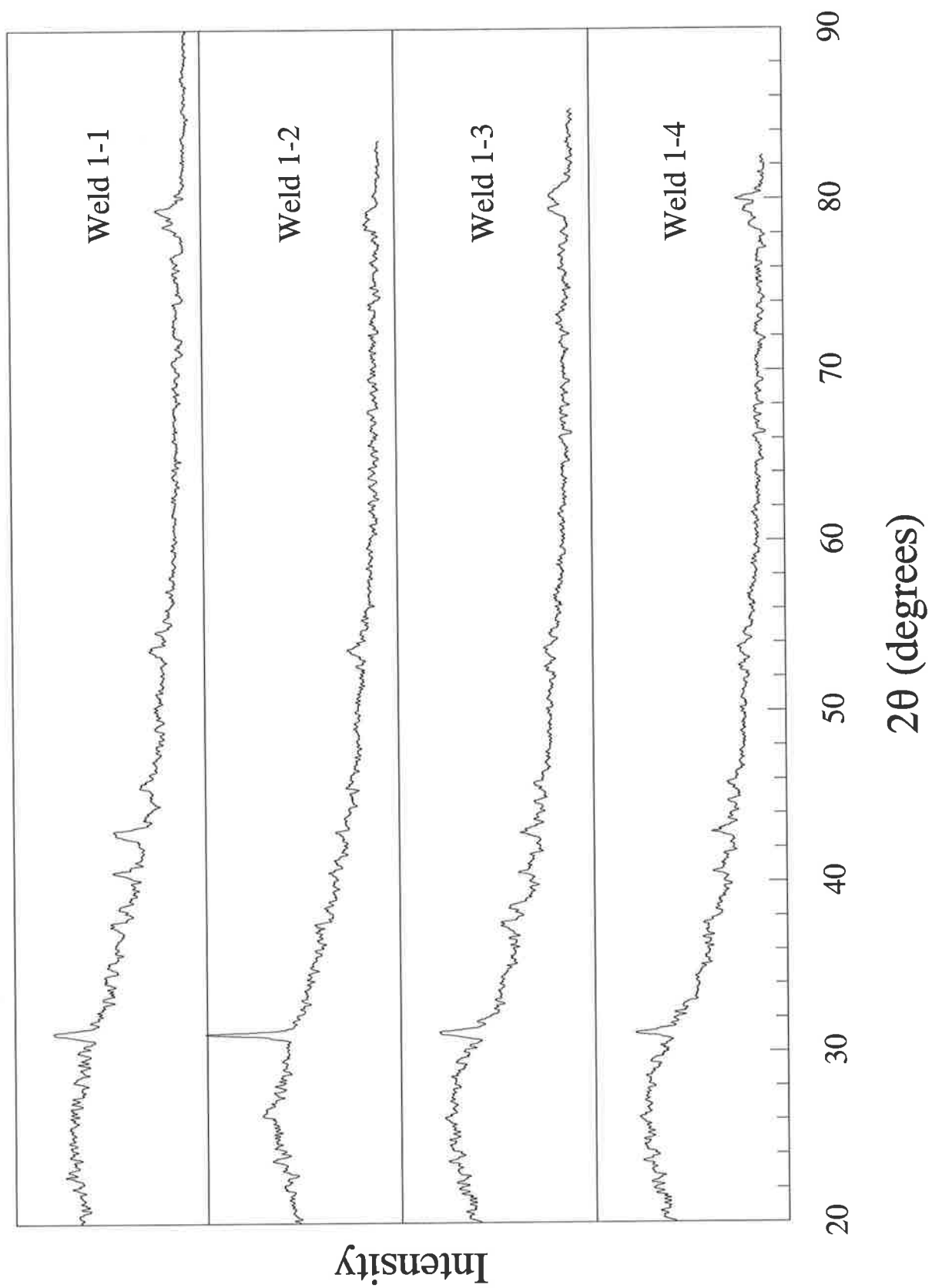


Figure 7-11: XRPD traces for series 1 welds.

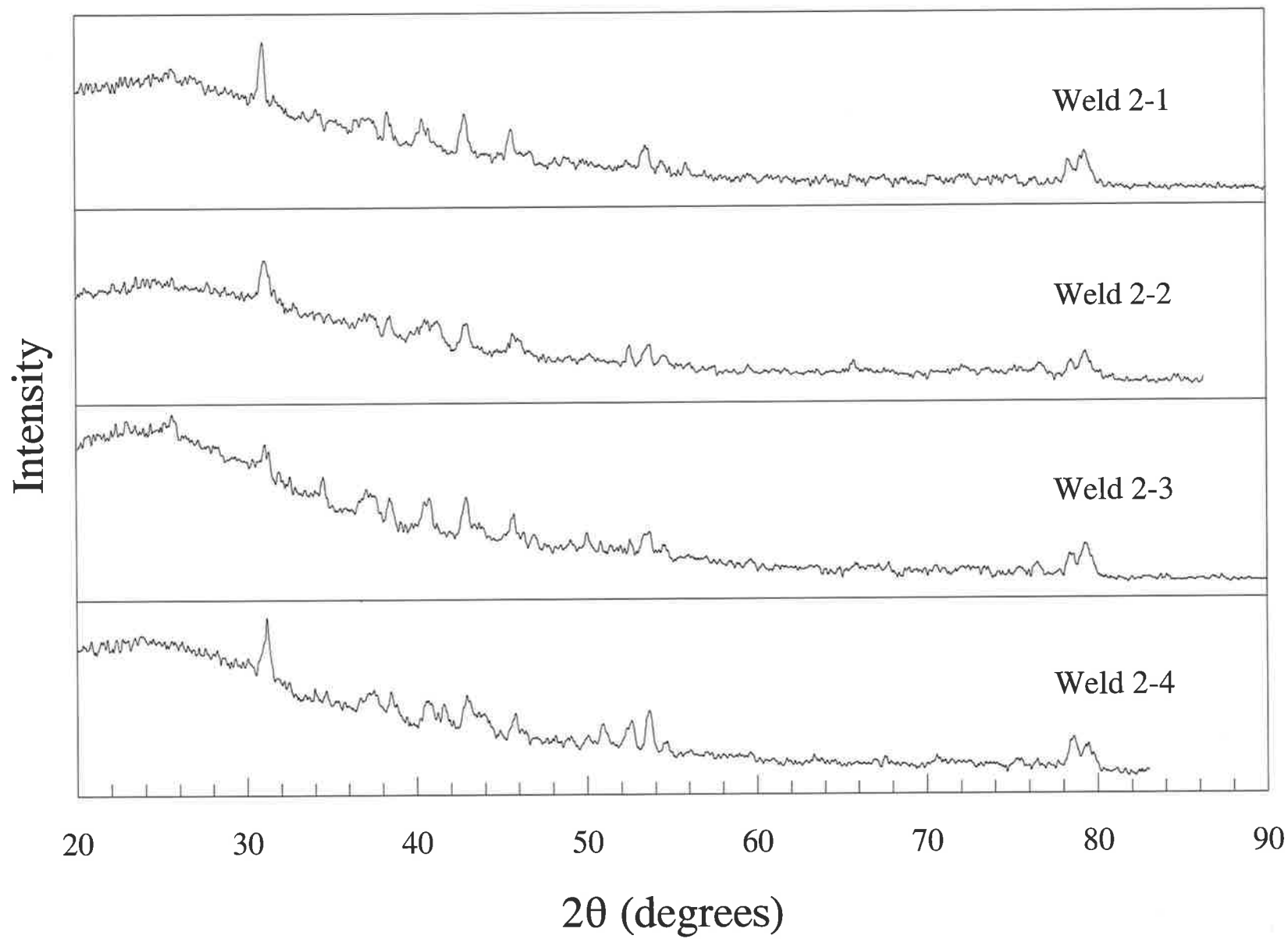


Figure 7-12: XRPD traces for series 2 welds.

d-spacings of experimental welds								d-spacings and intensities (<i>bold italics</i>) of inclusions commonly found in weld deposits					
1-1	1-2	1-3	1-4	2-1	2-2	2-3	2-4	Graphite	Al ₂ O ₃	MnS	MnO. Al ₂ O ₃	MnO. TiO ₂	2MnO. TiO ₂
3.35	3.35	3.35	3.35	3.35	3.34	3.35	3.36	3.36	<i>100</i>				
						3.26							
						3.20							
	3.01			3.05		3.02							3.04 <i>35</i>
											2.90 <i>75</i>		
					2.82	2.82	2.82						
2.79	2.79		2.79				2.80	2.80	<i>45</i>			2.79 <i>100</i>	
	2.72	2.72		2.72	2.72	2.72	2.73						
2.59	2.59	2.59	2.59	2.59	2.58	2.59	2.60			2.61 <i>100</i>			2.60 <i>100</i>
				2.57		2.57	2.57					2.57 <i>80</i>	
	2.54	2.54	2.53		2.54								
2.45	2.45	2.45		2.45	2.45	2.45	2.45				2.47 <i>100</i>		
2.40							2.40						
2.31	2.31	2.31	2.31	2.31	2.31	2.31	2.31						
		2.25				2.25	2.25					2.26 <i>30</i>	
		2.02	2.02	2.02	2.02	2.02	2.02	2.03 <i>30</i>			2.05 <i>30</i>		
1.98	1.98	1.98	1.98	1.98	1.98	1.98	1.98		1.98 <i>80</i>				
1.95		1.95		1.95	1.95	1.95	1.95						
				1.90								1.89 <i>40</i>	
					1.80					1.84 <i>50</i>			
												1.75 <i>40</i>	
	1.68		1.66		1.64			1.68 <i>80</i>					
	1.60	1.61	1.61			1.60	1.60				1.58 <i>50</i>		
		1.51	1.50									1.52 <i>50</i>	1.53 <i>55</i>
	1.48											1.48 <i>50</i>	
		1.43			1.43	1.43					1.45 <i>40</i>		
1.40	1.40	1.40	1.40	1.40	1.40	1.40	1.40		1.40 <i>100</i>				
1.39		1.39	1.39	1.39	1.39	1.39	1.39						

Table 7-5: Comparison of d-spacings measured in the experimental welds to the d-spacings of inclusions commonly found in weld deposits.

7.6 WAVELENGTH DISPERSIVE X-RAY ANALYSIS

The results of WDX element mapping are shown in Figures 7-13 to 7-22. As indicated in Figure 7-13(a), an area of interest was identified using hardness indents which allowed the same region to be examined under a light microscope and a SEM.

The resulting WDX analysis for weld 1-1 is shown in Figure 7-13. The microstructure of the region analysed is shown in Figure 7-13(a) and contains a wide band of ferrite formed on a prior austenite grain boundary running diagonally across the micrograph. The manganese distribution of the same region is shown in Figure 7-13(c). Manganese has a small equilibrium distribution coefficient permitting it to segregate during solidification. As a result of this segregation manganese is usually found in greater concentrations at solidification cell boundaries. Therefore, the structure revealed in Figure 7-13(c) is the delta ferrite solidification structure. The white spots in Figure 7-13(c) indicate inclusions that contain manganese.

The silicon distribution for the same area is shown in Figure 7-13(d). Although the silicon content is very low (0.08%), it appears to have segregated to the solidification cell boundaries. It is also apparent from the low number of white spots on the WDX map that only a small fraction of the inclusions in this weld deposit contain silicon.

The sulphur and aluminium distributions for weld 1-1 are shown in Figure 7-13(e) and (f) respectively. Both elements appear to be present only in the form of particles. From XRPD analysis the aluminium-rich inclusions are most likely γ alumina (Al_2O_3).

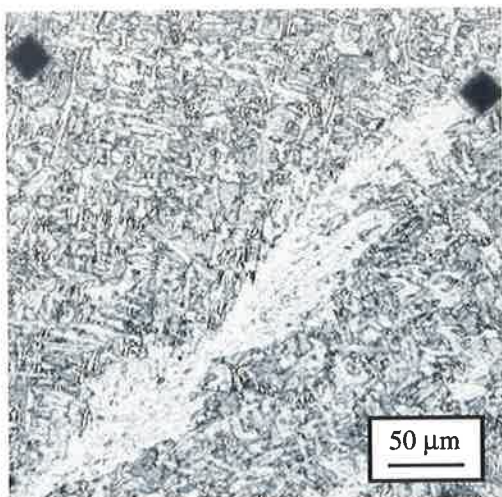
Figure 7-14(b) displays the manganese distribution for weld 1-1 overlayed with aluminium (red), sulphur (yellow) and silicon (blue) rich inclusions. Comparing this with Figure 7-14(a) it is evident that the sulphur distribution coincides with the manganese-rich particles indicating the presence of manganese sulphide (MnS). Some of the manganese sulphides are also associated with γ alumina particles suggesting that the manganese sulphides nucleated on the surface of the γ alumina inclusions.

It is also apparent from Figure 7-14 that the majority of the inclusions lie in the solidification cell boundaries. As mentioned in section 3.5.1.1, alumina forms in the melt and therefore is present before solidification. Results from Figure 7-14 confirm observation made by Dadian [48] and Sugden and Bhadeshia [52] that the inclusions are pushed to the intercellular regions during solidification.

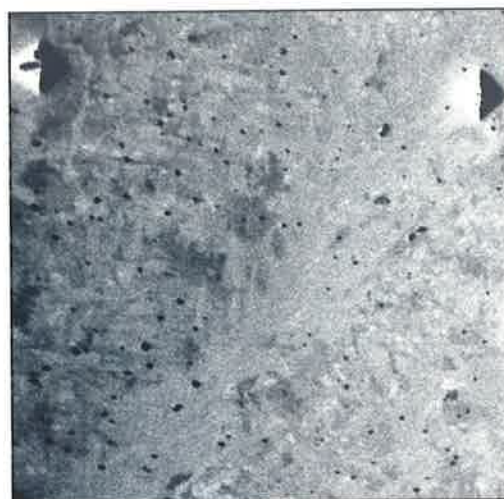
WDX mapping was carried out across the solute banding observed in weld 1-2 and the results shown in Figure 7-15. The results indicate that the banding region is rich in manganese, silicon, and phosphorus. There is also a large number of aluminium-rich (most likely γ alumina) inclusions in the vicinity of the solute banding.

Figure 7-16 and Figure 7-17 show the WDX mapping carried out on welds 1-3 and 1-4 respectively. Unfortunately a computer malfunctioned during the silicon mapping in weld 1-3 (Figure 7-16 (d)) and part of the map was not recorded.

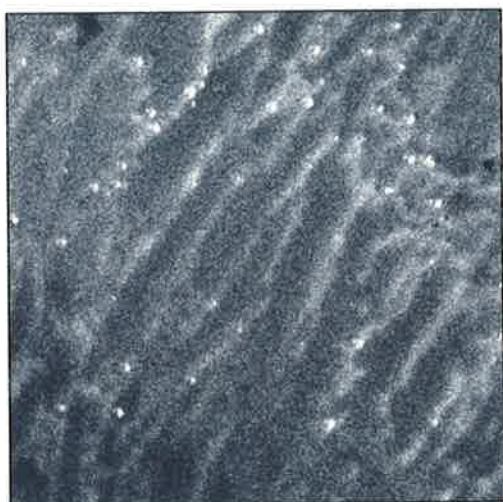
Comparing Figure 7-13 (d), Figure 7-16 (d), and Figure 7-17 (d), only a minor variation in the silicon distribution can be observed. As the silicon content increased the solidification boundaries became slightly more visible indicating a greater amount of silicon segregated.



(a) Light micrograph (2% nital)



(b) SEM Backscattered image (as polished)



(c) Manganese distribution



(d) Silicon distribution



(e) Sulphur distribution

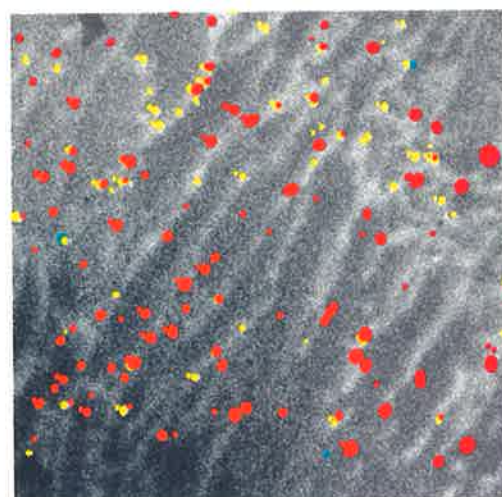


(f) Aluminium distribution

Figure 7-13: Element distributions in weld 1-1.



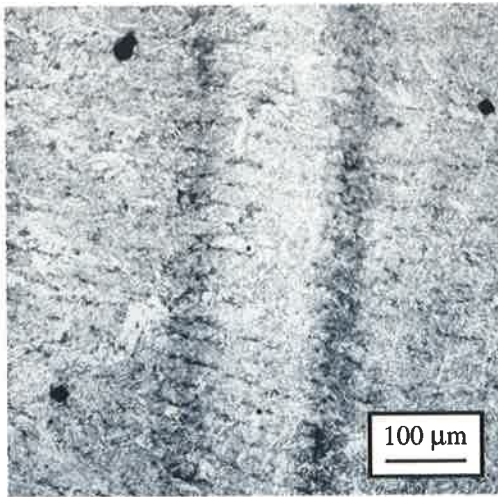
(a) Manganese distribution



Aluminium Sulphur Silicon

(b) Manganese distribution overlayed with Aluminium, Sulphur, and Silicon.

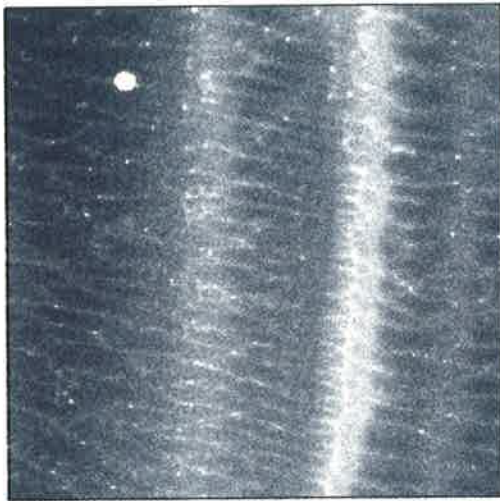
Figure 7-14: Overlay of element distributions in weld 1-1.



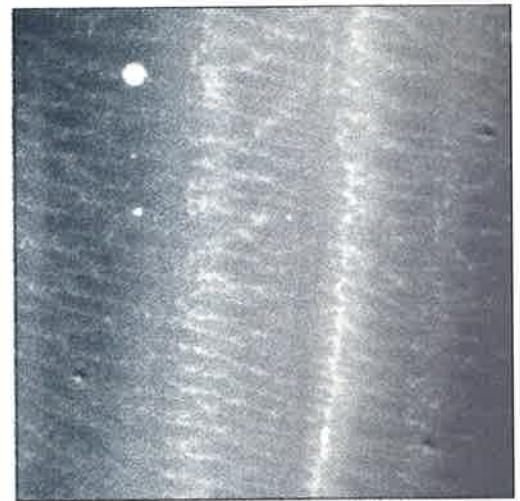
(a) Light micrograph (2% nital)



(b) SEM Backscattered image (as polished)



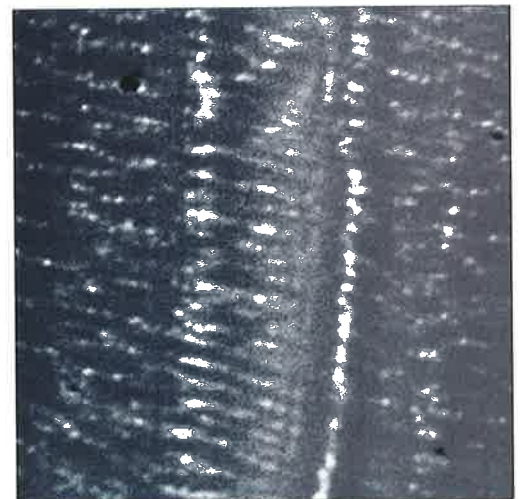
(c) Manganese distribution



(d) Silicon distribution

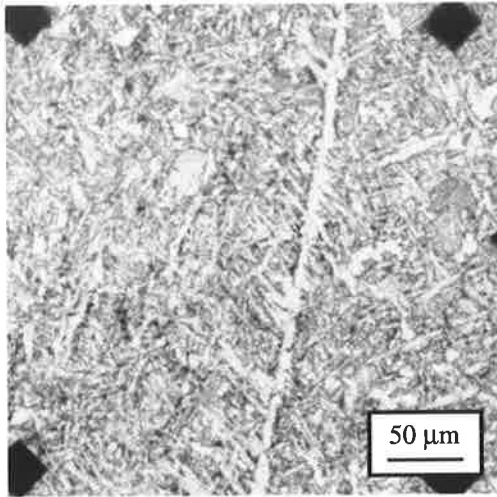


(e) Aluminium distribution

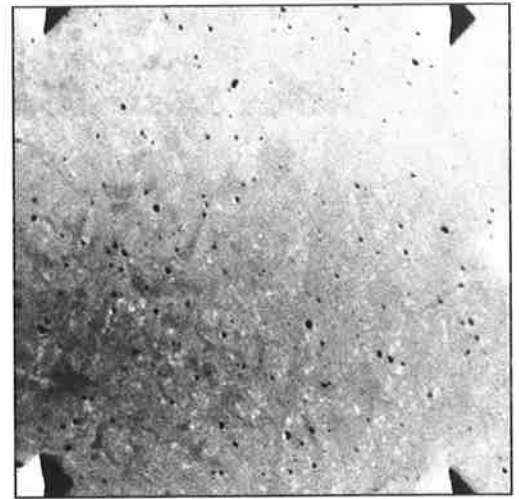


(f) Phosphorus distribution

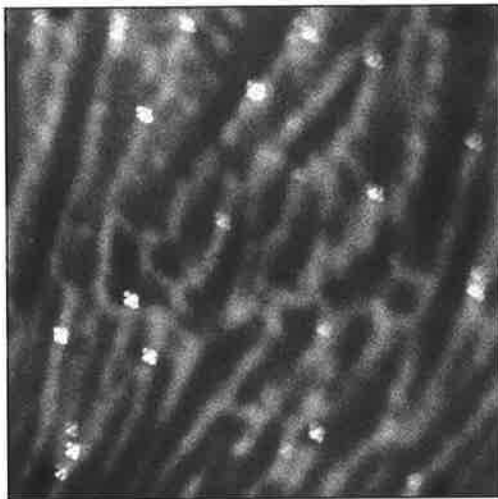
Figure 7-15: Element distributions in weld 1-2.



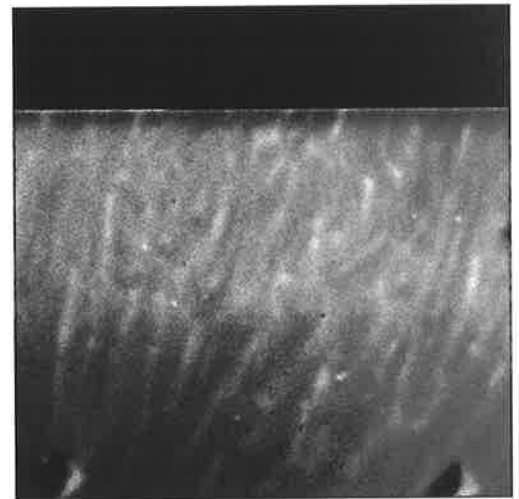
(a) Light micrograph (2% nital)



(b) SEM Backscattered image (as polished)

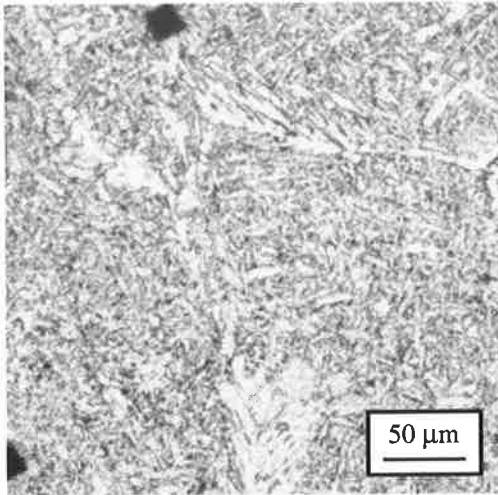


(c) Manganese distribution

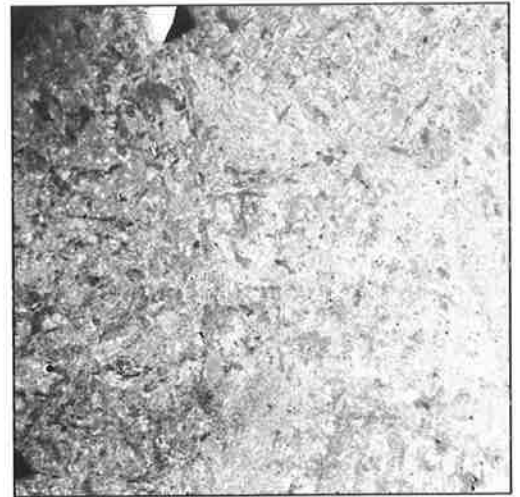


(d) Silicon distribution

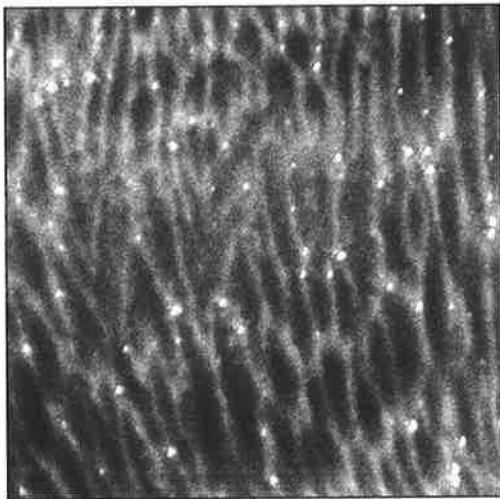
Figure 7-16: Element distributions in weld 1-3.



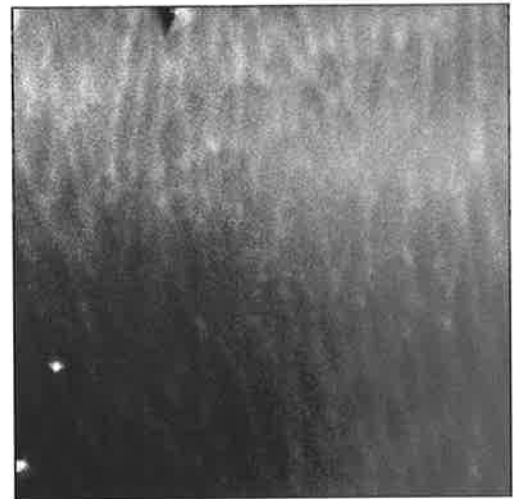
(a) Light micrograph (2% nital)



(b) SEM Backscattered image (as polished)



(c) Manganese distribution



(d) Silicon distribution



(e) Sulphur distribution



(f) Aluminium distribution

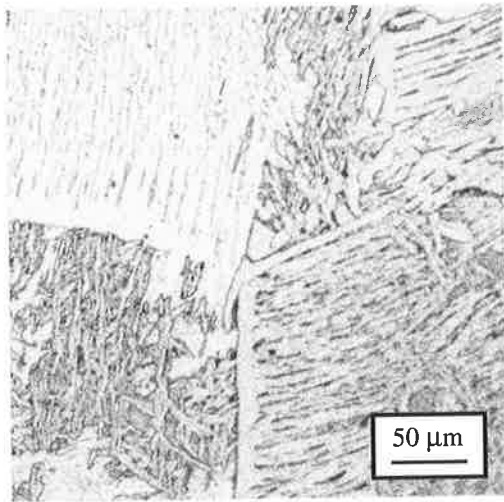
Figure 7-17: Element distributions in weld 1-4.

The elemental distributions for weld 2-1 are shown in Figure 7-18 and indicate that even at relatively low concentrations, manganese still segregates to the solidification cell boundaries.

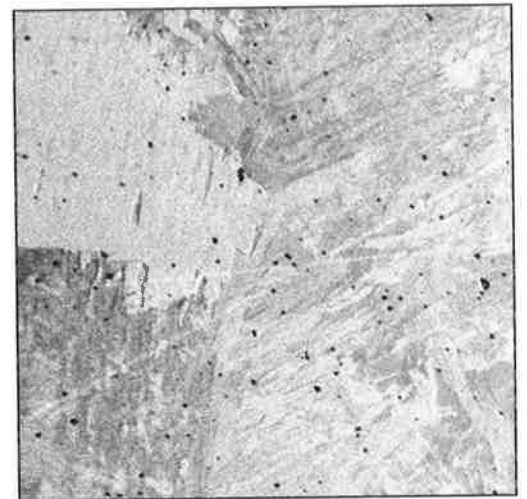
A comparison of the manganese distributions from Figures 7-18 to 7-21 indicate that there is little difference in the segregation characteristics as the manganese content increases. It should be noted however that these results are qualitative and therefore a direct measurement of the manganese concentration at the cell boundaries is not possible. To obtain quantitative data, the samples would require examination in a micro-probe.

A large inclusion was observed in the centre of the region marked for element mapping in weld 2-3 (see Figure 7-20). The entire inclusion is rich in manganese and half the inclusion is rich in silicon indicating a dual phase inclusion containing a manganese rich crystalline compound in a glassy matrix containing manganese and silicon.

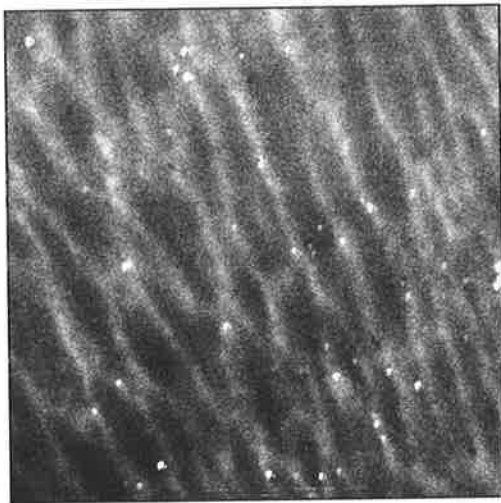
Figure 7-22 shows overlays of aluminium and sulphur on the manganese distribution in weld 2-4 (Figure 7-22(a) and (b)), and an overlay of aluminium over the oxygen distribution (Figure 7-22(c) and (d)). As indicated previously all the sulphur particles coincide with manganese-rich particles indicating the presence of manganese sulphides, and the majority of aluminium-rich particles lie on solidification cell boundaries. An overlay of aluminium and oxygen distributions (Figure 7-22(c) and (d)) show that most of the aluminium particles coincide with high oxygen concentrations confirming the presence of γ alumina.



(a) Light micrograph (2% nital)



(b) SEM Backscattered image (as polished)

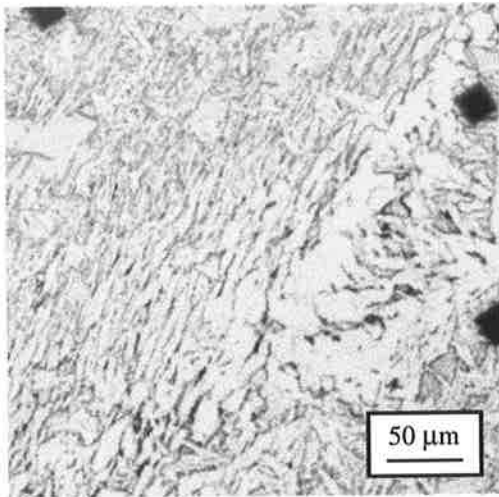


(c) Manganese distribution



(d) Sulphur distribution

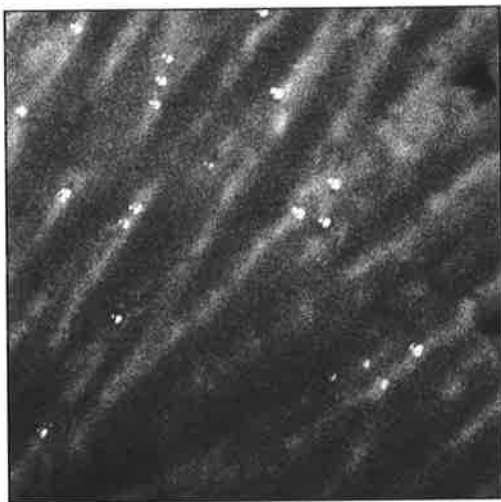
Figure 7-18: Element distributions in weld 2-1.



(a) Light micrograph (2% nital)



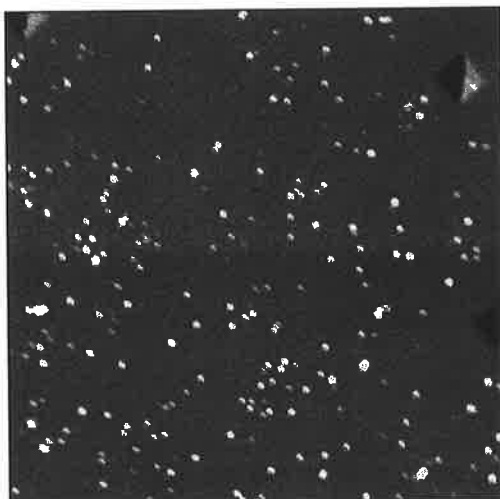
(b) SEM Backscattered image (as polished)



(c) Manganese distribution



(d) Sulphur distribution

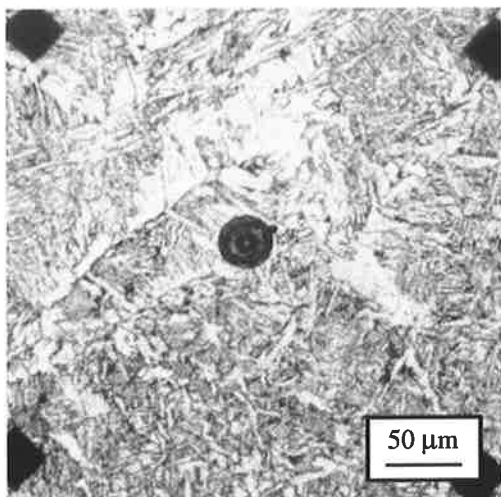


(e) Aluminium distribution

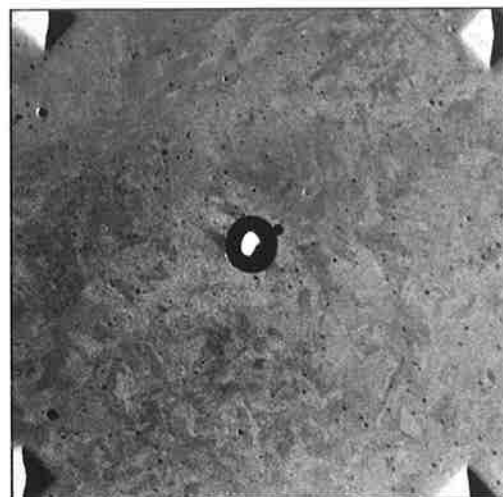


(f) Oxygen distribution

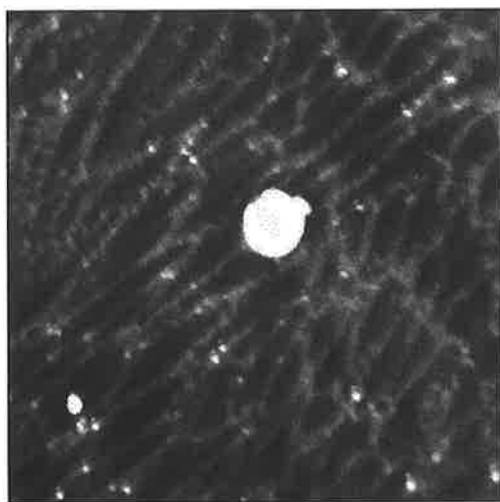
Figure 7-19: Element distributions in weld 2-2.



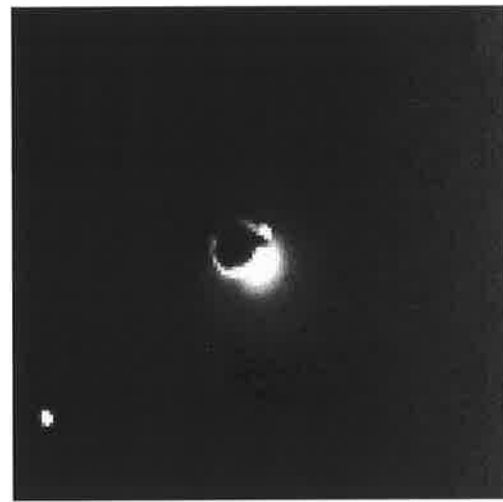
(a) Light micrograph (2% nital)



(b) SEM Backscattered image (as polished)

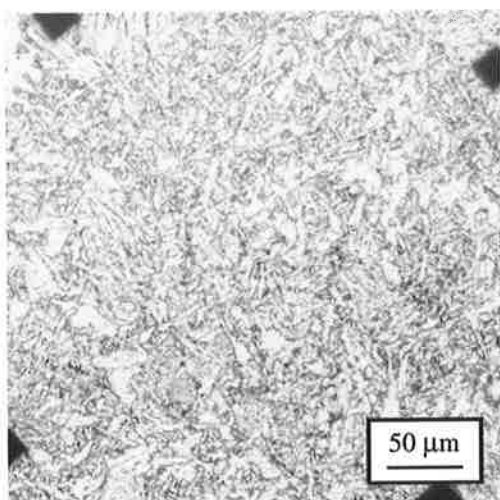


(c) Manganese distribution

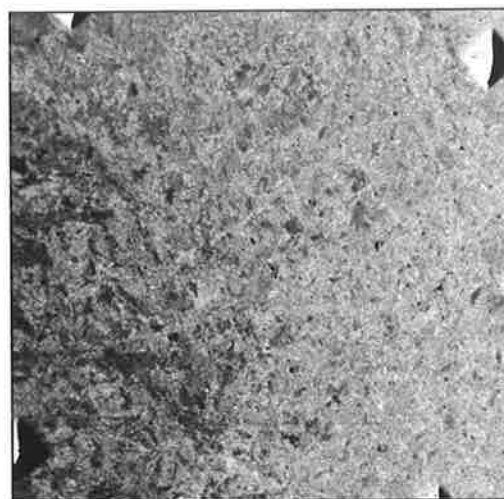


(d) Silicon distribution

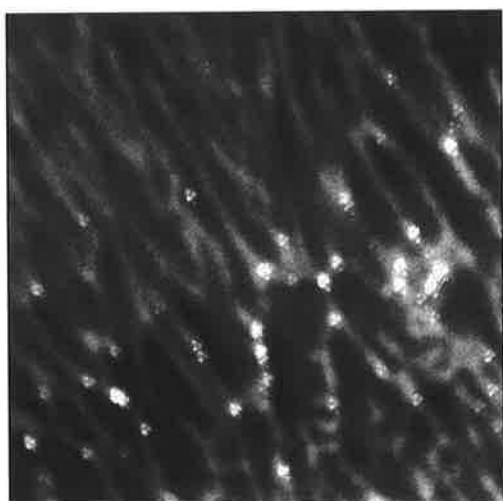
Figure 7-20: Element distributions in weld 2-3.



(a) Light micrograph (2% nital)



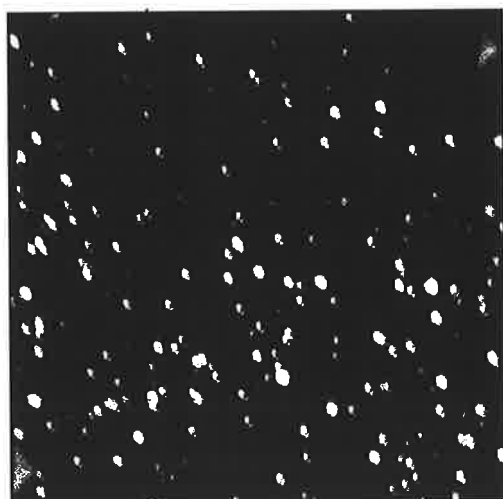
(b) SEM Backscattered image (as polished)



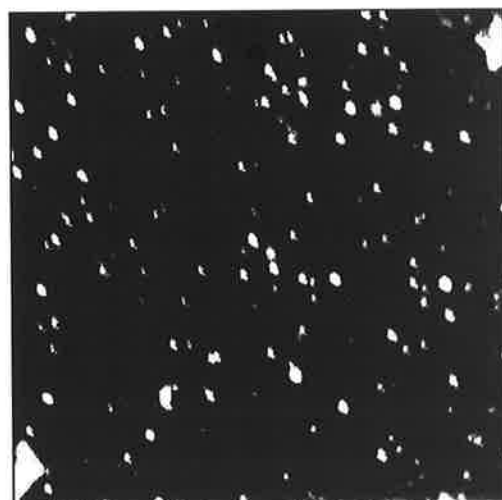
(c) Manganese distribution



(d) Sulphur distribution

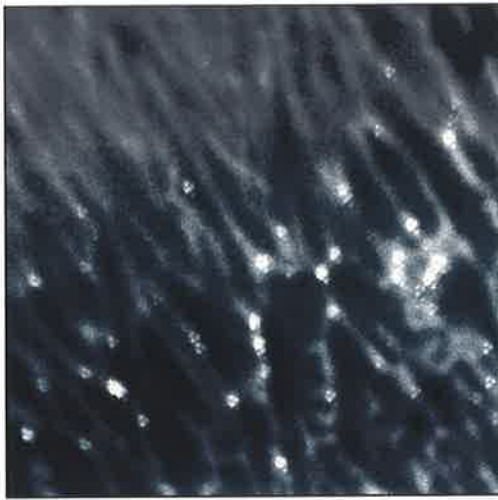


(e) Aluminium distribution

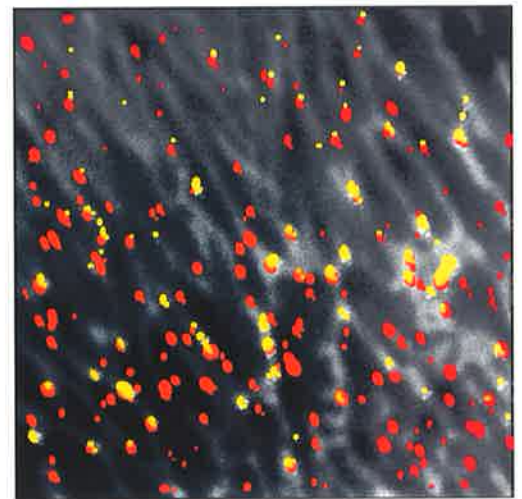


(f) Oxygen distribution

Figure 7-21: Element distributions in weld 2-4.



(a) Manganese distribution



Aluminium Sulphur

(b) Manganese distribution overlaid with Aluminium and Sulphur.



(c) Oxygen distribution



Aluminium

(d) Oxygen distribution overlaid with Aluminium.

Figure 7-22: Overlay of element distributions in weld 2-4.

7.7 SCANNING ELECTRON MICROSCOPY

7.7.1 The Effect of Silicon

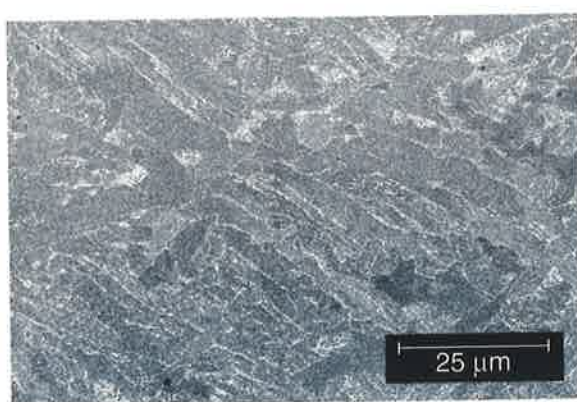
The results of scanning electron microscopy on series 1 weld deposits indicate that silicon does not appear to alter the ferrite lath size or the type of microphases formed between the ferrite laths.

In weld 1-1 (Figure 7-23) there appear to be isolated pockets of pearlite between the ferrite laths (see Figure 7-23(a), (b), and (e)). In some areas there are also continuous lengths of carbide film between the laths (Figure 7-23(d)).

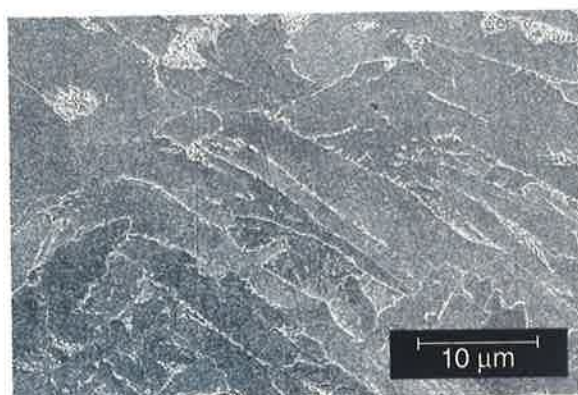
The solute banding mentioned previously in weld 1-2 is shown in Figure 7-24(a), (b), and (c). The banding region has transformed to martensite while the area adjacent has transformed to ferrite. Regions closer to the centre of the weld deposit, Figure 7-24(e) and (f), have a similar lath size and microphases to weld 1-1.

The microstructures of welds 1-3 and 1-4 shown in Figures 7-25 and 7-26 respectively are also very similar to weld 1-1.

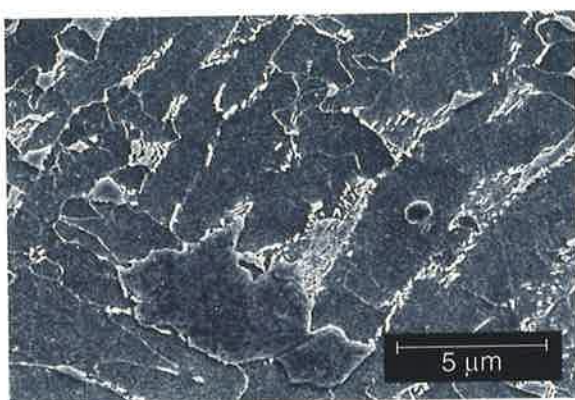
There was also a small proportion of a microphase present in all the series 1 weld deposits which had a slightly different morphology to the pearlite and carbide discussed above. This light etching blocky microphase indicated with red arrows in Figures 7-23(e), 7-24(d), 7-25(c) and 7-26(c) is generally located adjacent to or within a pearlite colony.



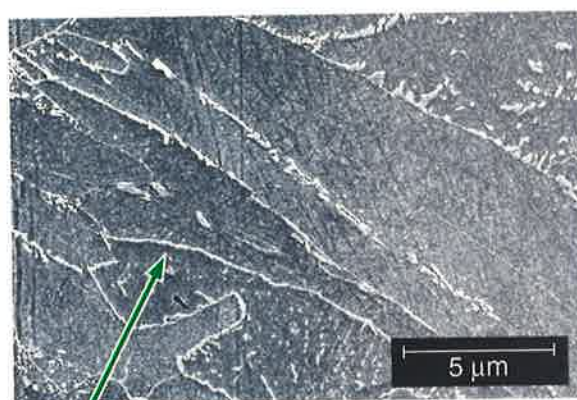
(a)



(b)

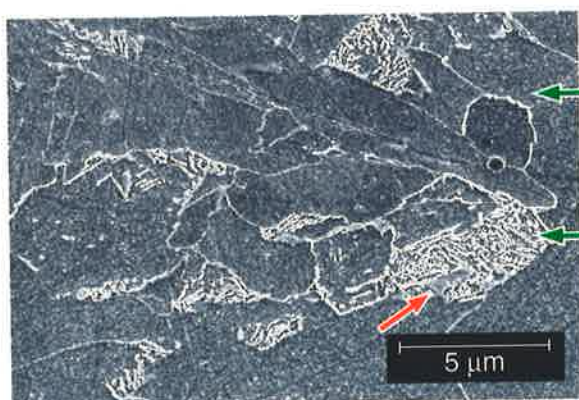


(c)



Carbide Film

(d)



(e)

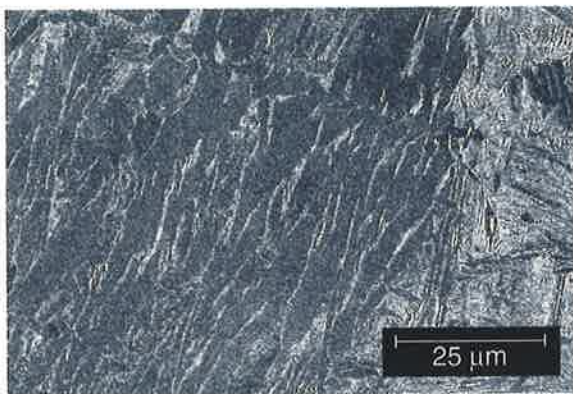
Figure 7-23: Secondary electron images from weld 1-1 (lightly etched in 2% nital).



(a)



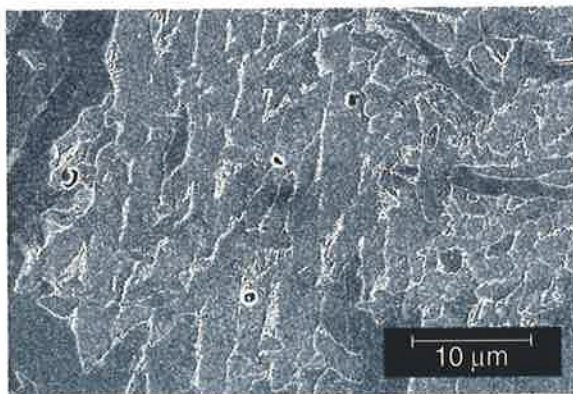
(b)



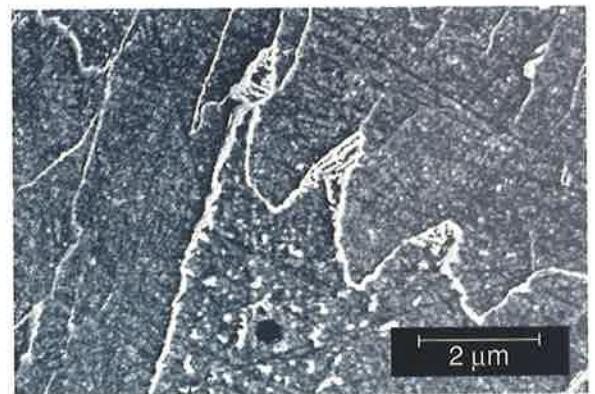
(c)



(d)

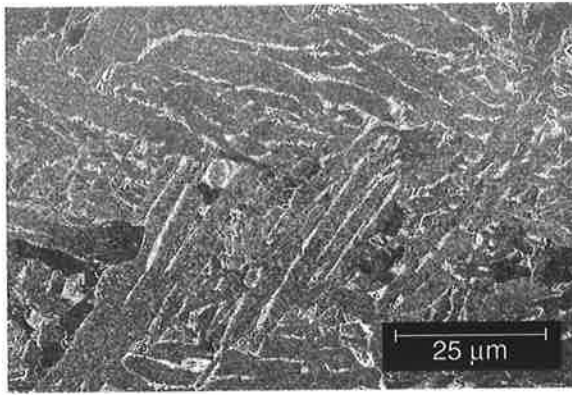


(e)

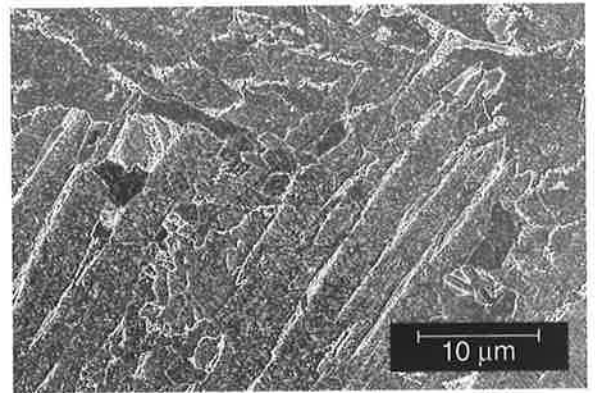


(f)

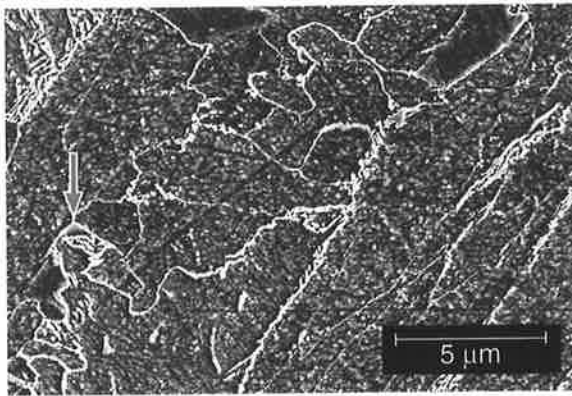
Figure 7-24: Secondary electron images from weld 1-2 (lightly etched in 2% nital).



(a)

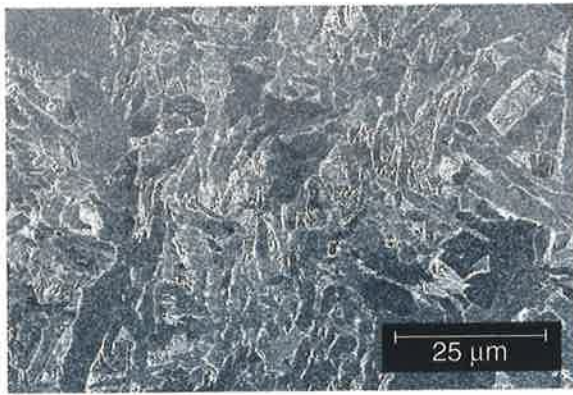


(b)

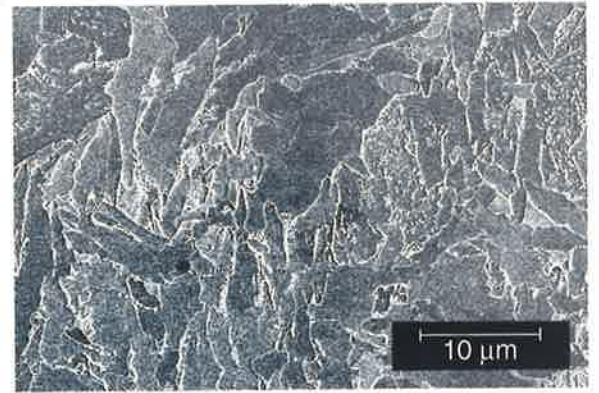


(c)

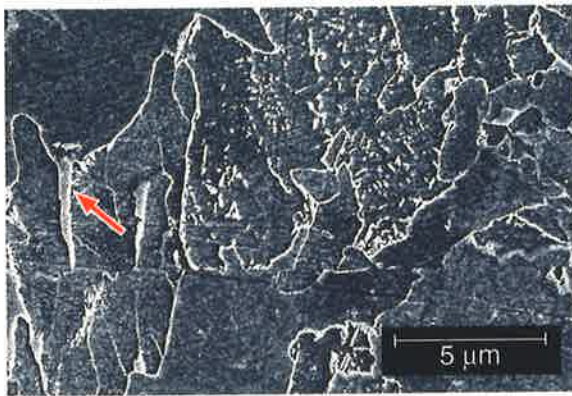
Figure 7-25: Secondary electron images from weld 1-3 (lightly etched in 2% nital).



(a)



(b)



(c)

Figure 7-26: Secondary electron images from weld 1-4 (lightly etched in 2% nital).

7.7.2 The Effect of Manganese

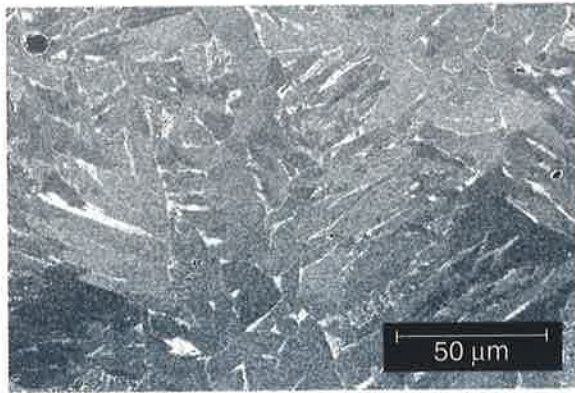
Scanning electron microscopy confirmed observations of series 2 weld deposits made using light microscopy. The microstructure varies as the manganese content increases.

The microstructure of weld 2-1 (0.53% Mn) is shown in Figure 7-27. The ferrite lath width is considerably larger than the lath widths observed in series 1 weld deposits (1.38 - 1.61% Mn). There is also a large proportion of pearlite between the ferrite laths.

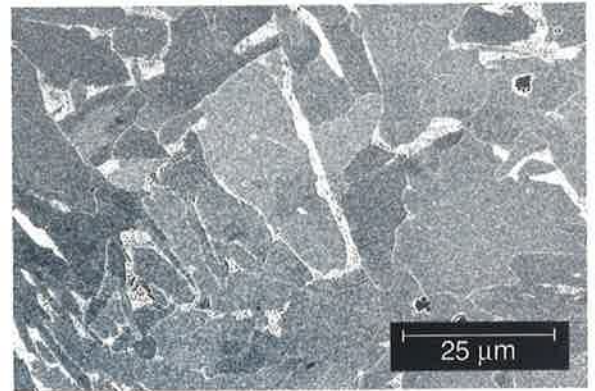
The microstructure of weld 2-2 (0.87% Mn), Figure 7-28, is slightly more refined than weld 2-1 and there is also less pearlite present between the ferrite laths.

The microstructures of welds 2-3 and 2-4 shown in Figures 7-29 and 7-30 respectively, continue to become more refined as the manganese content increases.

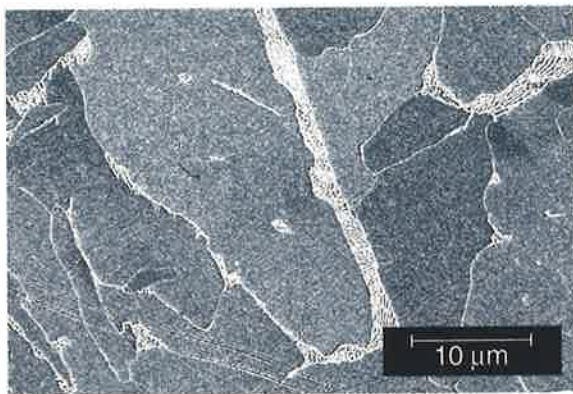
The proportion of the microphase mentioned in section 7.7.1 also increases as the manganese content increases. This microphase does not seem to be present in Figure 7-27 or Figure 7-28 (weld 2-1: 0.53% Mn, and weld 2-2: 0.87% Mn respectively). A small proportion of this microphase is present in weld 2-3 (1.15% Mn) arrowed in Figure 7-29(d), and a much larger proportion in weld 2-4 (1.71% Mn) arrowed in Figure 7-30(e).



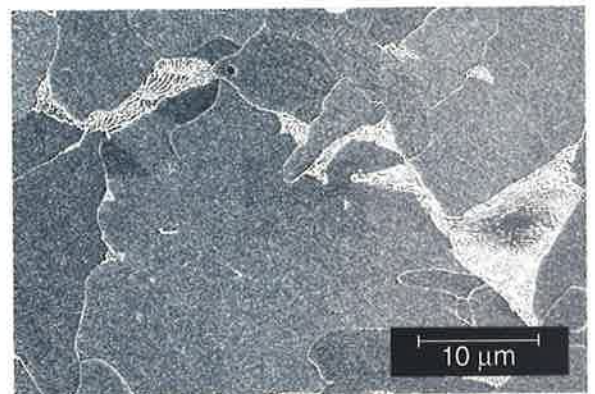
(a)



(b)

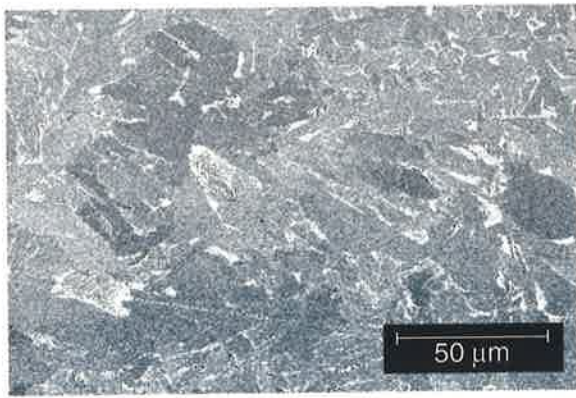


(c)

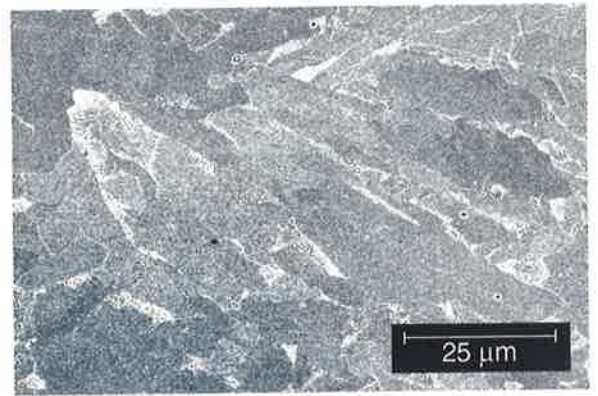


(d)

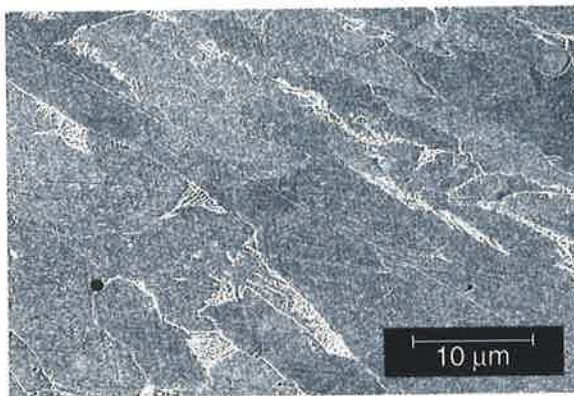
Figure 7-27: Secondary electron images from weld 2-1 (lightly etched in 2% nital).



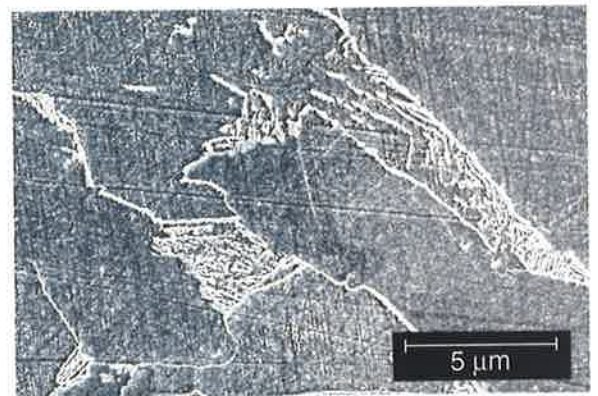
(a)



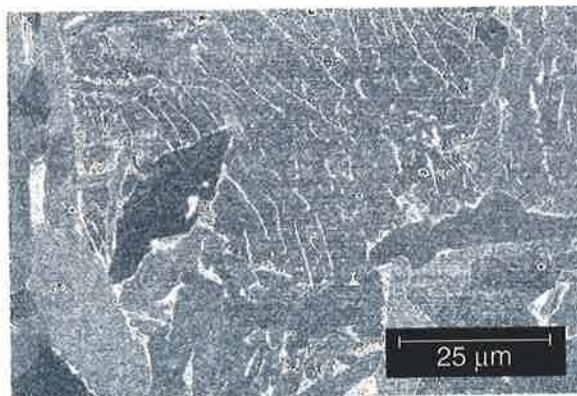
(b)



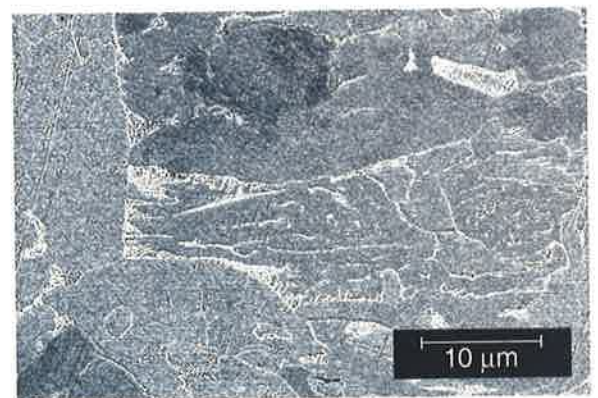
(c)



(d)



(e)

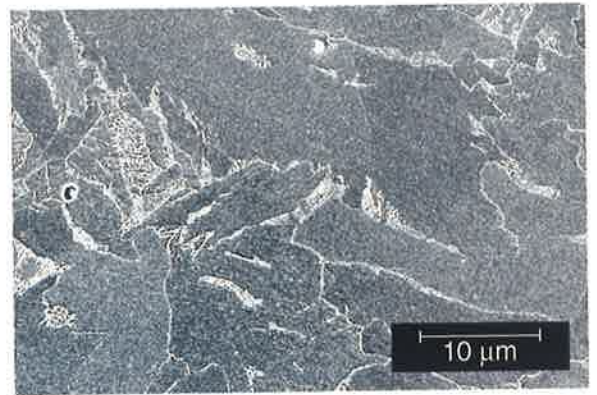


(f)

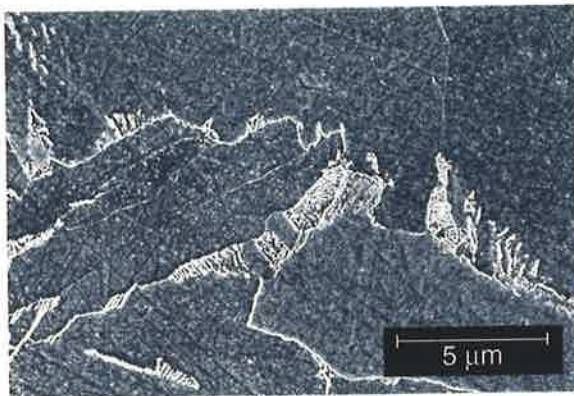
Figure 7-28: Secondary electron images from weld 2-2 (lightly etched in 2% nital).



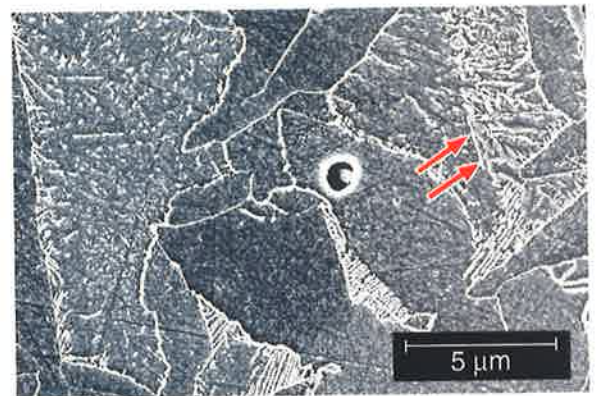
(a)



(b)



(c)



(d)

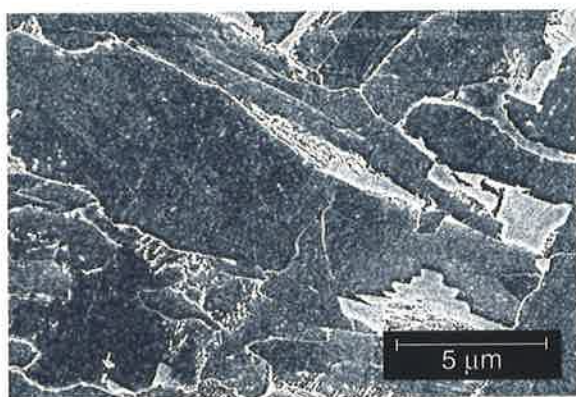
Figure 7-29: Secondary electron images from weld 2-3 (lightly etched in 2% nital).



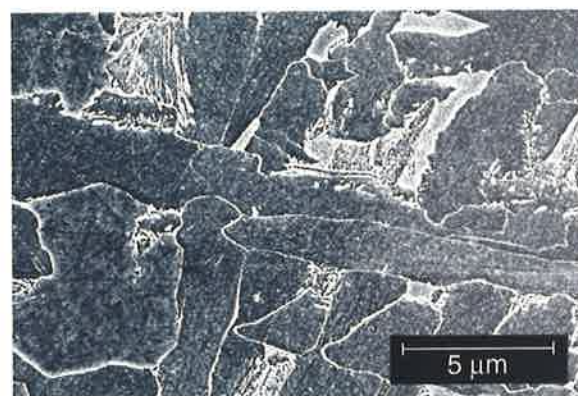
(a)



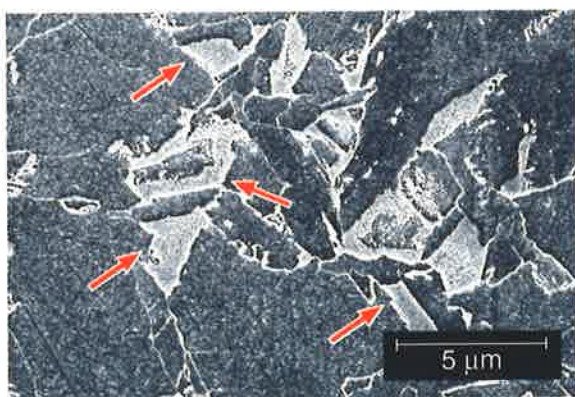
(b)



(c)



(d)



(e)

Figure 7-30: Secondary electron images from weld 2-4 (lightly etched in 2% nital).

7.8 TRANSMISSION ELECTRON MICROSCOPY

Due to limited time and access to a TEM it was decided to focus the TEM investigation on series 2 welds since this series showed a large variation in microstructure as the manganese content increased.

In weld 2-1 there was significant pearlite formation at the grain boundaries (Figure 7-31). This confirms the observation using the SEM (see Figure 7-27(c)).

In the SEM it was observed that as the manganese content increased less pearlite was observed and the grain boundaries became decorated with almost continuous thin carbide films. Examples were found of lath structures with very few precipitates (Figure 7-32(a)), and others with copious quantities of carbide (Figure 7-32(b)). The matrix was identified as ferrite using Selected Area Diffraction (SAD) (insert, Figure 7-32(a)).

In other areas, larger grains were observed (Figure 7-33(a)). This gave the diffraction pattern of Figure 7-33(b) which was indexed as austenite. Figure 7-33(c) is a Centred Dark Field (CDF) image using the $(1\bar{1}1)_\gamma$ reflection. This confirms that the blocky, light etching phase reported in section 7.7, is indeed retained austenite.

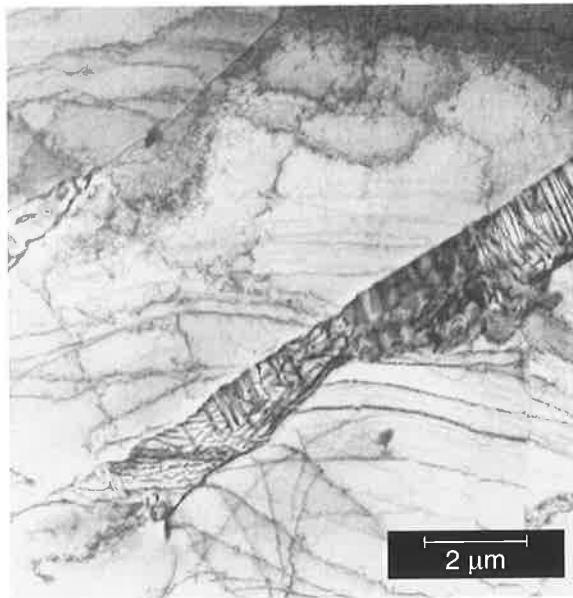
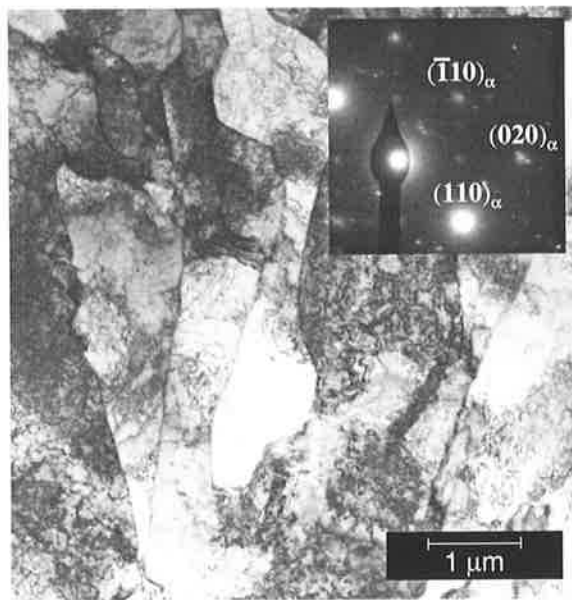
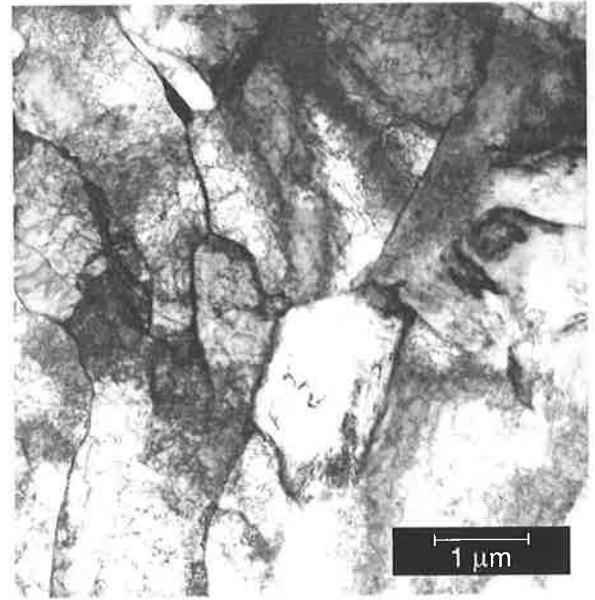


Figure 7-31: TEM BF image from weld 2-1 showing pearlite at a grain boundary.

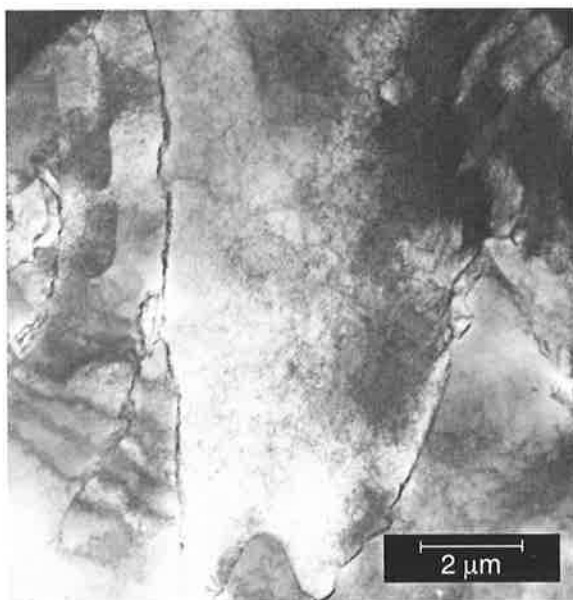


(a) BF image of ferrite with SAD pattern insert.

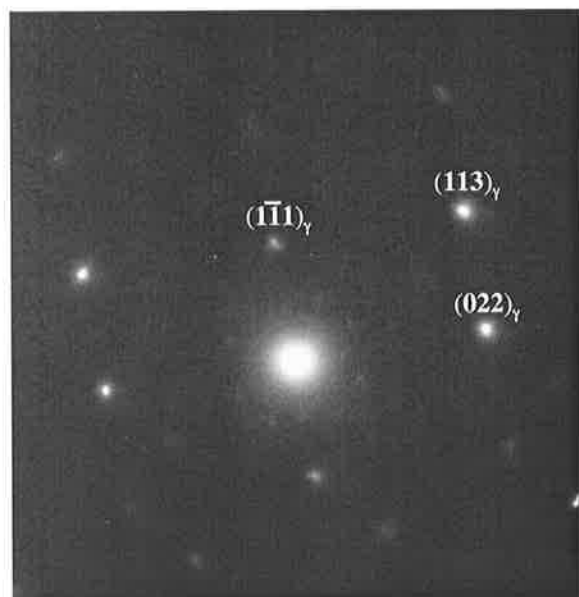


(b) BF image showing carbides at the grain boundaries.

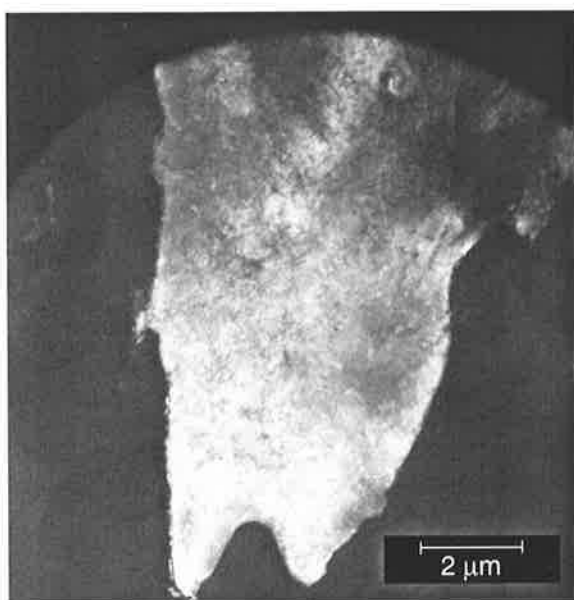
Figure 7-32: TEM BF images and SAD pattern from weld 2-4.



(a) BF image showing retained austenite.



(b) SAD pattern of (a) indexed to austenite.



(c) CDF image of (a) using $(1\bar{1}1)_\gamma$ reflection.

Figure 7-33: TEM BF image, SAD pattern and CDF image from weld 2-4 showing retained austenite.

7.9 MICROPHASES

As discussed in section 3.4.4, microphases are the result of carbon diffusing into the remaining austenite during transformation to ferrite. Light, scanning, and transmission microscopy has identified pearlite, carbide, retained austenite, and martensite (in the solute banding region) in the experimental weld deposits.

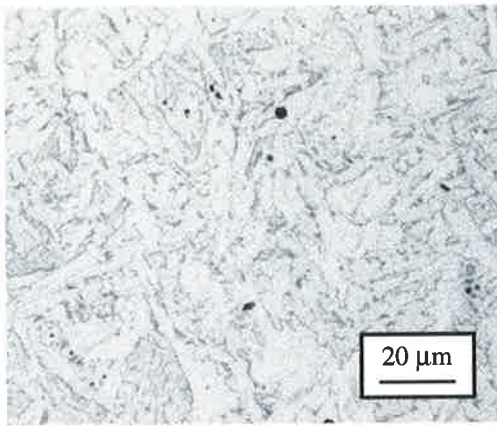
TEM identified a large region of retained austenite in sample 2-4 (see Figure 7-33). Weld 2-4 has the highest manganese content of all the experimental welds. Manganese is known as an austenite-stabilising element. It is therefore likely that increased levels of manganese will result in larger amounts of retained austenite in the transformed microstructure.

It was observed whilst performing WDX mapping that some regions within backscattered images of high manganese weld deposits were brighter (ie gave a stronger signal) relative to the surrounding matrix. The presence of these bright areas only in the high manganese welds suggests that they also represent retained austenite. It may be possible that the close packed FCC structure of austenite could have a higher backscatter coefficient than BCC ferrite of similar chemical composition.

To investigate the presence of retained austenite further, a region in weld 2-4 (1.71% Mn) was examined in greater detail and the results are displayed in Figure 7-34. The region investigated lies within the area previously examined using WDX mapping (see Figure 7-21). Figure 7-34(a) shows the region as viewed under the light microscope after light etching in 2% nital. A backscattered image of the same region is shown in Figure 7-34(c) and the manganese distribution of the same region is shown in Figure 7-34(d). In Figure 7-34(e) the bright areas (retained austenite) on the backscattered image have been coloured

yellow. Figure 7-34(f) displays the manganese distribution overlayed with retained austenite areas and shows that the majority of retained austenite is located in the manganese-rich cell boundaries.

Svensson and Gretoft [79] also reported the formation of microphases in the manganese-rich cell boundaries, and with increasing manganese content the microphases became more segregated.



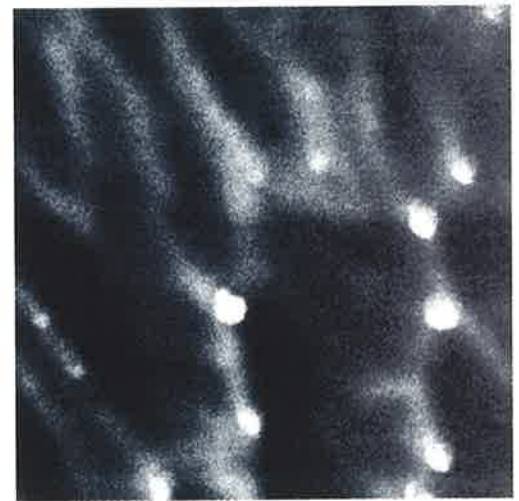
(a) Light micrograph (2% nital)



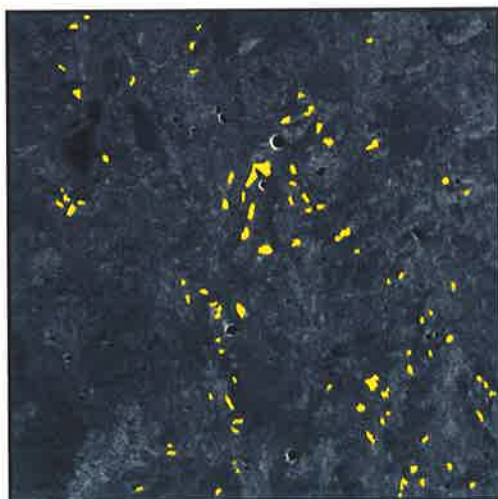
(b) Manganese Distribution



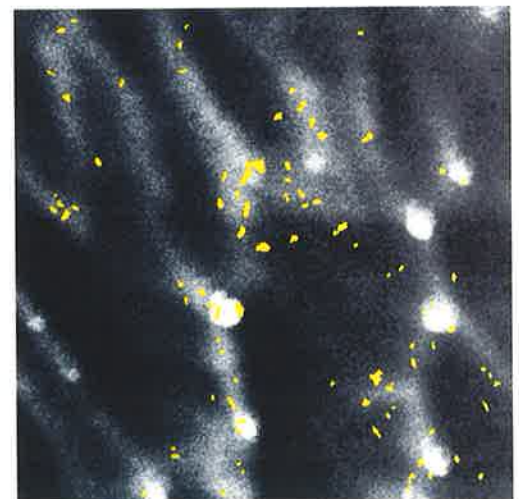
(c) Backscattered image of the area in (a) and the red square in (b)



(d) Manganese distribution of region outlined by the red square in (b).



(e) Backscattered image with retained austenite highlighted in yellow.



(f) Manganese distribution overlaid with retained austenite (yellow).

Figure 7-34: Location of retained austenite in weld 2-4.

7.10 HARDNESS TESTING RESULTS

The results of hardness testing are tabulated in Table 7-6 and plotted as a functions of silicon content in Figure 7-35 and manganese content in Figure 7-36.

Figure 7-35 indicates that there was a very slight increase in hardness as the silicon content increased. Similar results have been reported by Evans [145]. Figure 7-36 shows that manganese content has a significant influence on weld hardness. The hardness increases from 160 HV5 at 0.53% Mn to 205 HV5 at 1.71% Mn. In the manganese composition range studied, hardness is directly proportional to the manganese content.

Weld No.	Hardness (HV5)					
	Reading No.					Average & 95% Confidence Interval
	1	2	3	4	5	
1-1	212	208	180	197	203	200 $\pm$ 10
1-2	201	201	206	193	193	199 $\pm$ 4
1-3	199	197	199	206	201	200 $\pm$ 3
1-4	214	212	210	214	221	214 $\pm$ 3
2-1	152	164	160	161	164	160 $\pm$ 4
2-2	169	167	157	167	165	165 $\pm$ 4
2-3	180	175	168	171	169	173 $\pm$ 4
2-4	201	201	206	208	210	205 $\pm$ 3

Table 7-6: Results of hardness testing on experimental weld deposits.

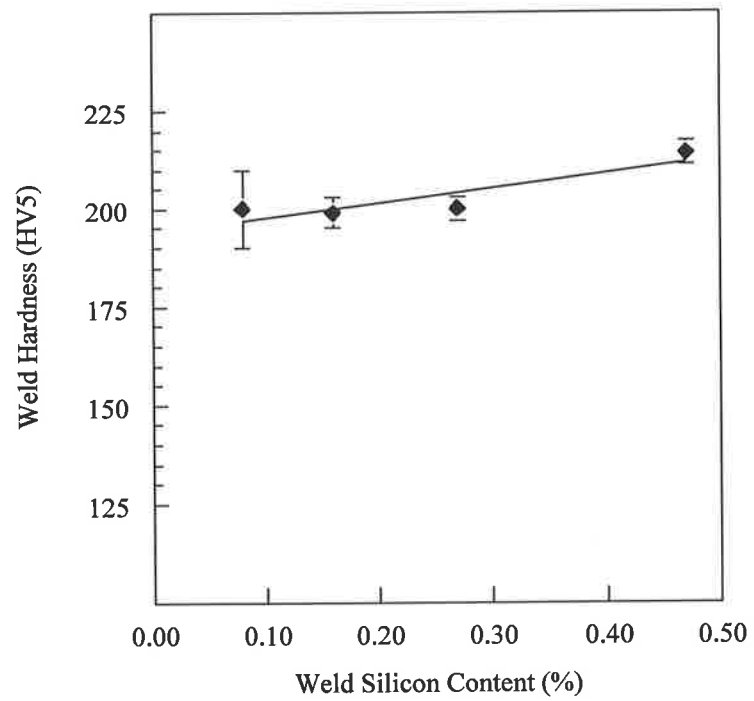


Figure 7-35: Weld hardness as a function of silicon content.

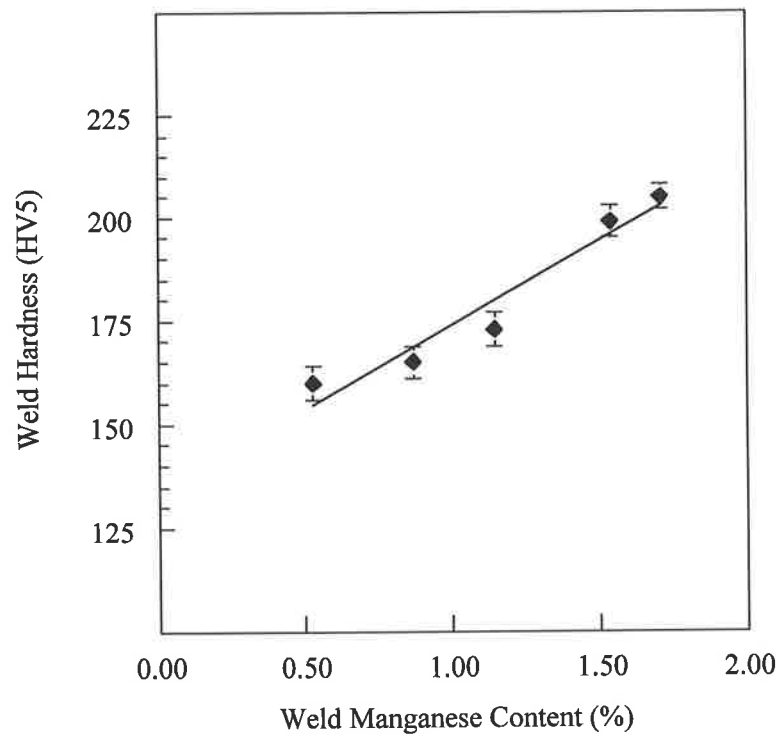


Figure 7-36: Weld hardness as a function of manganese content.

7.11 CHARPY V-NOTCH TESTING

The results of Charpy testing are listed in Table 7-7 and plotted in Figure 7-37 (the effect of silicon) and Figure 7-38 (the effect of manganese). There was considerable scatter amongst the results both within a weld and as part of a series which has resulted in large confidence intervals. This precluded the determination of any obvious trends.

In an attempt to understand the large variation in results selected broken samples (indicated in Table 7-7 by [†]) were examined firstly in the SEM then sectioned and prepared for light microscopy.

The fracture surface of sample 1-1b (100 J/cm²) is displayed in Figure 7-39. In Figure 7-39(a) shear lips are evident indicating ductile failure. Figure 7-39(b) is a high

Weld No.	Absorbed Energy (J/cm ²) at 0°C			
	Sample No.			Average & 95% Confidence Interval
	A	B	C	
1-1	64	100 [†]	56	73.3 ± 22
1-2	44 [†]	74	40	52.7 ± 17
1-3	88	98	126 [†]	104.0 ± 18
1-4	20 [†]	50	24	31.3 ± 15
2-1	64	40	88 [†]	64.0 ± 22
2-2	72	86	24 [†]	60.7 ± 30
2-3	56	64	56 [†]	58.7 ± 4
2-4	30	16 [†]	86 [†]	44.0 ± 34

[†] Fracture surface examined in the SEM and light microscope.

Table 7-7: Charpy V Notch test results.

magnification image of the fracture surface close to the notch, and shows a typical ductile fracture surface with microvoid coalescence. Figure 7-39(c) is a light micrograph showing the microstructure beneath the fracture surface and the crack path close the notch. Below the fracture surface a number of voids are visible and the linking of these voids results in the fracture surface observed in Figure 7-39(b).

The fracture surface of sample 1-2a (44 J/cm^2) is shown in Figure 7-40. In contrast to the previous sample there are no shear lips evident in Figure 7-40(a) and the fracture surface, shown in detail in Figure 7-40(b), has large cleavage facets. In Figure 7-40(c) a few microvoids are visible below the fracture surface (arrowed), as are surface cracks. The fracture path appears to be made by the joining of several cleavage facets. From Figure 7-40(c) it appears that cleavage cracking is not limited to a particular type of phase and crosses grain boundaries without impedance. However, it should be noted that the cracks are three dimensional and viewing in a two dimensional plane can be misleading.

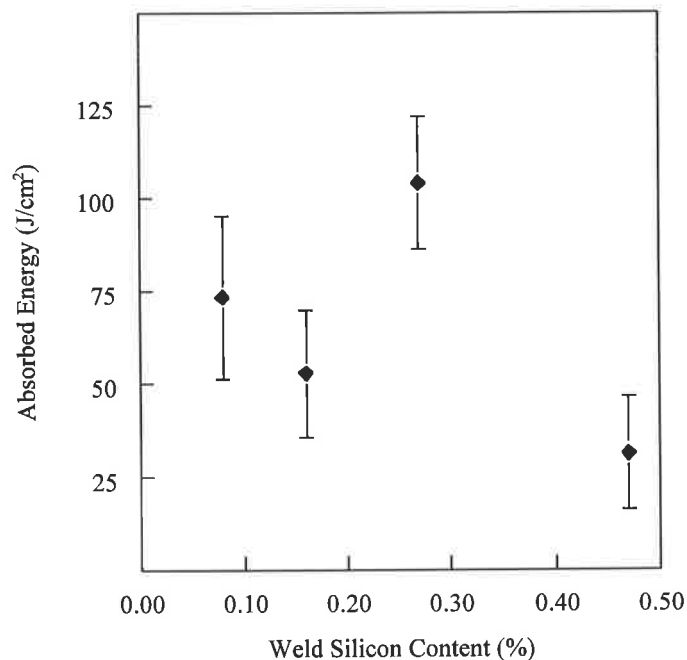


Figure 7-37: Absorbed energy as a function of weld silicon content.

The fracture surface of 1-3c (126 J/cm^2), shown in Figure 7-41 is similar to sample 1-1b with shear lips and a ductile fracture surface.

The fracture surface of sample 1-4a (20 J/cm^2) shown in Figure 7-42 has a similar cleavage fracture appearance to sample 1-2a described above. The cleavage facets observed in Figure 7-42 appear to be larger than those observed in sample 1-2a and could explain the lower absorbed energy in sample 1-4a.

Although sample 2-1 had a coarse microstructure (see Figure 7-9), the weld deposit had acceptable toughness. Sample 2-1c (88 J/cm^2) presented in Figure 7-43, had a ductile fracture surface. The improved toughness may be the result of the reduced hardness in the weld deposit.

An examination of the fracture surface of sample 2-2c, (Figure 7-44) revealed a large gas pore near the notch explaining the low absorbed energy result.

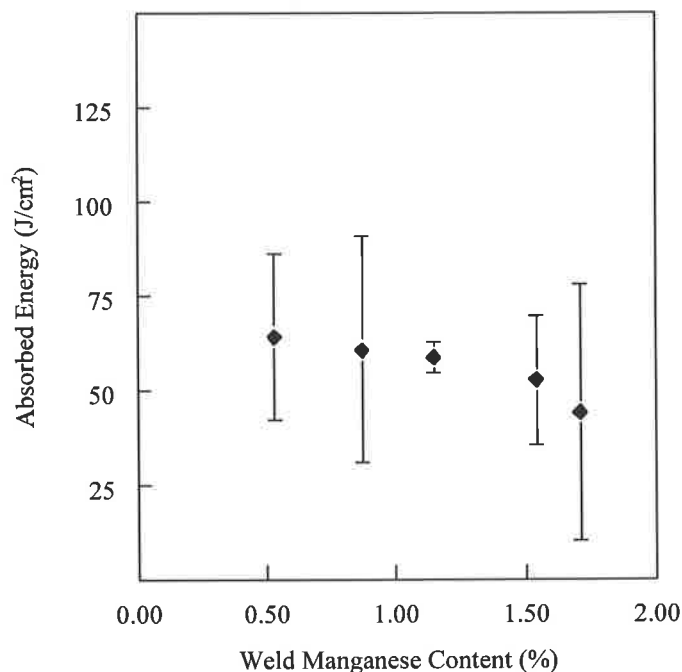
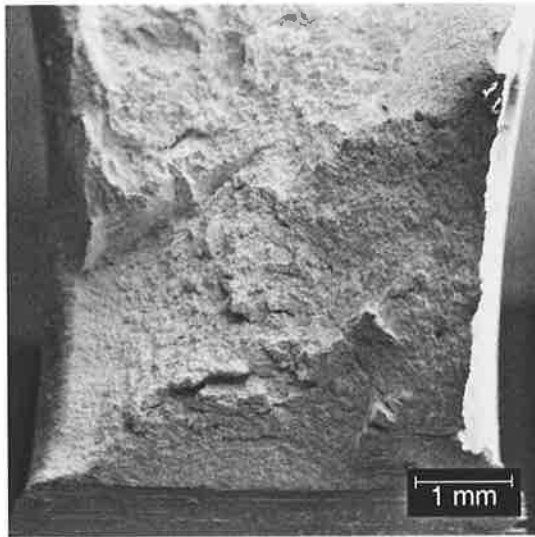


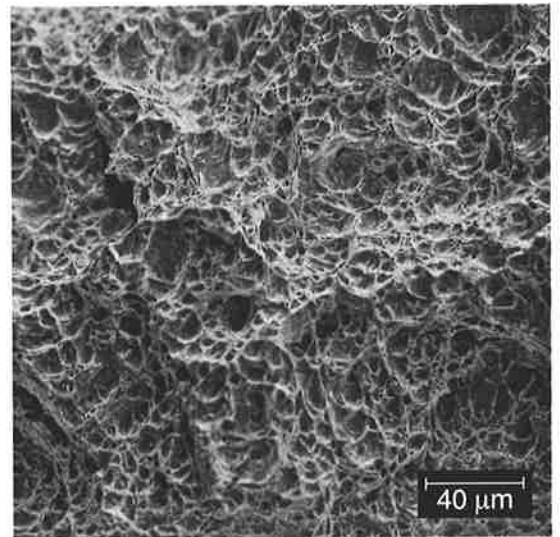
Figure 7-38: Absorbed energy as a function of weld manganese content.

Weld 2-3 (1.15% Mn) displayed the most consistent Charpy V-Notch test results with an average of $58.7 \pm 4 \text{ J/cm}^2$. Figure 7-45 shows the fracture surface of sample 2-3c (56 J/cm^2). From Figure 7-45(b), the fracture surface looks to have mainly cleavage facets. However, examination of Figure 7-45(c) indicates that parts of the fracture surface display a ductile appearance. In this sample the cleavage facets have been linked by ductile microvoid coalescence.

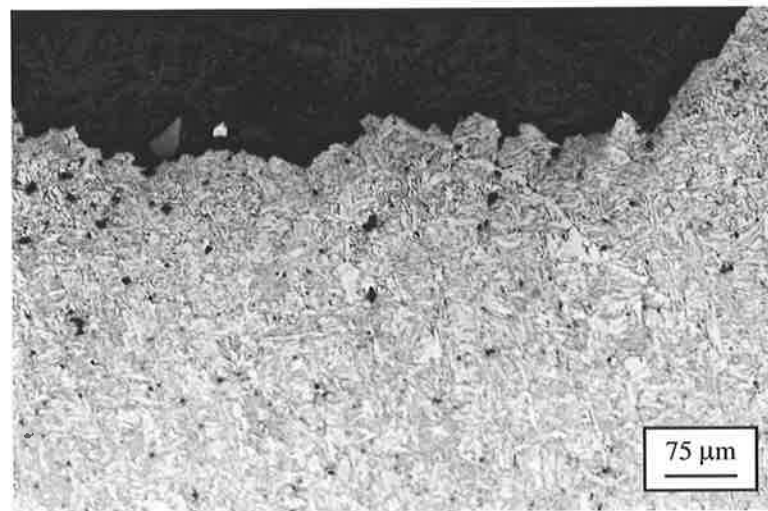
Due to the large variation in results for weld 2-4 (1.71% Mn) two Charpy V-Notch samples were examined. Figure 7-46 and Figure 7-47 show the fracture surfaces of samples 2-4b (16 J/cm^2) and 2-4c (86 J/cm^2). Sample 2-4b does not appear to have any defects such as porosity and displays a typical cleavage facet fracture surface. However, the fracture surface of sample 2-4c contains some areas of cleavage facets and some areas of microvoid coalescence. Therefore the large variation in absorbed energies indicates that the test temperature (0°C) is near the ductile to brittle transition temperature.



(a) SEM

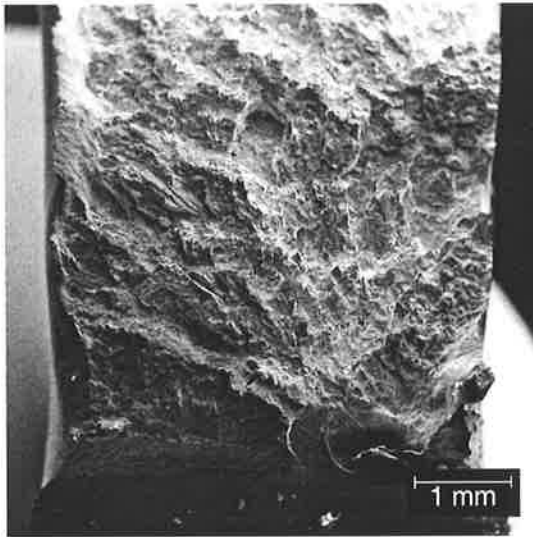


(b) SEM

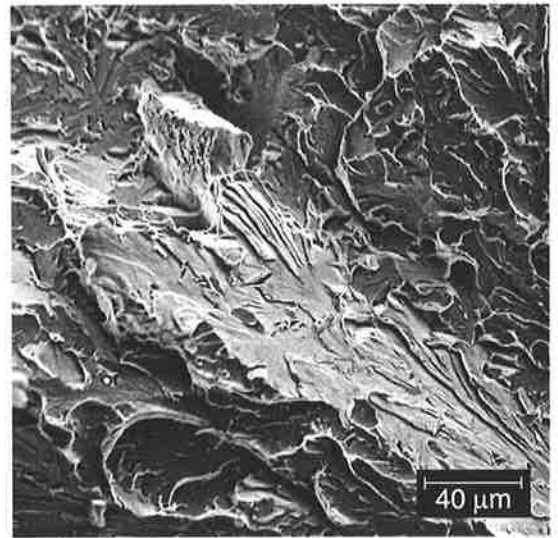


(c) Light micrograph

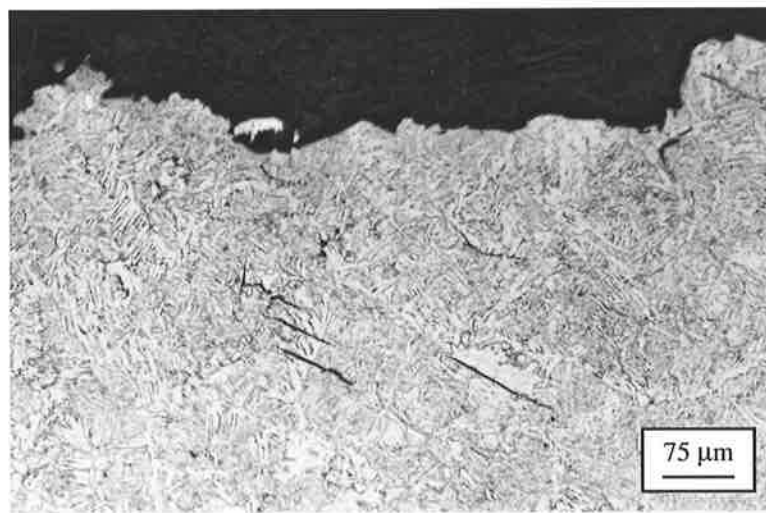
Figure 7-39: Examination of weld 1-1b Charpy fracture surface (100 J/cm^2).



(a) SEM

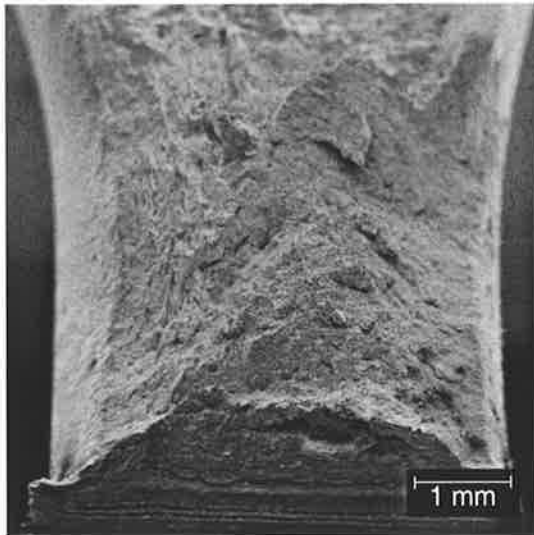


(b) SEM

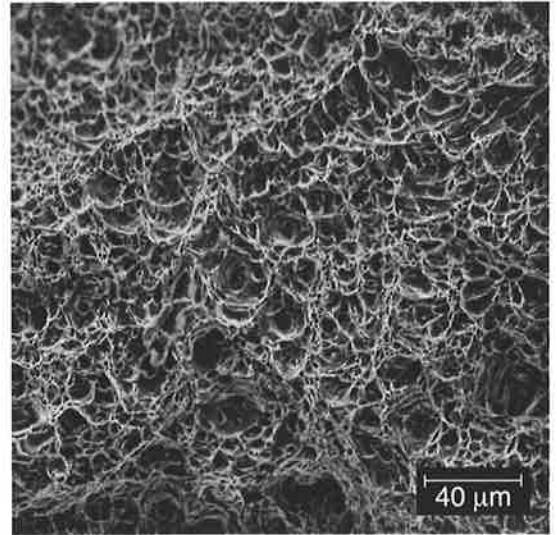


(c) Light micrograph

Figure 7-40: Examination of weld 1-2a Charpy fracture surface (44 J/cm^2).



(a) SEM

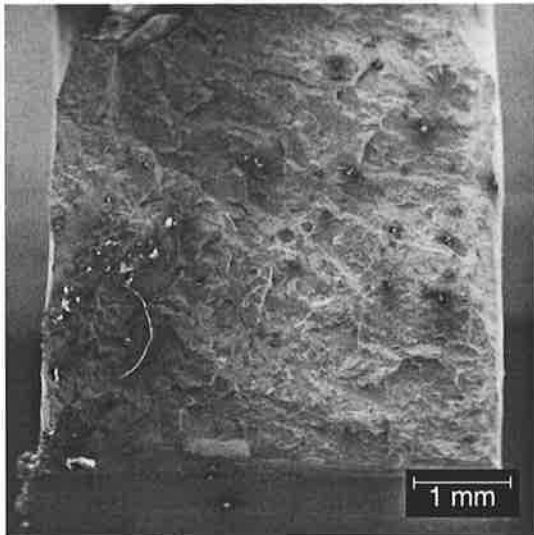


(b) SEM

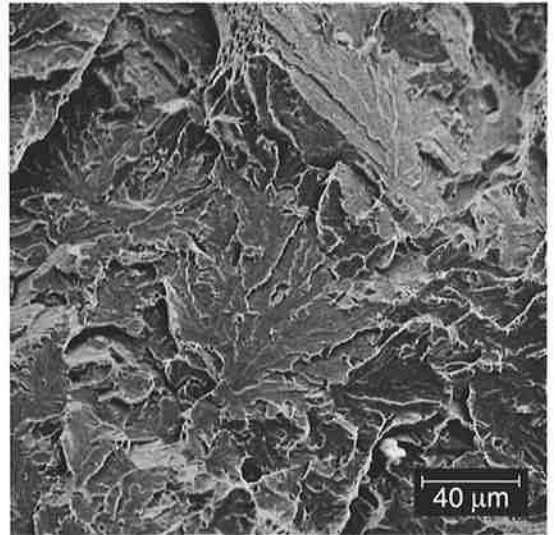


(c) Light micrograph

Figure 7-41: Examination of weld 1-3c Charpy fracture surface (126 J/cm^2).



(a) SEM

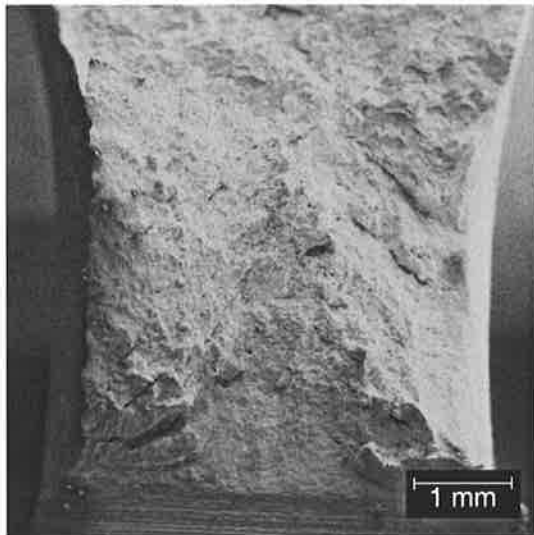


(b) SEM

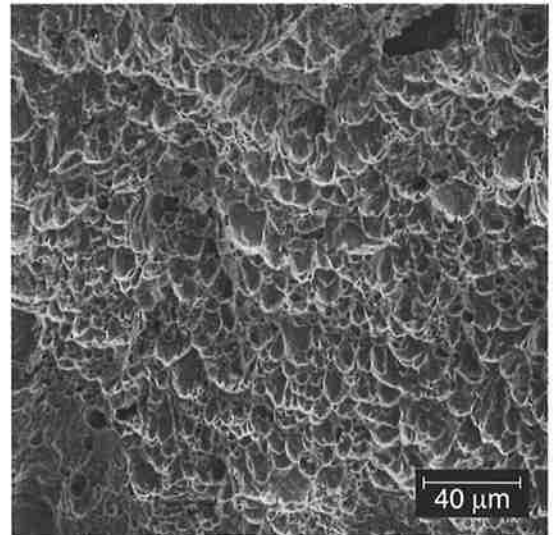


(c) Light micrograph

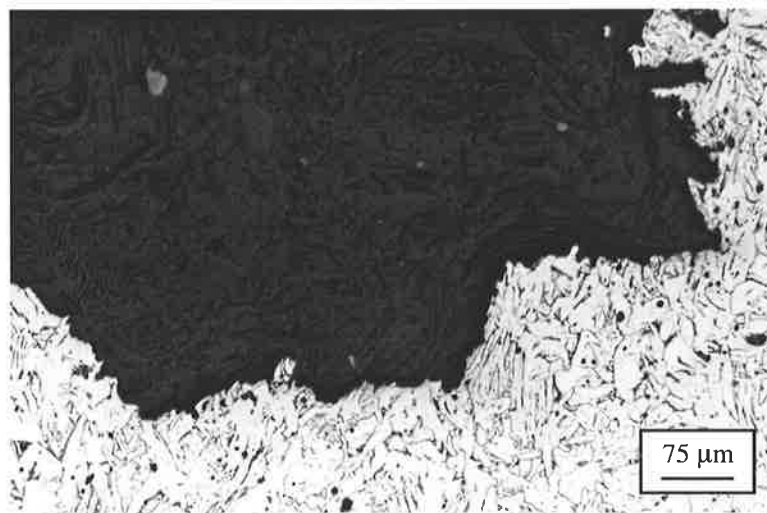
Figure 7-42: Examination of weld 1-4a Charpy fracture surface (20 J/cm^2).



(a) SEM

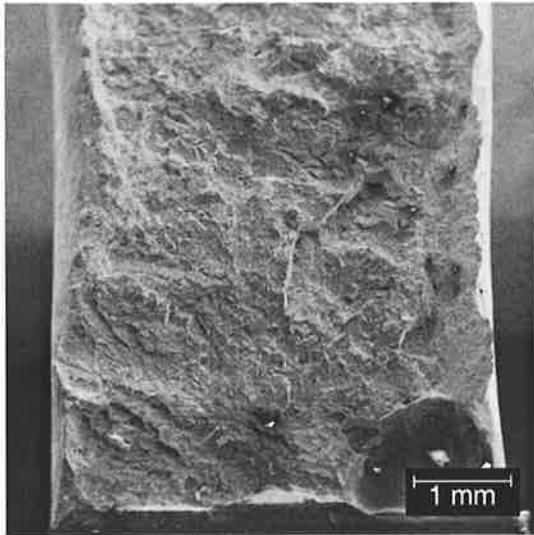


(b) SEM

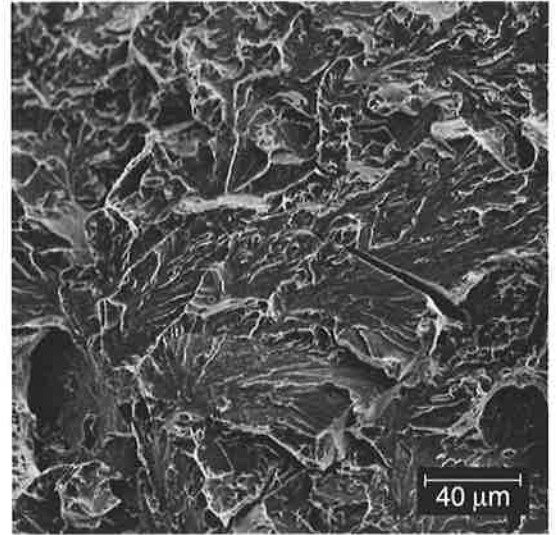


(c) Light micrograph

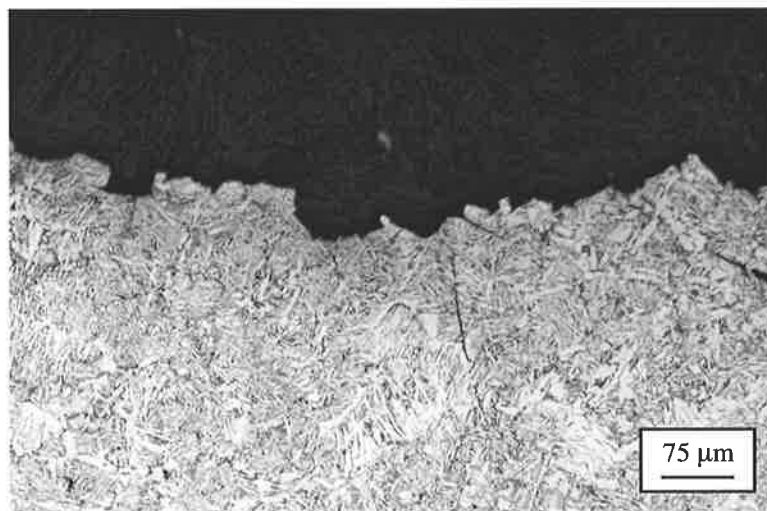
Figure 7-43: Examination of weld 2-1c Charpy fracture surface (88 J/cm^2).



(a) SEM

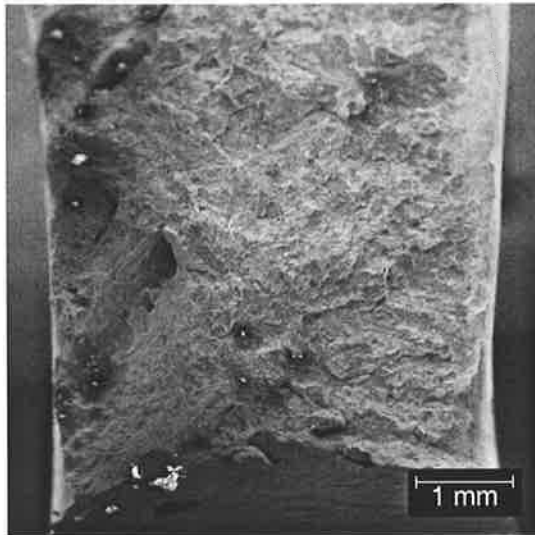


(b) SEM

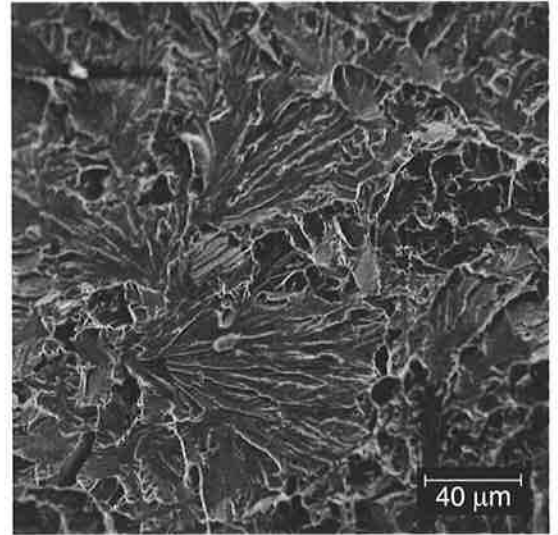


(c) Light micrograph

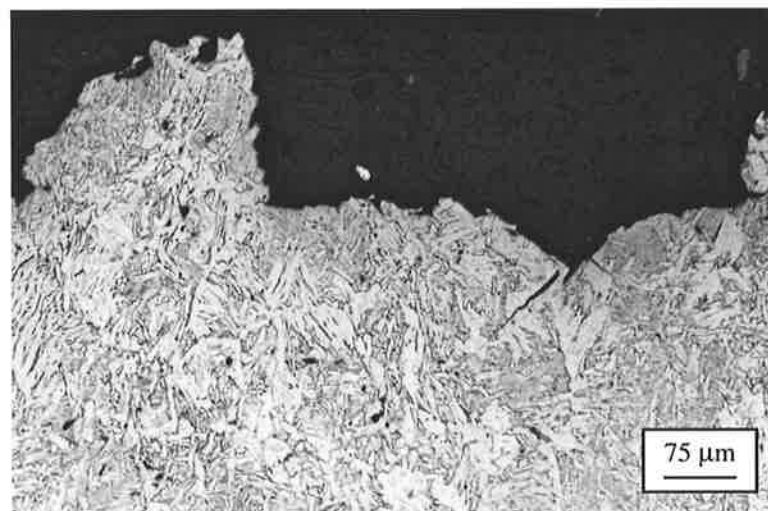
Figure 7-44: Examination of weld 2-2c Charpy fracture surface (24 J/cm^2).



(a) SEM

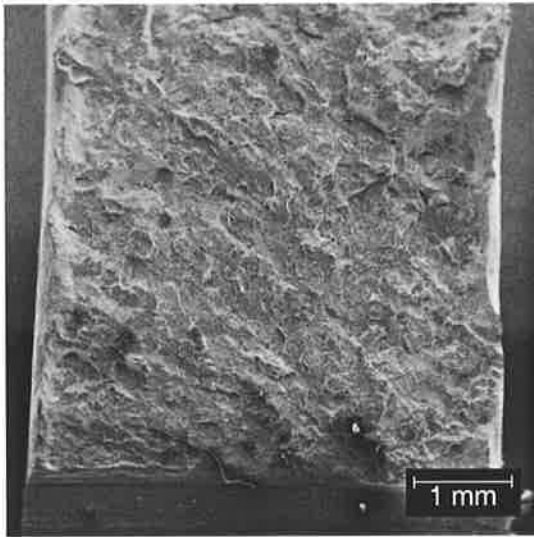


(b) SEM

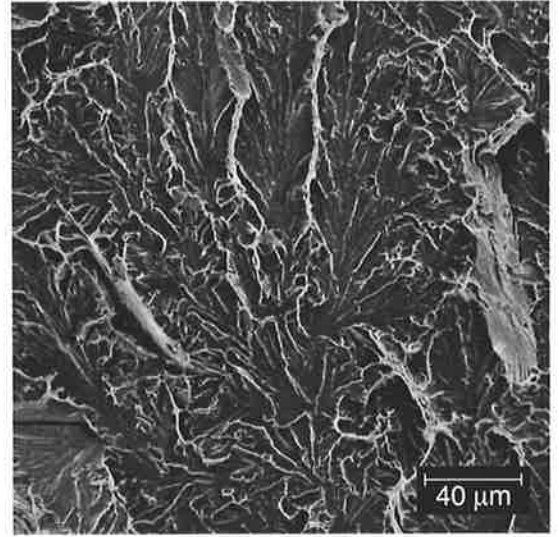


(c) Light micrograph

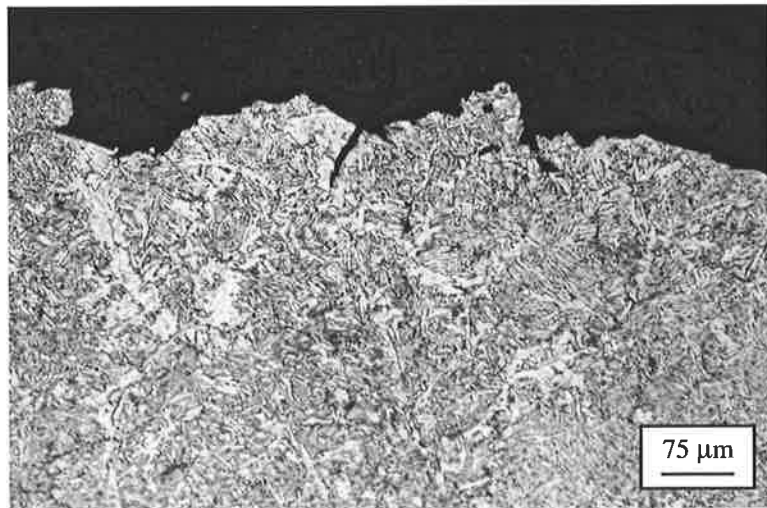
Figure 7-45: Examination of weld 2-3c Charpy fracture surface (56 J/cm^2).



(a) SEM

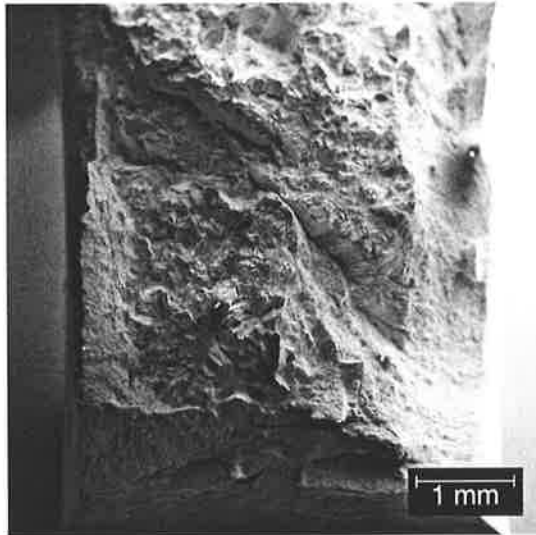


(b) SEM

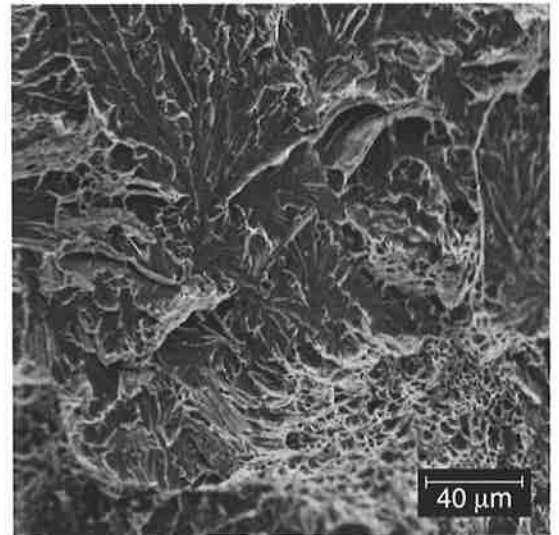


(c) Light micrograph

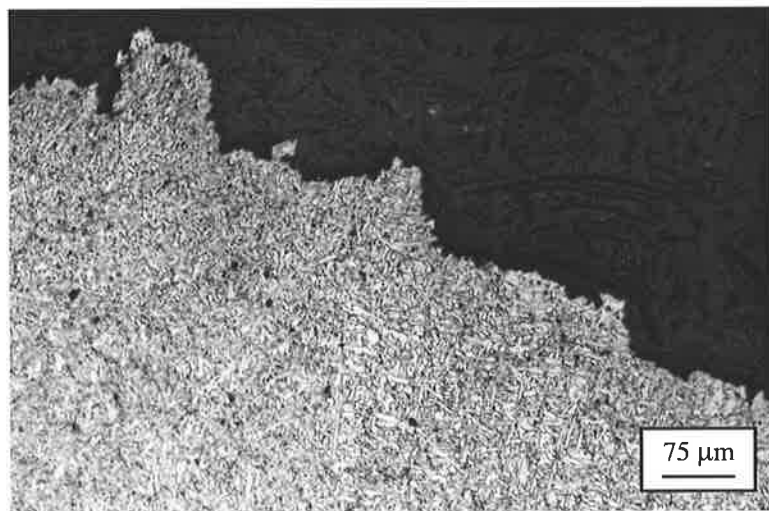
Figure 7-46: Examination of weld 2-4b Charpy fracture surface (16 J/cm^2).



(a) SEM



(b) SEM



(c) Light micrograph

Figure 7-47: Examination of weld 2-4c Charpy fracture surface (86 J/cm^2).

7.12 TENSILE TESTING

The results of tensile testing are listed in Table 7-8 and plotted in Figures 7-48 and 7-49.

For comparison purposes tensile tests were also conducted on grade 250 base material.

Load vs Strain graphs plotted from test data files have been reproduced in Appendix D.

Weld No.	0.2% Proof Stress (MPa)	Ultimate Tensile Strength (MPa)	Elongation (%)	Fracture surface examined	Comment
Grade 250a	304	458	36		
Grade 250b	279	463	37		
Average	292	461	37		
1-1a	485	604	24		
1-1b	494	621	21	✓	Small inclusions
Average	490	612	23		
1-2a	473	634	21		
1-2b	509	648	25		
Average	491	641	23		
1-3a	509	658	22		
1-3b	502	645	21	✓	Small inclusions
Average	506	652	22		
1-4a	509	629	11	✓	Large gas pores
1-4b	508	635	11		Large gas pores
Average	509	632	11		
2-1a	389	494	14	✓	Lack of penetration
2-1b	385	517	23		Lack of penetration
Average	387	506	18		
2-2a	409	543	30		
2-2b	414	545	30		
Average	412	544	30		
2-3a	465	586	17	✓	Small inclusions
2-3b	450	585	25		
Average	458	586	21		
2-4a	496	639	22		
2-4b	500	652	23	✓	
Average	498	645	23		

Table 7-8: Results from tensile testing on grade 250 steel and weld deposits.

Table 7-8 also notes any defects observed on the fracture surface. Some of these tensile samples containing defects were examined in the SEM. Backscattered images of the fracture surface of selected samples are presented in Figure 7-50.

From Figure 7-48 it is evident that the addition of silicon appears to have little effect on the tensile strength of the weld deposit. However since the tensile strength of weld 1-4 was reduced by the presence of gas pores in the sample, there may be a trend to some increase in tensile strength with silicon content. The gas pores also had a marked effect on elongation which dropped to 11% compared with ~22% for the other welds in series 1. The effect on elongation is also evident by comparing Figure 7-50(c) and Figure 7-50 (b).

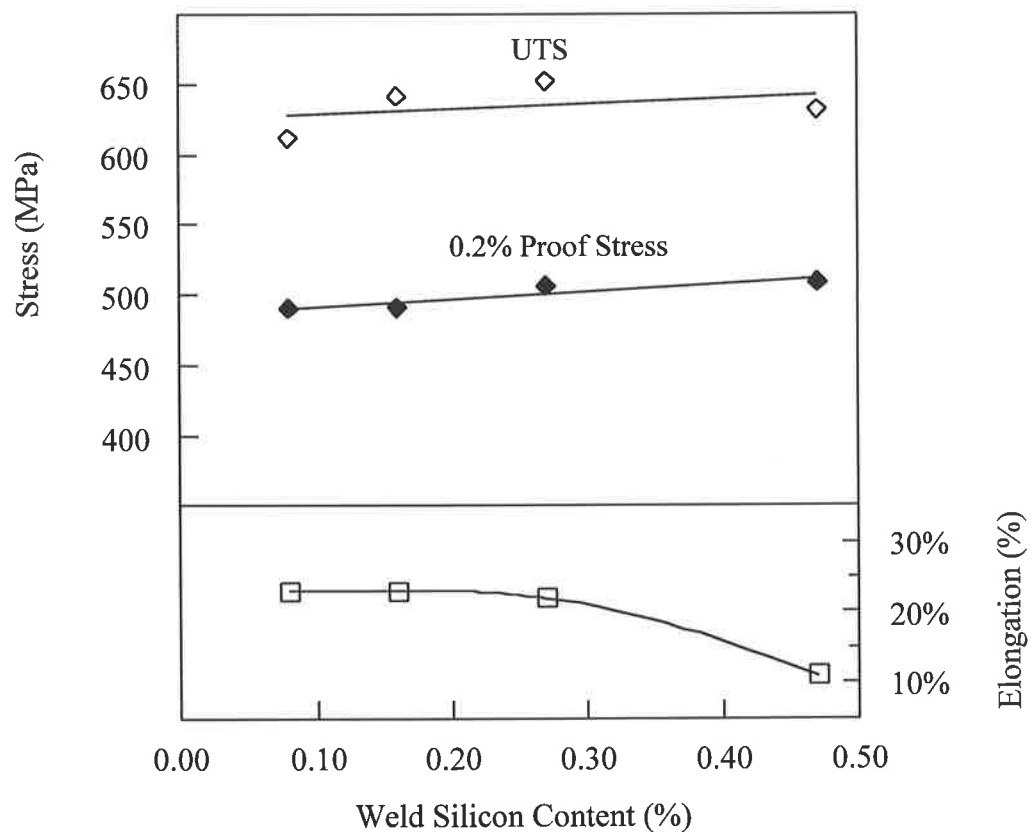


Figure 7-48: The effect of silicon content on weld tensile properties.

In Figure 7-49 tensile properties are plotted as a function of the weld manganese content. Both 0.2% proof stress and ultimate tensile strength increase with increasing manganese content. Other researchers have reported similar findings [142]. The two tensile samples extracted from weld 2-1 contained defects which is reflected in the lower elongation values.

The tensile properties of both series correlate well with the hardness values reported in section 7.10. As expected the tensile strength increases as the hardness increases.

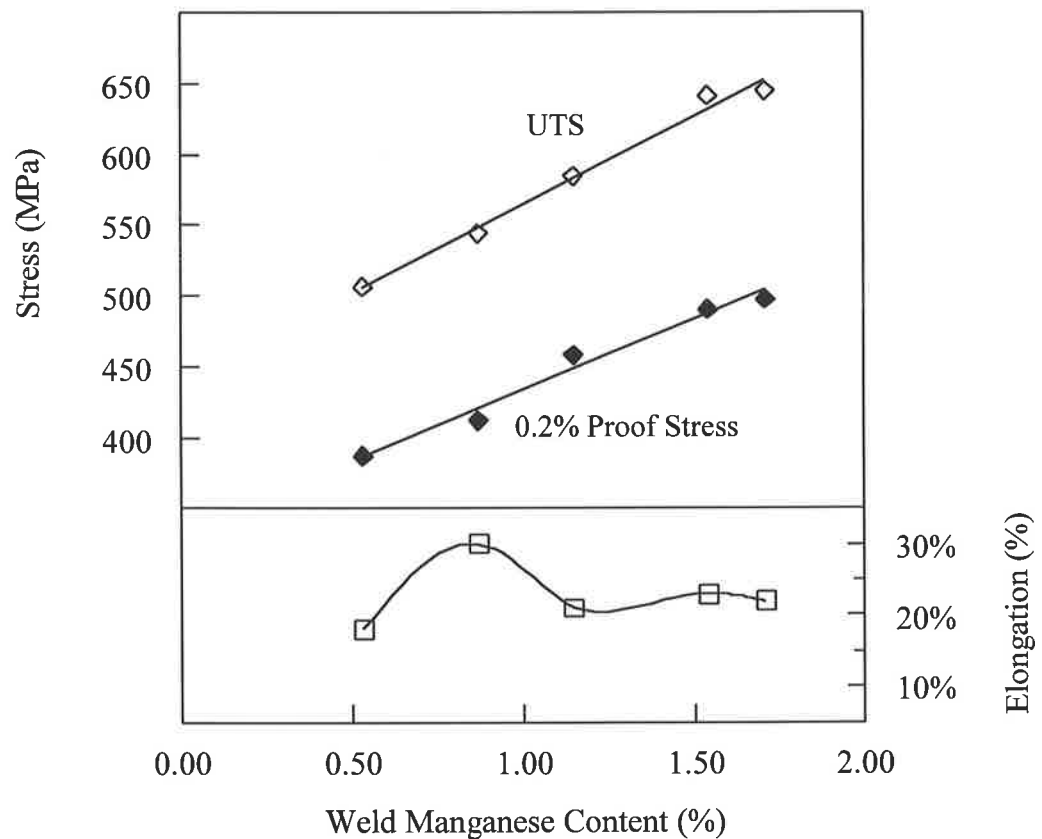
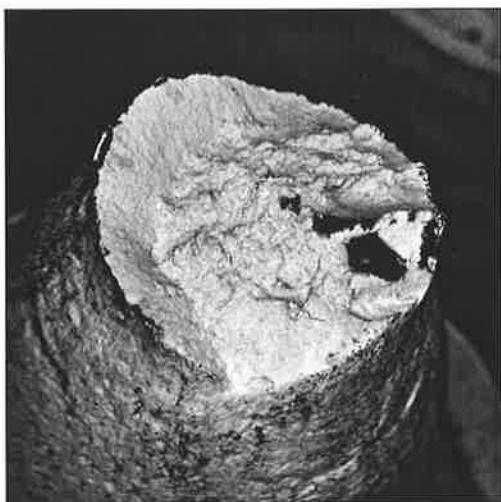


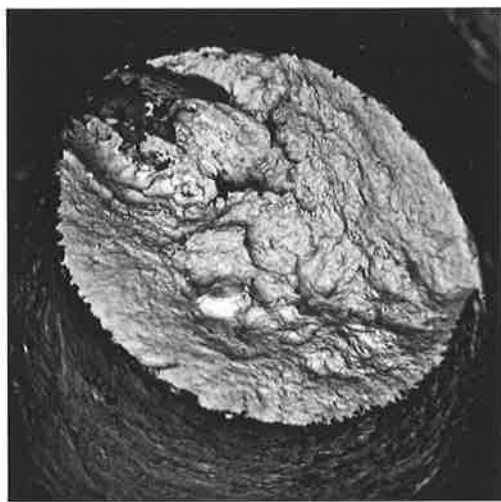
Figure 7-49: The effect of manganese content on weld tensile properties.



(a) Fracture surface from sample weld 1-1b.



(b) Fracture surface from sample weld 1-3b.



(c) Fracture surface from sample weld 1-4a.



(d) Fracture surface from sample weld 2-1a.



(e) Fracture surface from sample weld 2-3a.



(f) Fracture surface from sample weld 2-4b.

Figure 7-50: SEM backscattered images of selected tensile fracture surfaces.

7.13 STRUCTURE-PROPERTY RELATIONSHIPS

7.13.1 The Influence of Silicon on Structure-Property Relationships

The addition of silicon in the range 0.08 to 0.47% did not have any major effects on the microstructure or mechanical properties of the weld deposits examined. The proportion of acicular ferrite remained uniform while hardness and tensile strength only increased slightly. Evans [145] found that in weld deposits with high manganese levels (1.4 to 1.8%) silicon had little influence on microstructure or mechanical properties. The manganese level of the series 1 weld deposits ranged between 1.38 and 1.61%. Therefore the results of this experimental programme agree with those of Evans.

The addition of silicon did however influence the weld macrostructure in that it decreased the penetration of the weld bead.

7.13.2 The Influence of Manganese on Structure-Property Relationships

A dramatic refinement of the microstructure was observed as the manganese content increased from 0.53% to 1.71%. This was due to manganese lowering the $\gamma \rightarrow \alpha$ transformation temperature which resulted not only in restricting the growth of allotriomorphic ferrite but also allowed the intragranular nucleation of acicular ferrite.

In the SEM the coarseness of the low manganese welds was clearly evident as well as the large amount of pearlite between the laths. As the manganese content increased the lath size decreased and the amount of retained austenite increased.

Manganese also increased the weld hardness and tensile strength. These increases could be attributed directly to grain refinement and the increased presence of acicular ferrite.

However, it was clearly evident that the increased presence of acicular ferrite did not translate into improved Charpy impact toughness. A similar phenomenon was observed by Svensson and Gretoft [79]. They attributed the drop-off in toughness to the increased presence of microphases. They state that since the retained austenite islands are rich in carbon they are unstable and are likely to transform to brittle martensite during impact testing.

In the current experimental programme, TEM and SEM results confirmed the presence of large islands of retained austenite in the high manganese welds. It was also shown that the bulk of retained austenite islands were aligned on manganese rich cell boundaries. It is therefore probable that under impact testing these islands of retained austenite transformed to martensite and acted as crack initiation sites.

In light of this information it appears that the manganese level has to be carefully balanced. Sufficient manganese has to be added to achieve enough acicular ferrite (Svensson and Gretoft [79] recommend about 50% acicular ferrite) but not too much to cause excessive levels of microphases.

8. CONCLUSIONS

This study investigated the influence of silicon and manganese on the microstructure and mechanical properties of HC-GTA weld deposits in grade 250 steel. From the study several correlations were observed.

- Increasing the silicon content decreases the depth of penetration. Manganese additions on the other hand appear to increase depth of penetration.
- Increasing the silicon content of the weld deposit from 0.08% to 0.47% does not vary the as deposited microstructure greatly. Increasing the manganese content of the weld deposit refines the microstructure and promotes the nucleation of acicular ferrite.
- In weld deposits containing a high aluminium content γ alumina inclusions will form regardless of the silicon or manganese contents.
- Even at relatively low concentrations, manganese will segregate during solidification. The presence of manganese lowers the equilibrium partition coefficient of the weld metal allowing silicon to also segregate to cell boundaries.
- Varying the silicon content has little effect on the microphases formed between ferrite laths. At low manganese contents the microphase found between ferrite laths is pearlite. As the manganese content increases the amount of retained austenite also increases.
- Retained austenite is generally found within the manganese rich solidification cell boundaries.

- Increasing the silicon content slightly increases the hardness of the weld deposit.
Increasing the manganese content dramatically increases the hardness of the weld deposit.
- Increasing the silicon content slightly increased the tensile strength of the weld deposit.
Increasing the manganese content dramatically increases the tensile strength at the expense of elongation.
- A high proportion of acicular ferrite does not necessarily translate to high Charpy impact toughness if there is also a high proportion of microphases such as retained austenite present in the weld.

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APPENDIX A

Metal core wire mass balance.

A1.1 Introduction

Experiments at CSIRO found a metal core wire which showed promising results. The ingredients of the metal core are listed in Table A1-1.

	C wt%	Mn wt%	Si wt%	Al wt%	Fe	Metal core ratio	Wire ratio
Iron powder (Fe)	0.01	0.02	0.07	-	rem	69%	20%
Manganese powder (Mn)	-	99.5	-	-	-	25%	
Ferrosilicon powder (Fe/Si)	0.02	-	46.0	1.0	rem	6%	
Steel casing	0.05	0.22	0.01	0.015	rem		80%

Table A1-1: Experimental wire ingredients and their chemical composition.

A1.2 Mass balance of silicon in the metal core wire:

% Si in iron powder	$69\% \times 0.07\% \times 20\%$	0.01%
+ % Si in manganese powder		0.00%
+ % Si in ferrosilicon powder	$6\% \times 46\% \times 20\%$	0.55%
+ % Si in steel casing	$0.01\% \times 80\%$	0.01%
= % Si in the filler wire		0.57%

Actual wire Si content = 0.43%

$$\therefore \text{Wire correction factor} = \frac{\text{Predicted Si content}}{\text{Actual Si content}} = \frac{0.57\%}{0.43\%} = 1.33$$

Weld Si content: 0.24%

Weld area: 103 mm^2

$$\therefore \text{Mass of weld per linear mm} = 103 \text{ mm}^2 \times 7.96 \times 10^{-3} \text{ g mm}^{-3} = 0.8217 \text{ g per mm}$$

Dilution: 76%

$$\text{Mass of base material melted (per linear mm)} = \text{Weld Mass} \times \text{Dilution}$$

$$0.8217 \times 76\% = 0.6227 \text{ g/mm}$$

Mass of filler wire melted (per linear mm)

$$\text{Wire weight: } 5.81 \text{ g/m} \Rightarrow 0.00581 \text{ g/mm}$$

$$\text{Travel speed: } 150 \text{ mm/min} \Rightarrow 2.5 \text{ mm/s} \therefore \text{Time to travel 1mm} = 0.4\text{s}$$

Wire feed rates:

	Wire feed speed (mm/s)	Wire used per linear mm (speed x 0.4)
Side 1	50	20
Side 2	13	5
Total		25

$$\therefore \text{Mass of wire melted per linear mm} = 25 \text{ mm} \times 0.00581 \text{ g/mm} = 0.1453 \text{ g}$$

A1.2.1 Silicon weld mass balance:

Si from base material	$0.6227 \text{ g} \times 0.10\% =$	$6.23 \times 10^{-4} \text{ g}$
+ Si from filler wire	$0.1453 \text{ g} \times 0.43\% =$	$6.25 \times 10^{-4} \text{ g}$
= Mass of Si in the weld		$1.25 \times 10^{-3} \text{ g}$

$$\text{Predicted Weld Si \%} = \frac{\text{Mass of Si in the weld}}{\text{Mass of Weld metal}} = \frac{1.25 \times 10^{-3} \text{ g}}{0.8217 \text{ g}} = 0.0015 \Rightarrow 0.15 \%$$

$$\text{Weld Correction factor (Weld C.F.)} = \frac{\text{Predicted Si Content}}{\text{Actual Si Content}} = \frac{0.15\%}{0.24\%} = 0.63$$

$$\begin{aligned}
\text{Si content of filler wire} &= \frac{\text{Mass of Si in the weld} - \text{Mass of Si from melted Base Material}}{\text{Mass of wire used}} \\
&= \frac{\text{Weld Si\%} \times \text{Weld C.F.} \times 0.8219\text{g} - 0.6227\text{g} \times 0.10\%}{0.1453\text{g}} \\
&= 3.5788 \times \text{Weld Si\%} - 0.0043
\end{aligned}$$

To examine the effect of Silicon on the weld microstructure and mechanical properties the Manganese content of the filler wires will remain constant.

A1.2.2 Determination of the metal powder mixture

Assuming the sum of the fractions of metal powder equal 1 and Manganese powder fraction equals 0.25

$$\therefore \text{Iron powder fraction} = 0.75 - \text{ferrosilicon powder fraction}$$

$$\text{Wire Si content} = \frac{\text{Si content of metal powder} \times 0.2 + \text{Si content of steel casing} \times 0.8}{\text{Wire correction factor}}$$

$$\text{Wire correction factor} = 1.33$$

$$\text{Si content of steel casing} = 0.01\%$$

$$\text{Si content of metal powder} = (0.75 - \chi) \times 0.07\% + \chi \times 46\%$$

where χ = ferrosilicon powder fraction

Entering these values into the equation:

$$\text{Wire Si content} \times 1.33 = ((0.75 - \chi) \times 0.07\% + \chi \times 46\%) \times 0.2 + 0.01\% \times 0.8$$

Simplifying:

$$\begin{aligned} \text{Wire Si content} \times 1.33 &= 1.05 \times 10^{-4} - 1.4 \times 10^{-4}\chi + 0.092\chi + 8 \times 10^{-5} \\ &= 0.0919\chi + 1.85 \times 10^{-4} \end{aligned}$$

$$\therefore \chi = 14.4786 \times \text{Wire Si Content} - 1.391 \times 10^{-4}$$

Incorporating wire Si content equation

$$\chi = 51.8173 \times \text{Weld Si\%} - 0.0622$$

From this formula it was possible to calculate the metal powder ratios to give the required weld Si contents as shown in Table A1-2.

Wire Number	Predicted Weld Silicon Content (%)	Predicted Wire Silicon content (%)	Metal powder ratios		
			Fe/Si (χ)	Fe ($0.75 - \chi$)	Mn
1-1	Minimum	0.0	0.00	0.75	0.25
1-2	.25	0.89	0.07	0.68	0.25
1-3	.30	1.07	0.10	0.65	0.25
1-4	.40	1.42	0.15	0.60	0.25

Table A1-2: Metal powder ratios for the silicon series wires.

A1.3 Manganese mass balance for metal core wire

% Mn in iron powder	$69\% \times 0.02\% \times 20\%$	0.00%
+ % Mn in manganese powder	$25\% \times 99.5\% \times 20\%$	4.98%
+ % Mn in ferrosilicon powder		0.00%
+ % Mn in steel casing	$0.22\% \times 80\%$	0.18%
= % Mn in the filler wire		5.16%

Actual wire Mn content = 4.95%

$$\therefore \text{Wire correction factor} = \frac{\text{Predicted Mn content}}{\text{Actual Mn content}} = \frac{5.16\%}{4.95\%} = 1.04$$

A1.3.1 Manganese weld mass balance:

Mn from base material	$0.6227 \text{ g} \times 0.68\% =$	$4.23 \times 10^{-3} \text{ g}$
+ Mn from filler wire	$0.1453 \text{ g} \times 4.95\% =$	$7.19 \times 10^{-3} \text{ g}$
= Mass of Mn in the weld		$1.14 \times 10^{-2} \text{ g}$

$$\text{Predicted Weld Mn \%} = \frac{\text{Mass of Mn in the weld}}{\text{Mass of Weld metal}} = \frac{1.14 \times 10^{-2} \text{ g}}{0.8217 \text{ g}} = 0.0139 \Rightarrow 1.39 \%$$

$$\text{Weld Correction factor (Weld C.F.)} = \frac{\text{Predicted Mn Content}}{\text{Actual Mn Content}} = \frac{1.39\%}{1.32\%} = 1.05$$

$$\text{Filler wire Mn content} = \frac{\text{Mass of Mn in the weld} - \text{Mass of Mn from melted Base Material}}{\text{Mass of wire used}}$$

$$= \frac{\text{Weld Mn\%} \times \text{Weld C.F.} \times 0.8219\text{g} - 0.6227\text{g} \times 0.68\%}{0.1453\text{g}}$$

$$= 5.9394 \times \text{Weld Mn\%} - 0.0291$$

To examine the effect of manganese on the weld microstructure and mechanical properties the silicon content of the filler wires will remain constant.

A1.3.2 Determination of the metal powder mixture

Assuming the sum of the fractions of metal powder equal 1 and Ferrosilicon powder fraction equals 0.07.

$$\therefore \text{Iron powder fraction} = 0.93 - \text{manganese fraction}$$

$$\text{Wire Mn content} = \frac{\text{Mn content of metal powder} \times 0.2 + \text{Mn content of steel casing} \times 0.8}{\text{Wire correction factor}}$$

$$\text{Wire correction factor} = 1.04$$

$$\text{Mn content of steel casing} = 0.22\%$$

$$\text{Mn content of metal powder} = (0.93 - \chi) \times 0.02\% + \chi \times 99.5\%$$

where χ = manganese powder fraction

Entering these values into the equation:

$$\text{Wire Mn content} \times 1.04 = ((0.93 - \chi) \times 0.02\% + \chi \times 99.5\%) \times 0.2 + 0.22\% \times 0.8$$

Simplifying:

$$\text{Wire Mn content} \times 1.04 = 3.72 \times 10^{-5} - 4 \times 10^{-5}\chi + 0.199\chi + 0.0018$$

$$= 0.199\chi + 0.0018$$

$$\therefore \chi = 5.2356 \times \text{Wire Mn Content} - 0.009$$

Incorporating wire Mn content equation

$$\chi = 31.0963 \times \text{Weld Mn\%} - 0.1614$$

Table A1-3 lists the metal powder ratios calculated to give the desired weld Manganese content. Note that wire No. 1-3 is used in both the silicon and manganese series.

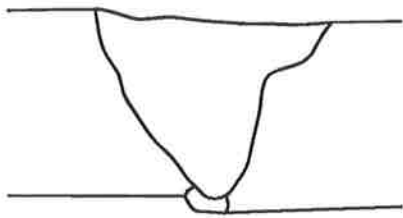
Wire Number	Predicted Weld Manganese Content (%)	Predicted Wire Manganese content (%)	Metal powder ratios		
			Fe/Si	Fe (0.93 - χ)	Mn (χ)
2-1	Minimum	0.17	0.07	0.93	0.000
2-2	0.90	2.44	0.07	0.81	0.12
2-3	1.10	3.62	0.07	0.75	0.18
1-2	1.30	4.80	0.07	0.68	0.25
2-4	1.50	6.00	0.07	0.62	0.31

Table A1-3: Metal powder ratios for the Manganese series wires.

APPENDIX B

Dilution Calculations.

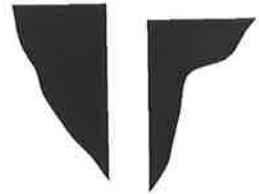
Weld 1-1



Weld Profile



Weld Area: 102.9 mm²

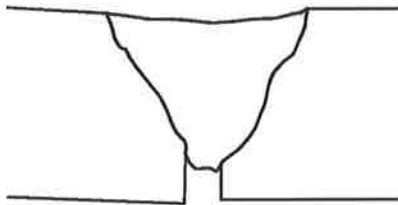


Base Area: 42.9
+ 31.7

74.6 mm²

$$\text{Dilution} = \frac{74.6 \text{ mm}^2}{102.9 \text{ mm}^2} \times 100\% = 72.5\%$$

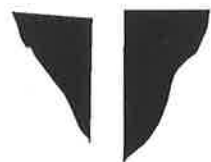
Weld 1-2



Weld Profile



Weld Area: 75.4 mm²

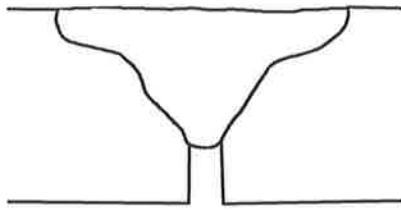


Base Area: 33.7
+ 22.6

56.3 mm²

$$\text{Dilution} = \frac{56.3 \text{ mm}^2}{75.4 \text{ mm}^2} \times 100\% = 74.4\%$$

Weld 1-3



Weld Profile



Weld Area: 88.4 mm²

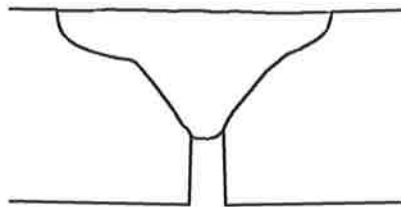


Base Area: 33.8
+ 33.7

67.5 mm²

$$\text{Dilution} = \frac{67.5 \text{ mm}^2}{88.4 \text{ mm}^2} \times 100\% = 76.4\%$$

Weld 1-4



Weld Profile



Weld Area: 78.3 mm²

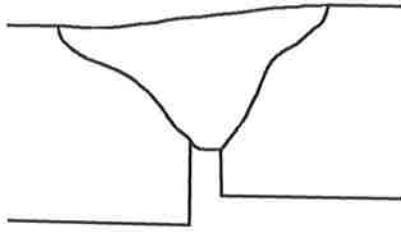


Base Area: 32.2
+ 28.3

60.5 mm²

$$\text{Dilution} = \frac{60.5 \text{ mm}^2}{78.3 \text{ mm}^2} \times 100\% = 72.5\%$$

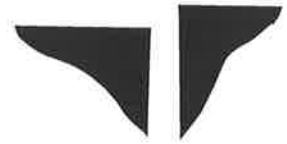
Weld 2-1



Weld Profile



Weld Area: 79.4 mm²

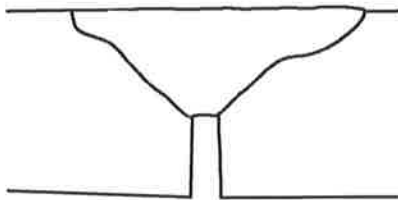


Base Area: 28.5
+ 30.2

58.7 mm²

$$\text{Dilution} = \frac{58.7 \text{ mm}^2}{79.4 \text{ mm}^2} \times 100\% = 73.9\%$$

Weld 2-2



Weld Profile



Weld Area: 77.2 mm²

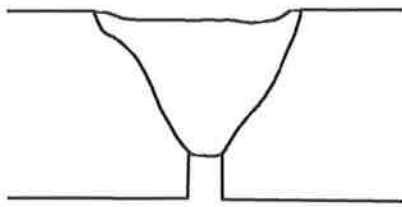


Base Area: 25.2
+ 29.2

54.4 mm²

$$\text{Dilution} = \frac{54.4 \text{ mm}^2}{77.2 \text{ mm}^2} \times 100\% = 70.5\%$$

Weld 2-3



Weld Profile



Weld Area: 69.3 mm²

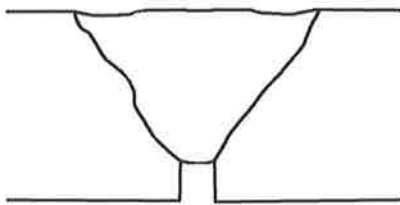


Base Area: 27.5
+ 25.0

52.5 mm²

$$\text{Dilution} = \frac{52.5 \text{ mm}^2}{69.3 \text{ mm}^2} \times 100\% = 75.8\%$$

Weld 2-4



Weld Profile



Weld Area: 92.1 mm²



Base Area: 33.8
+ 35.3

69.1 mm²

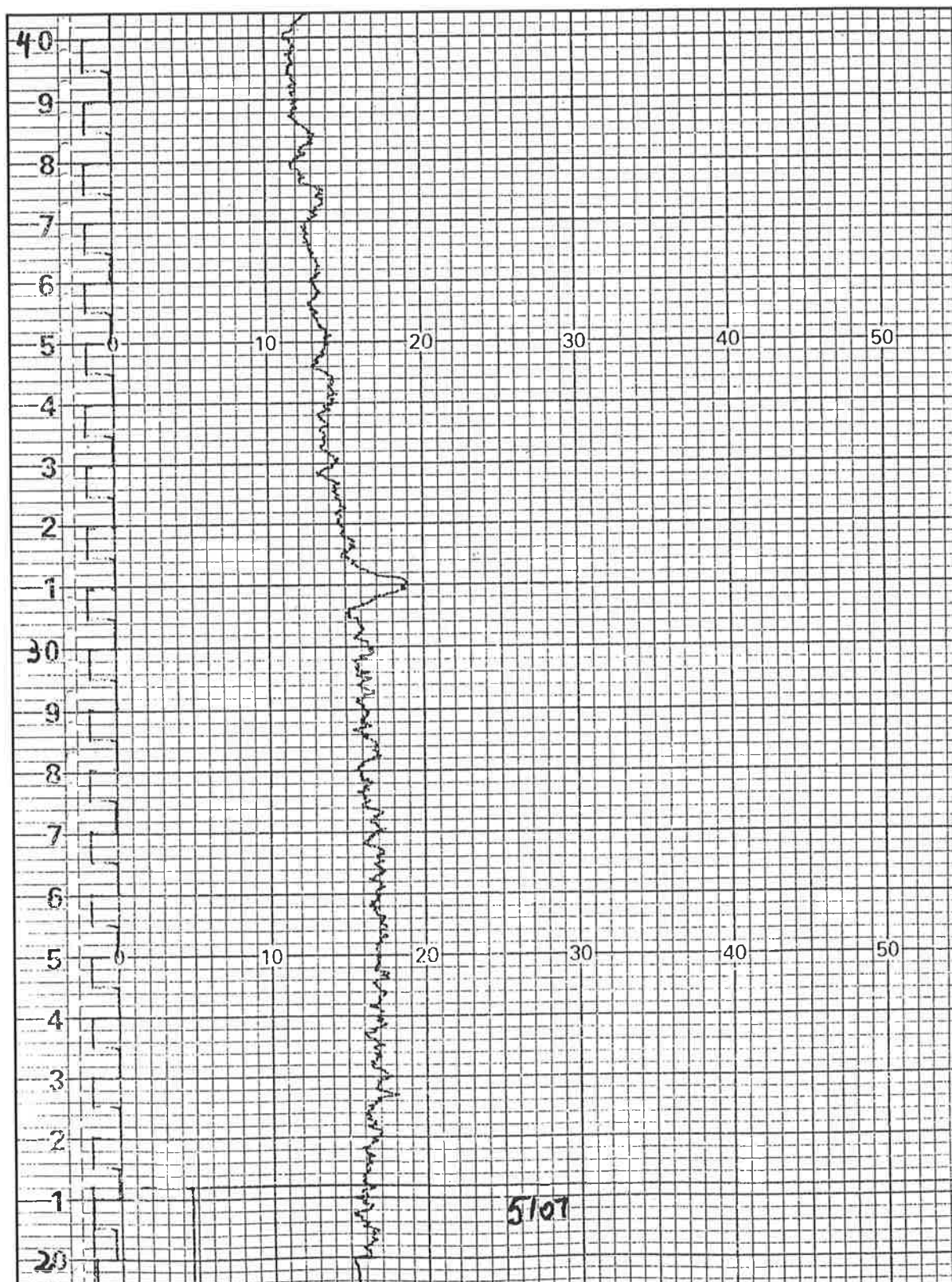
$$\text{Dilution} = \frac{69.1 \text{ mm}^2}{92.1 \text{ mm}^2} \times 100\% = 75.0\%$$

APPENDIX C

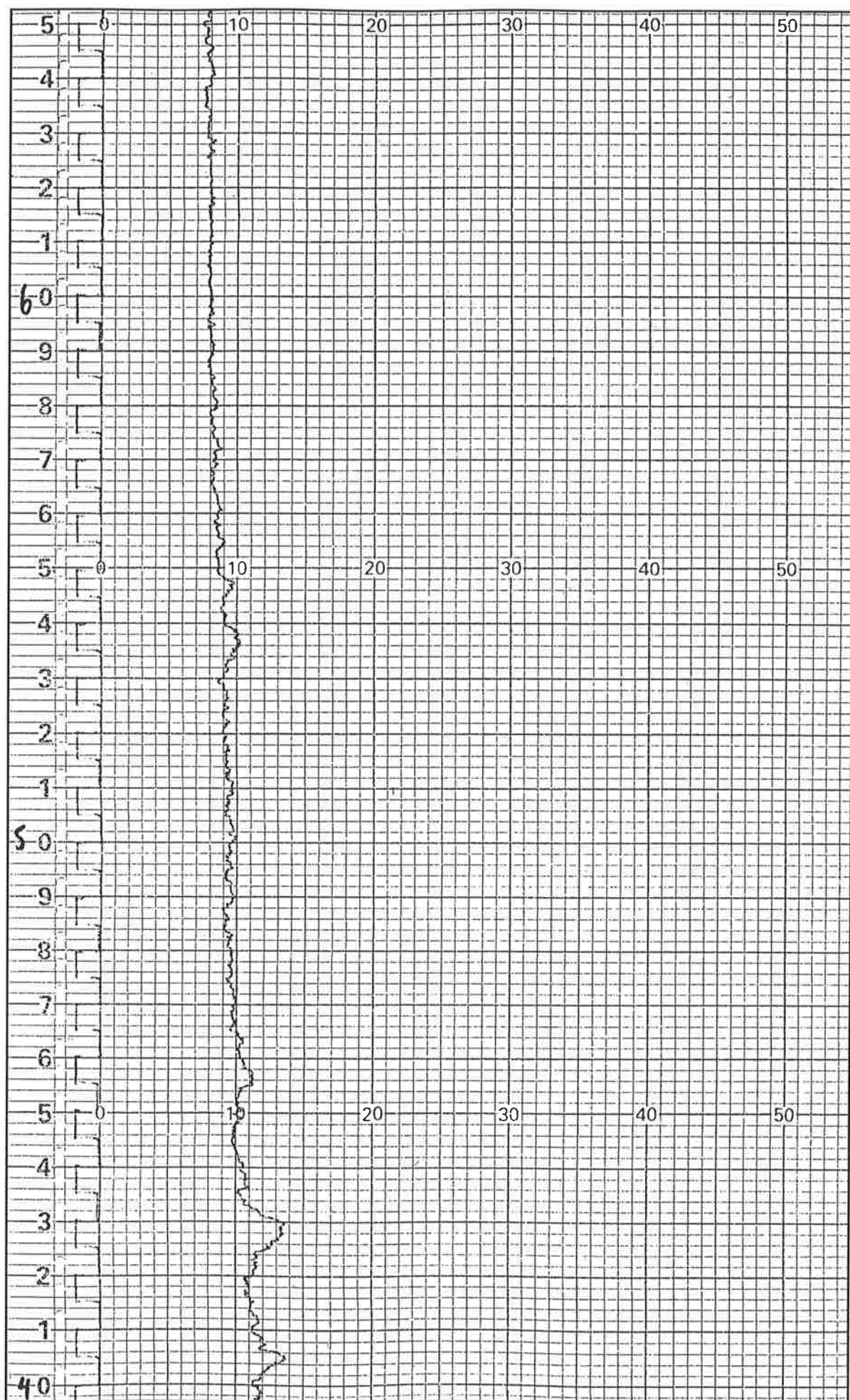
XRPD Chart Recorder Traces.

Weld 1-1

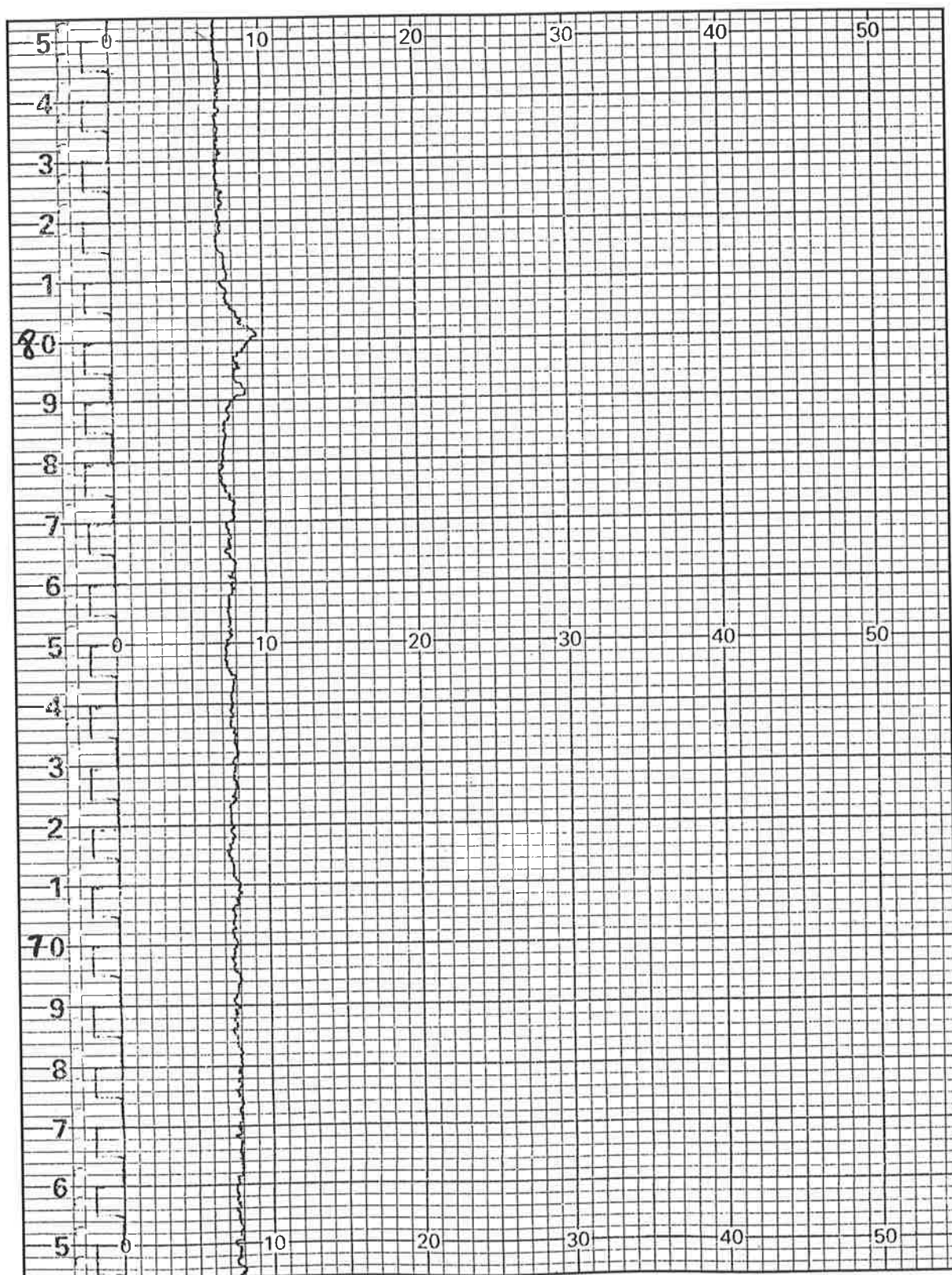
XRPD Trace: Weld 1-1, $2\theta = 20^\circ$ to 40°



XRPD Trace: Weld 1-1, $2\theta = 40^\circ$ to 65°

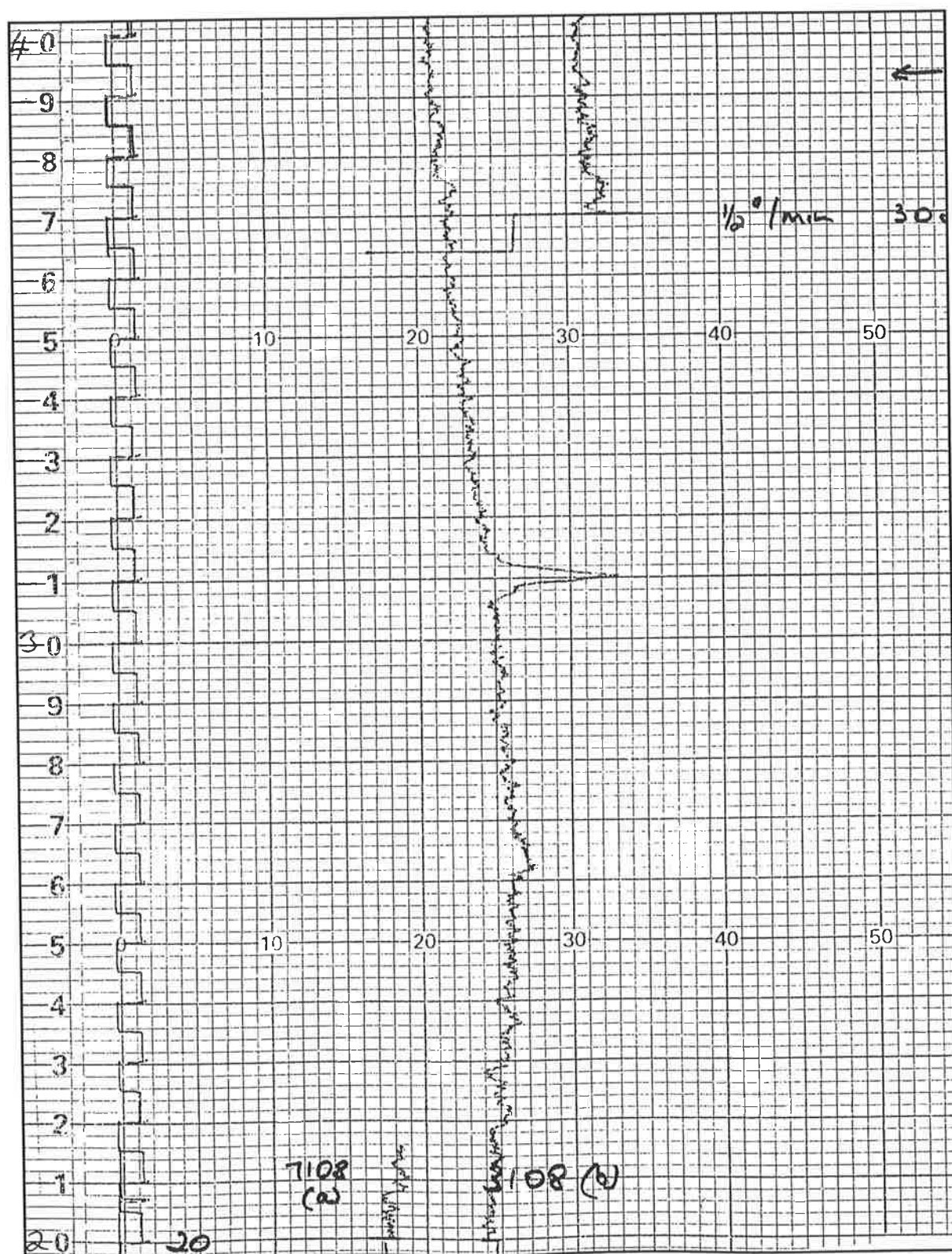


XRPD Trace: Weld 1-1, $2\theta = 65^\circ$ to 85°

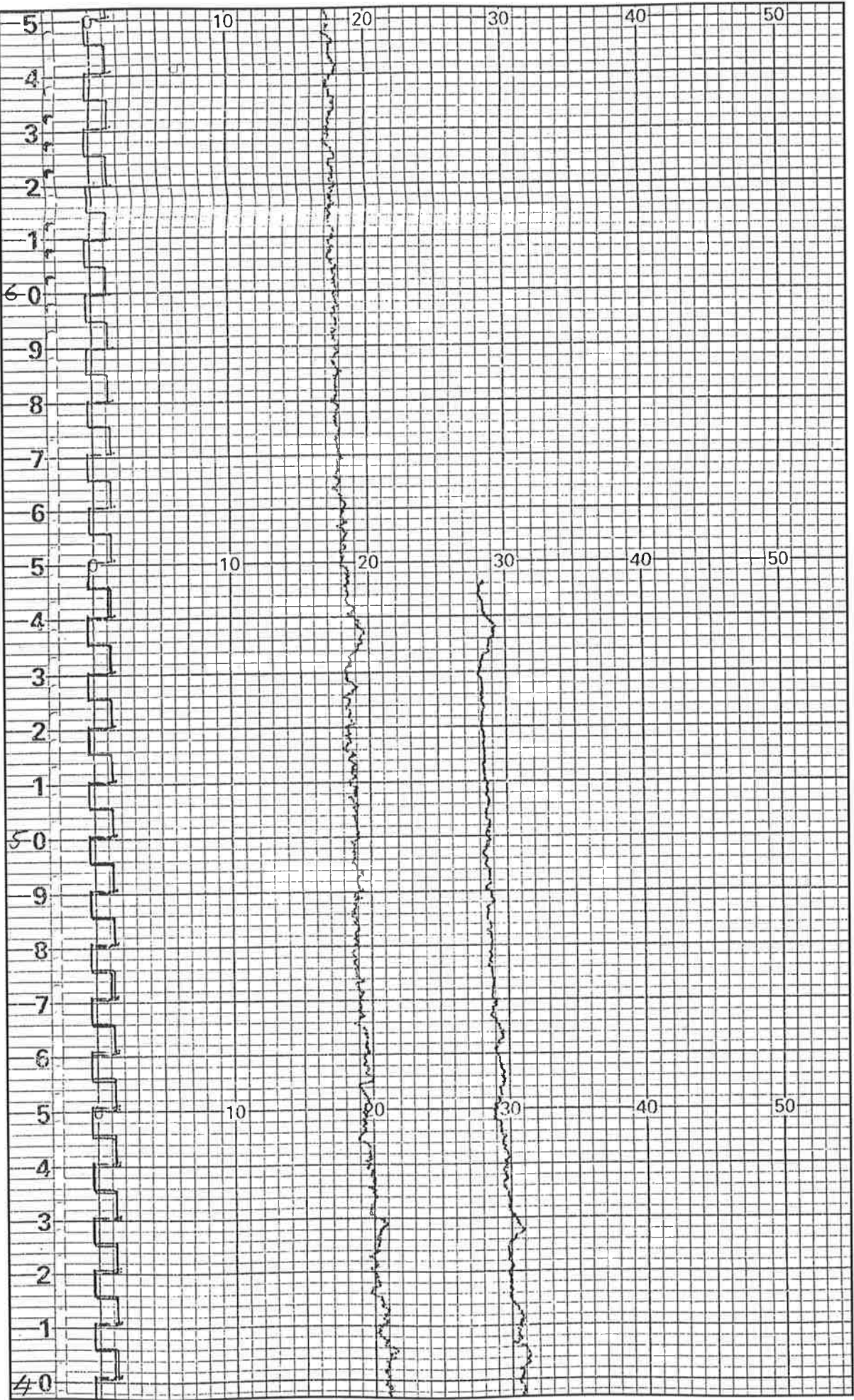


Weld 1-2

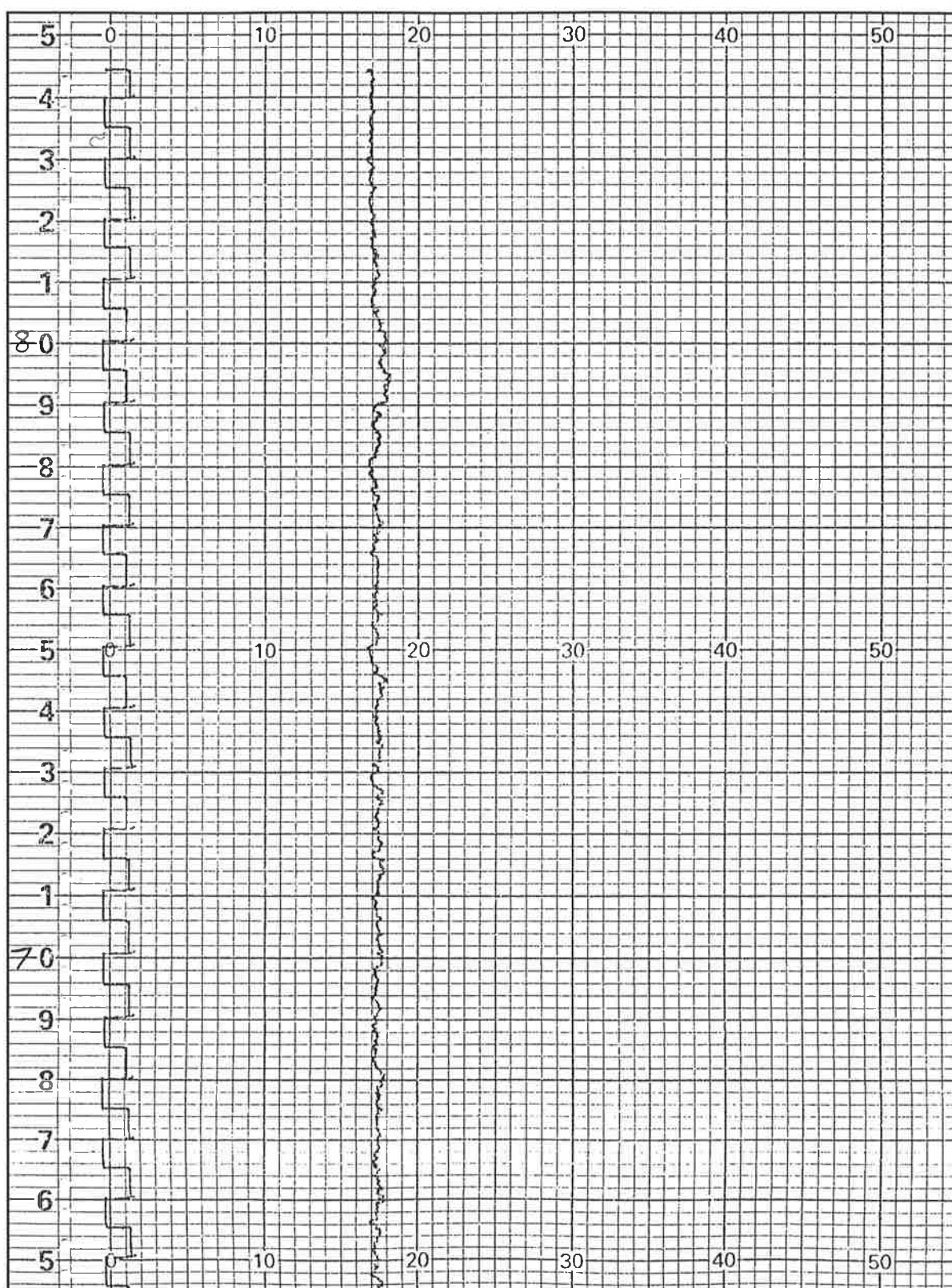
XRPD Trace: Weld 1-2, $2\theta = 20^\circ$ to 40°



XRPD Trace: Weld 1-2, $2\theta = 40^\circ$ to 65°

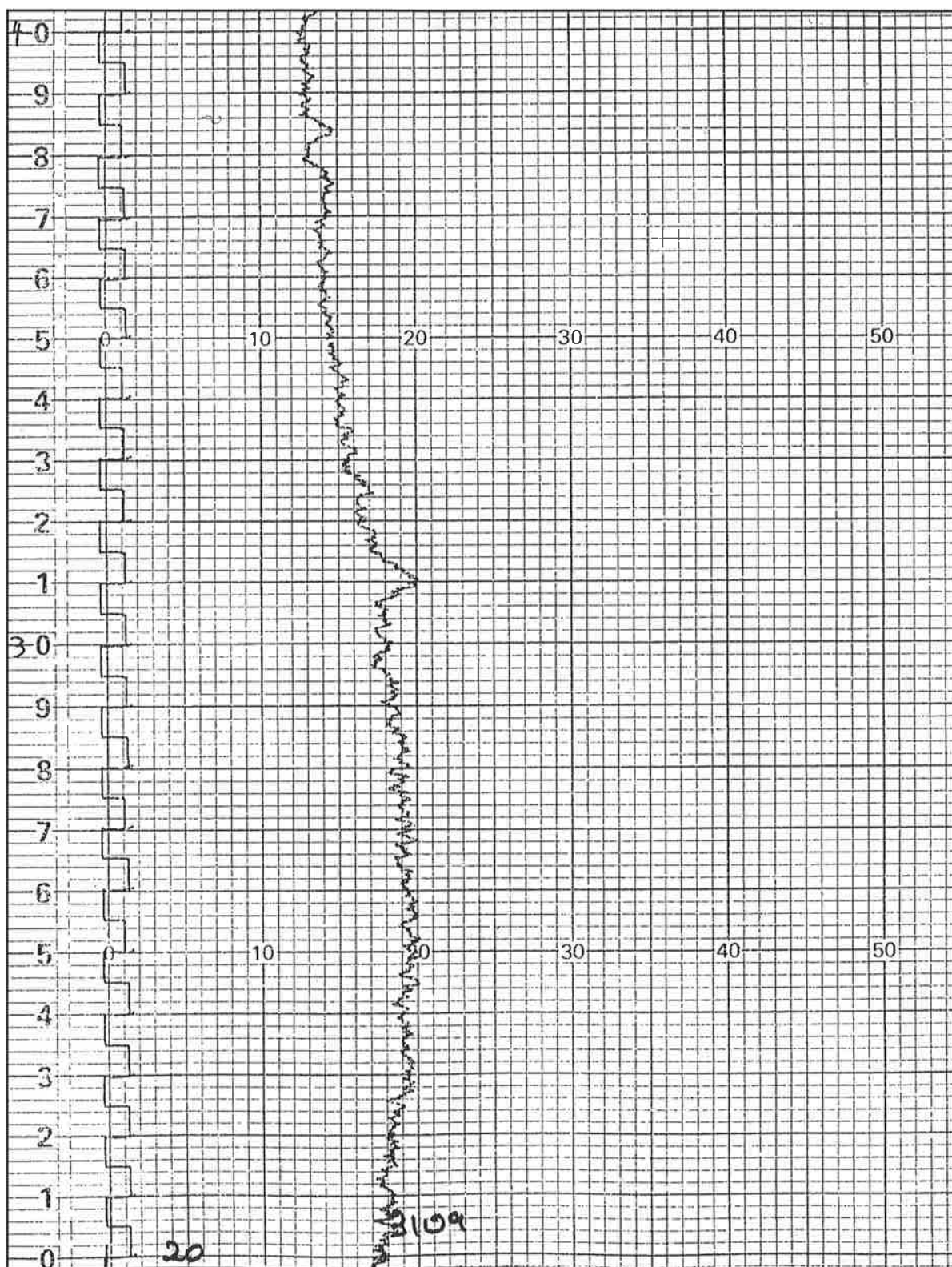


XRPD Trace: Weld 1-2, $2\theta = 65^\circ$ to 85°

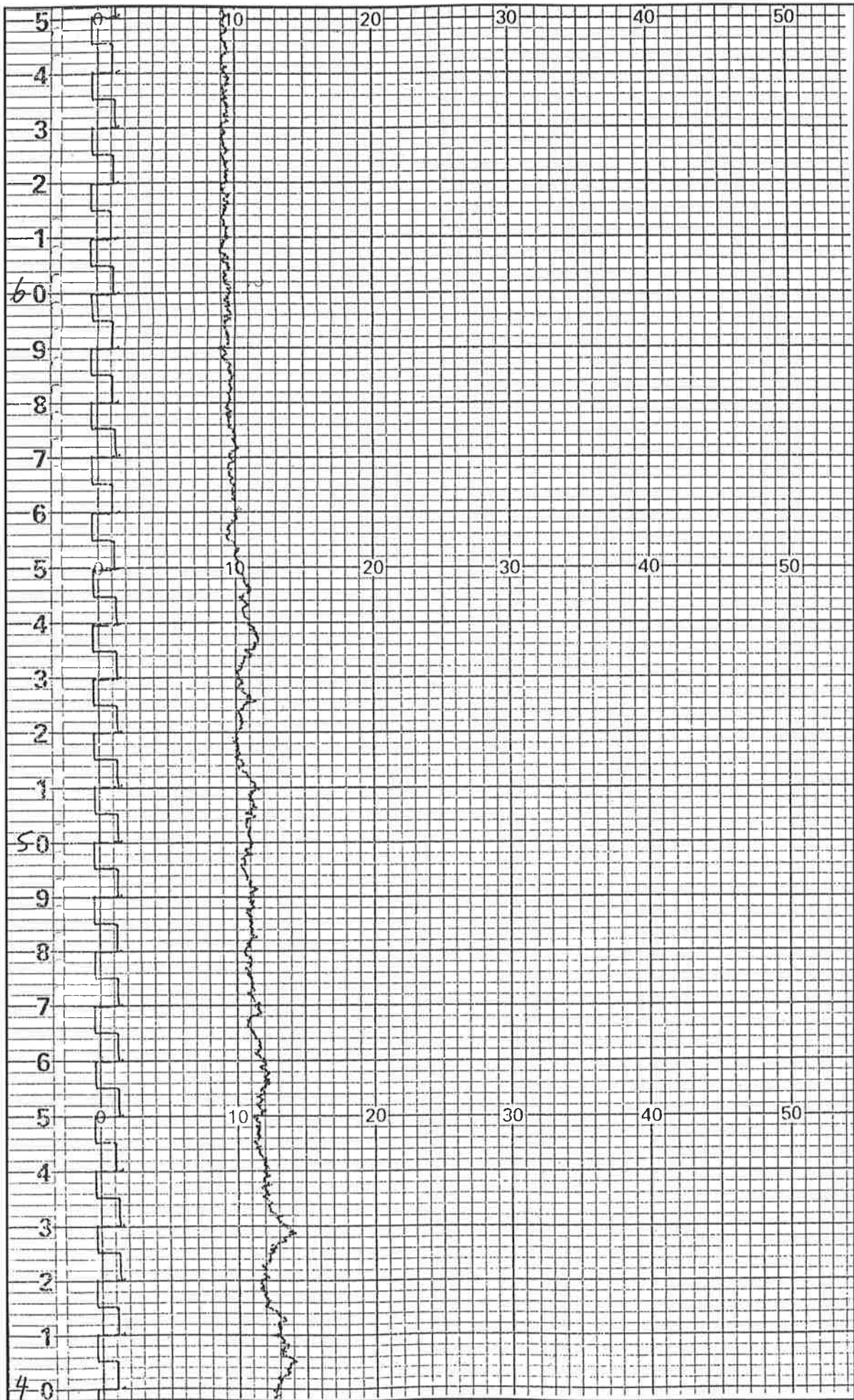


Weld 1-3

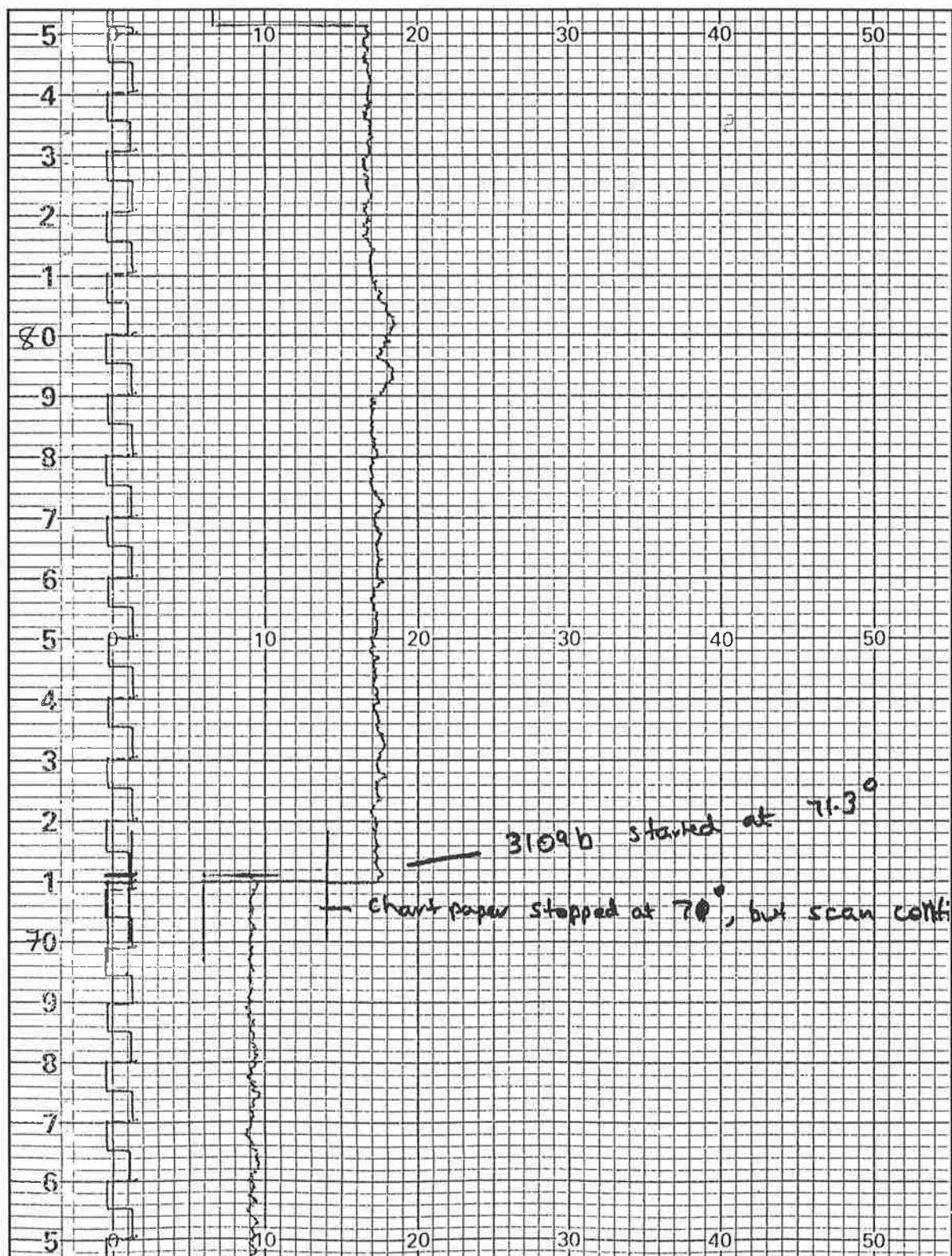
XRPD Trace: Weld 1-3, $2\theta = 20^\circ$ to 40°



XRPD Trace: Weld 1-3, $2\theta = 40^\circ$ to 65°

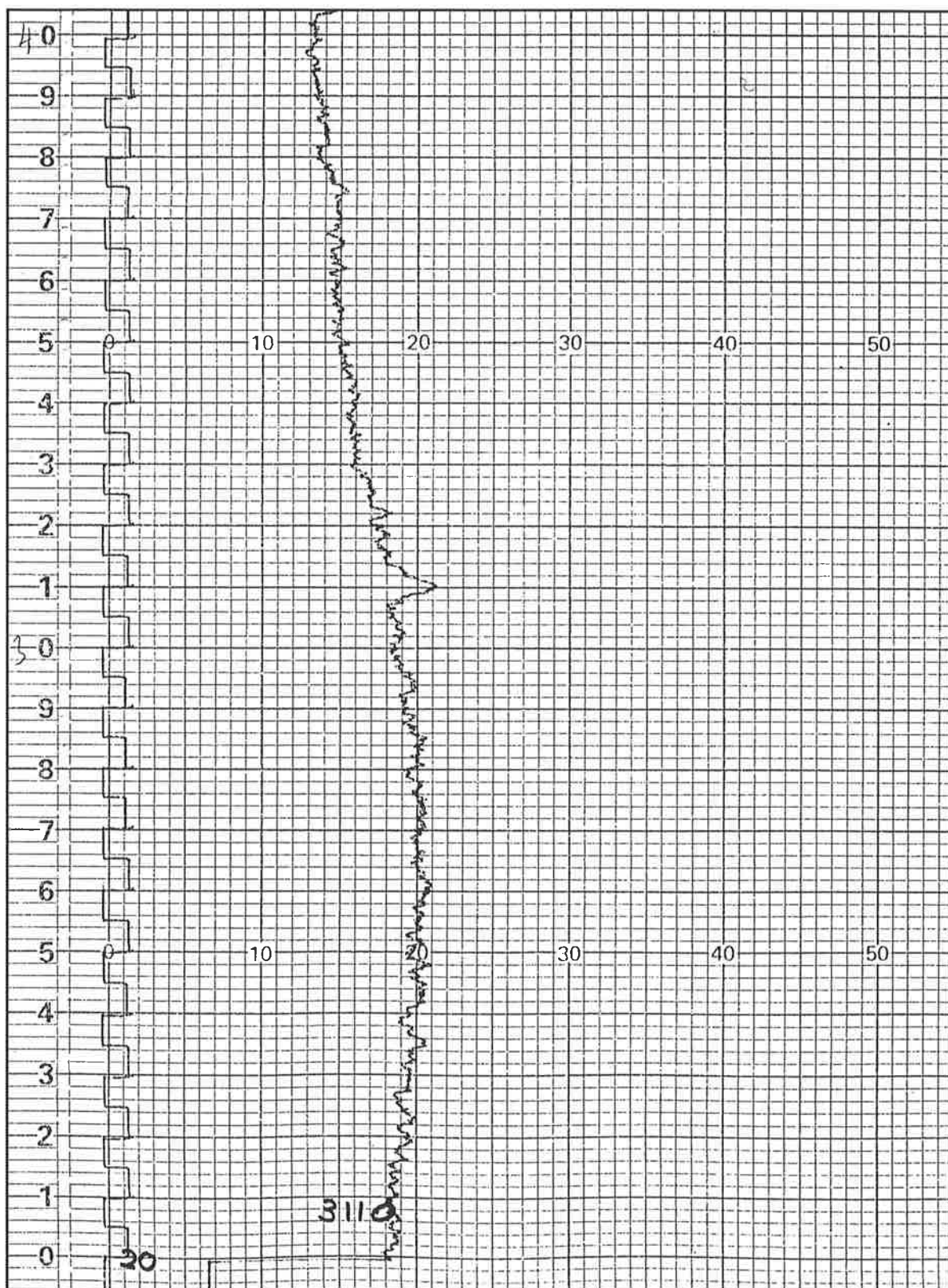


XRPD Trace: Weld 1-3, $2\theta = 65^\circ$ to 85°

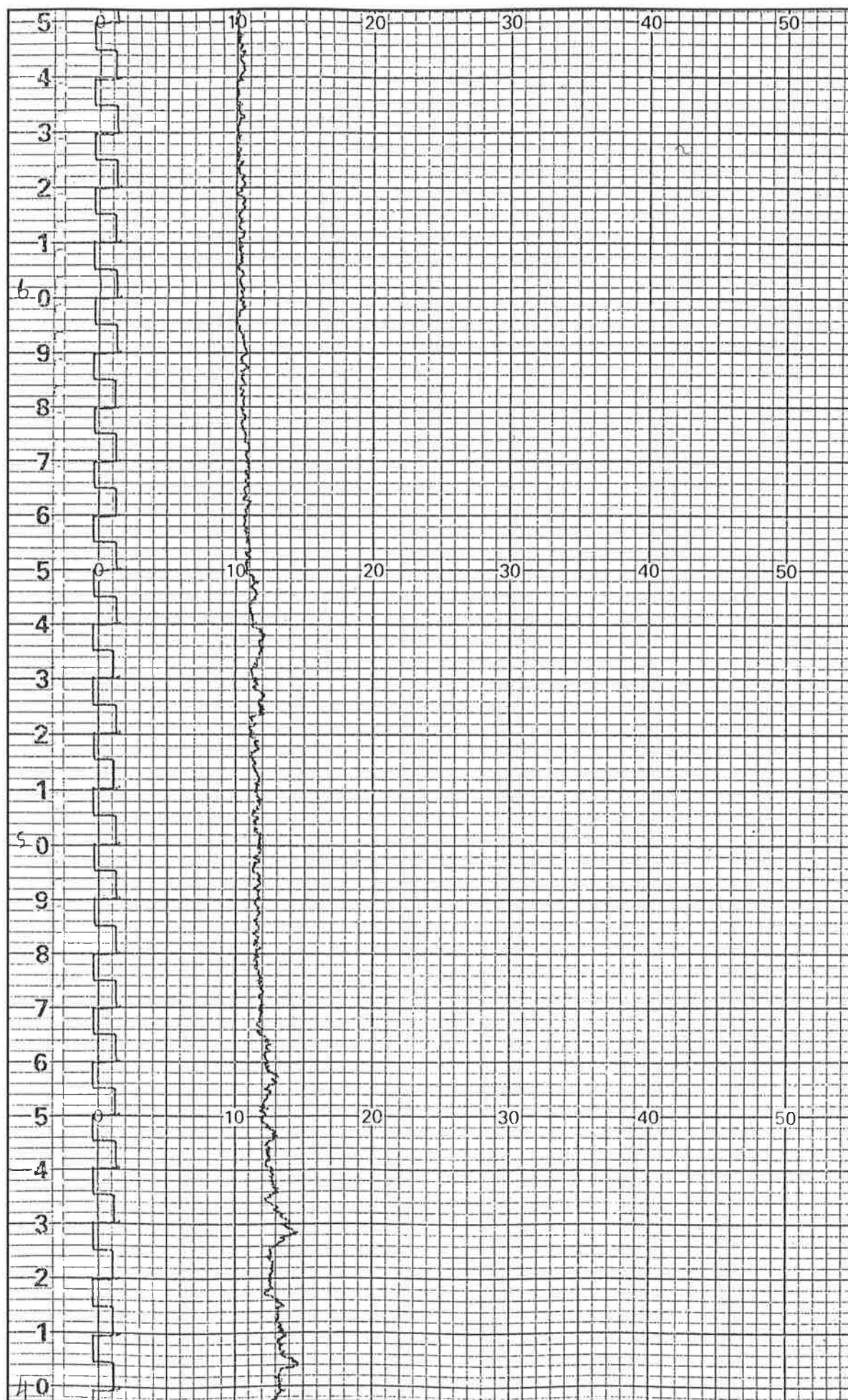


Weld 1-4

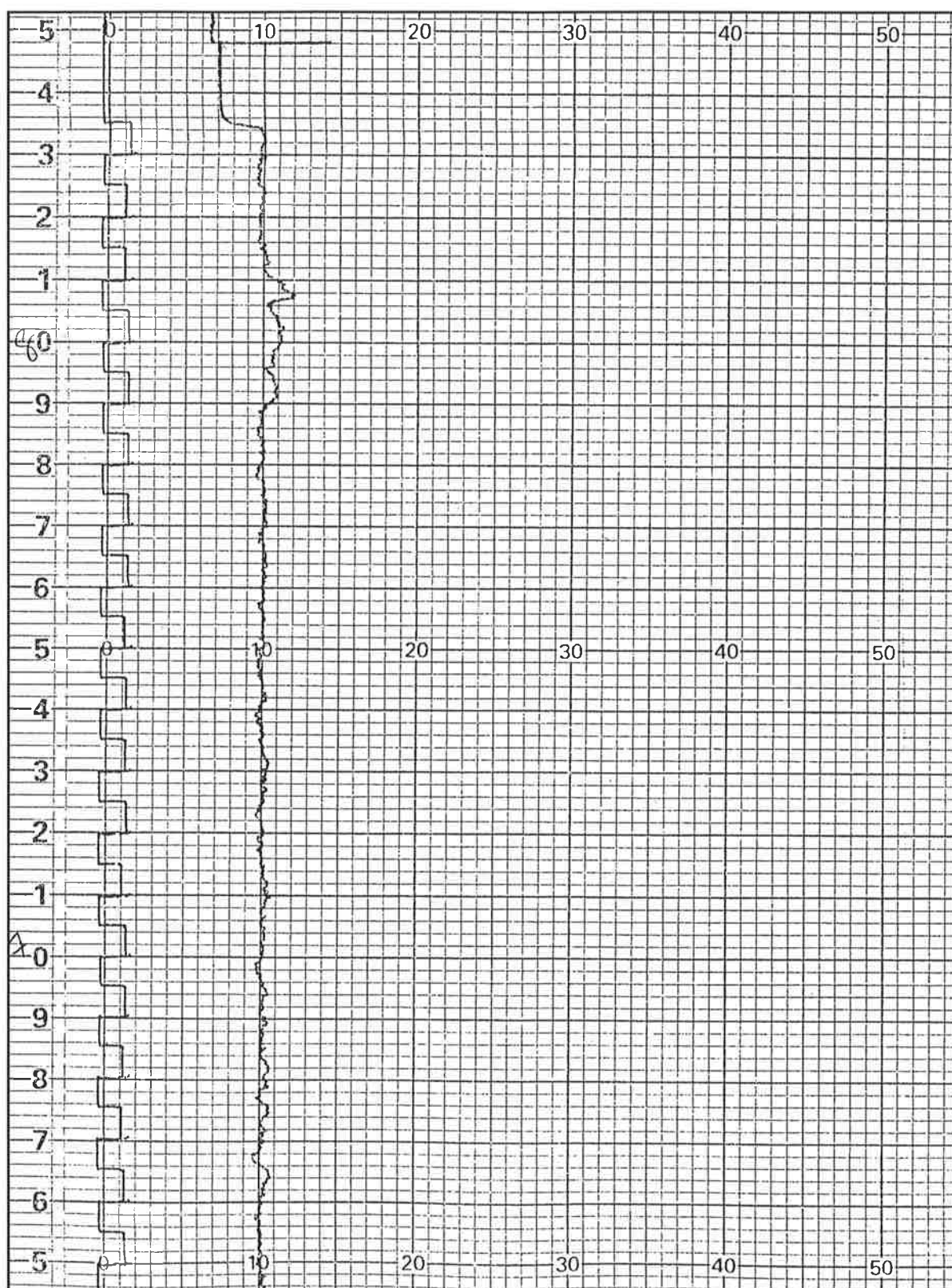
XRPD Trace: Weld 1-4, $2\theta = 20^\circ$ to 40°



XRPD Trace: Weld 1-4, $2\theta = 40^\circ$ to 65°

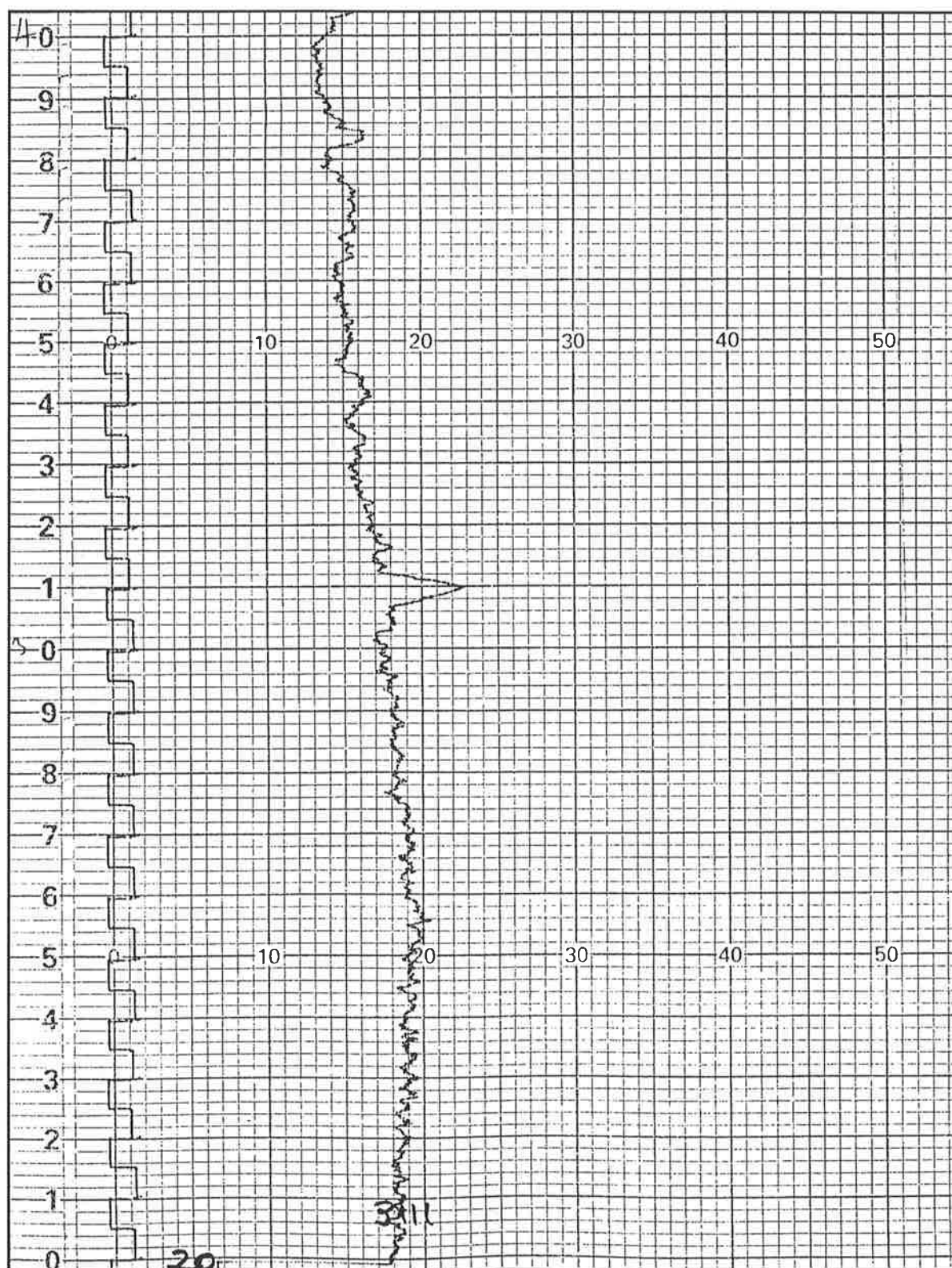


XRPD Trace: Weld 1-4, $2\theta = 65^\circ$ to 85°

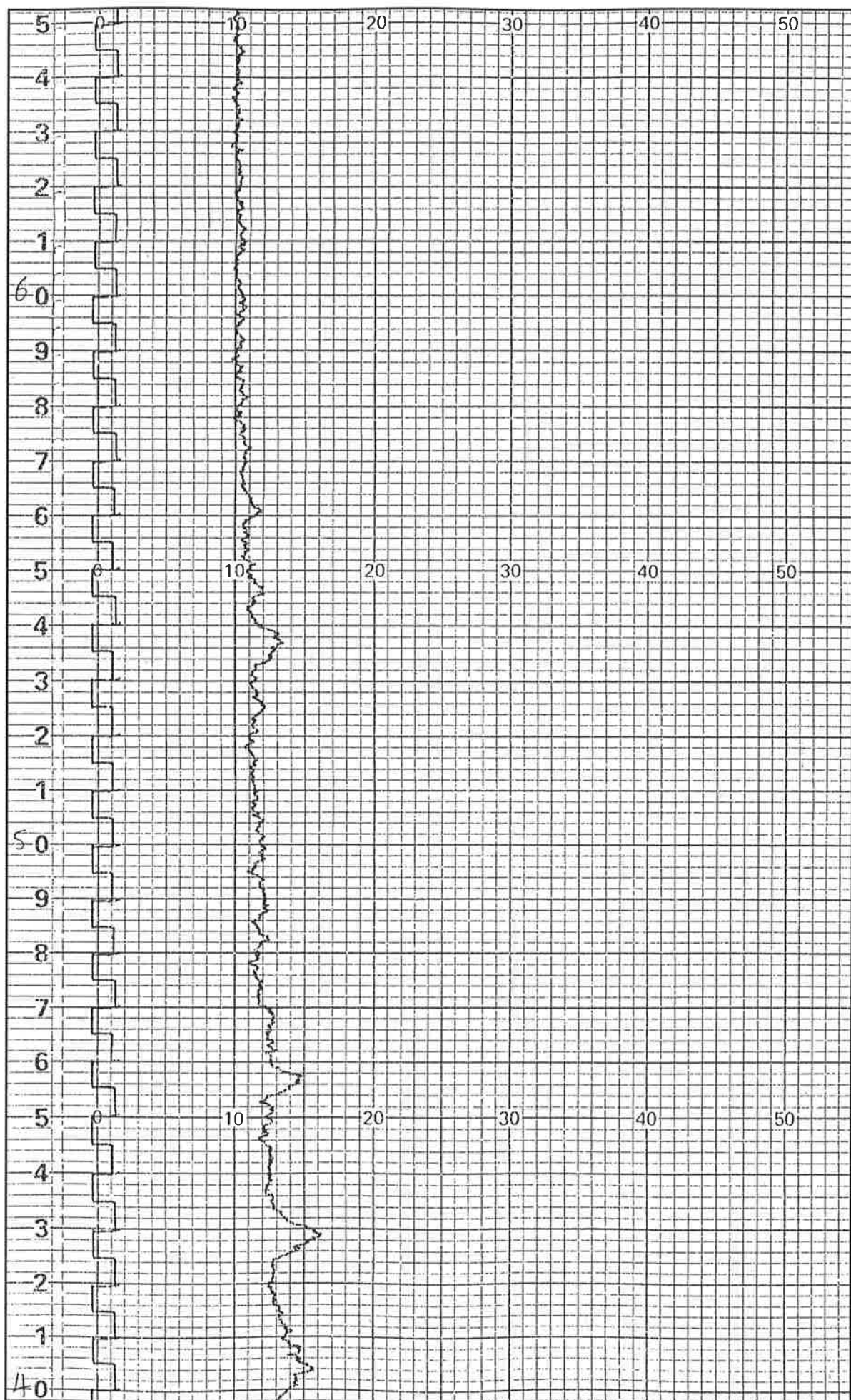


Weld 2-1

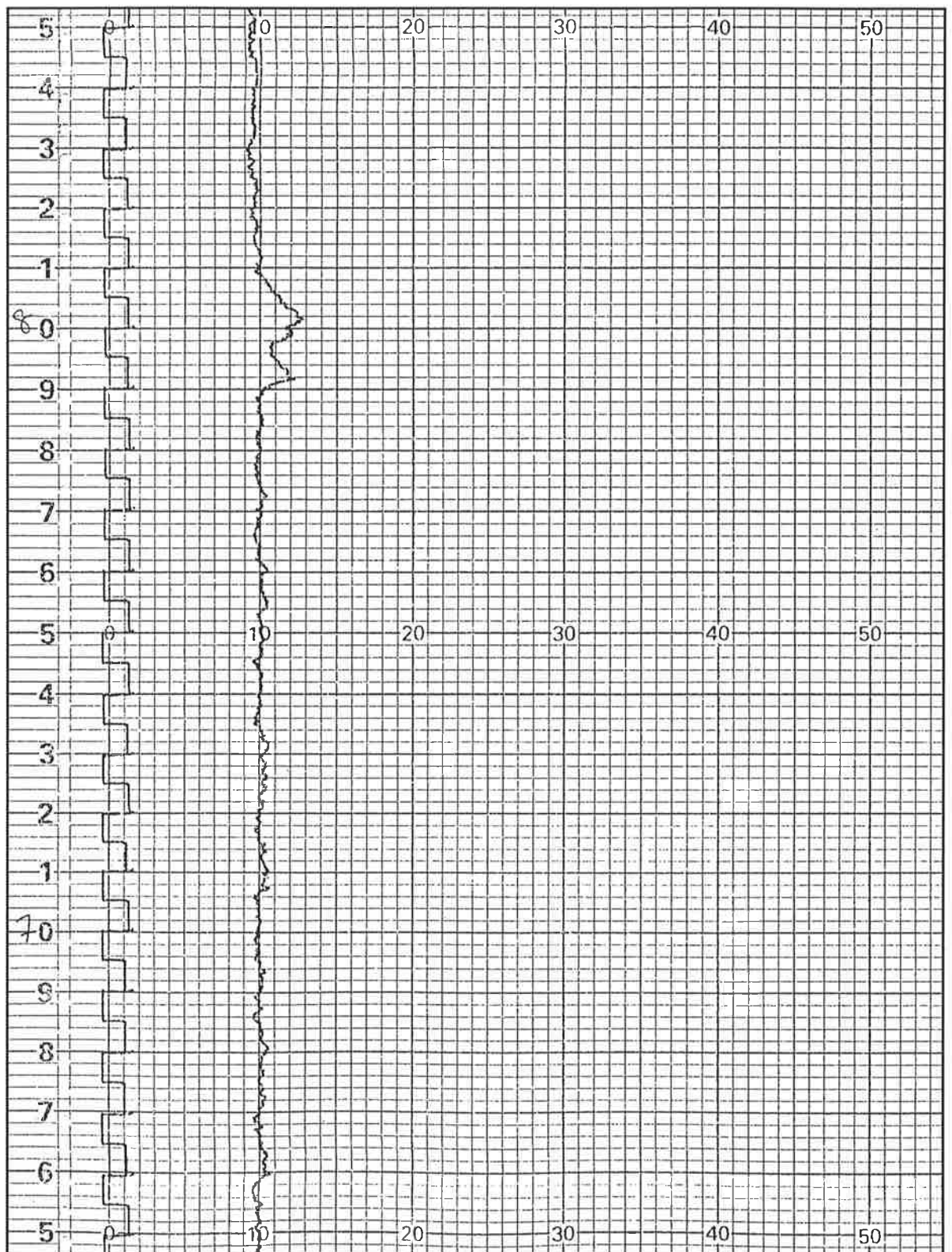
XRPD Trace: Weld 2-1, $2\theta = 20^\circ$ to 40°



XRPD Trace: Weld 2-1, $2\theta = 40^\circ$ to 65°

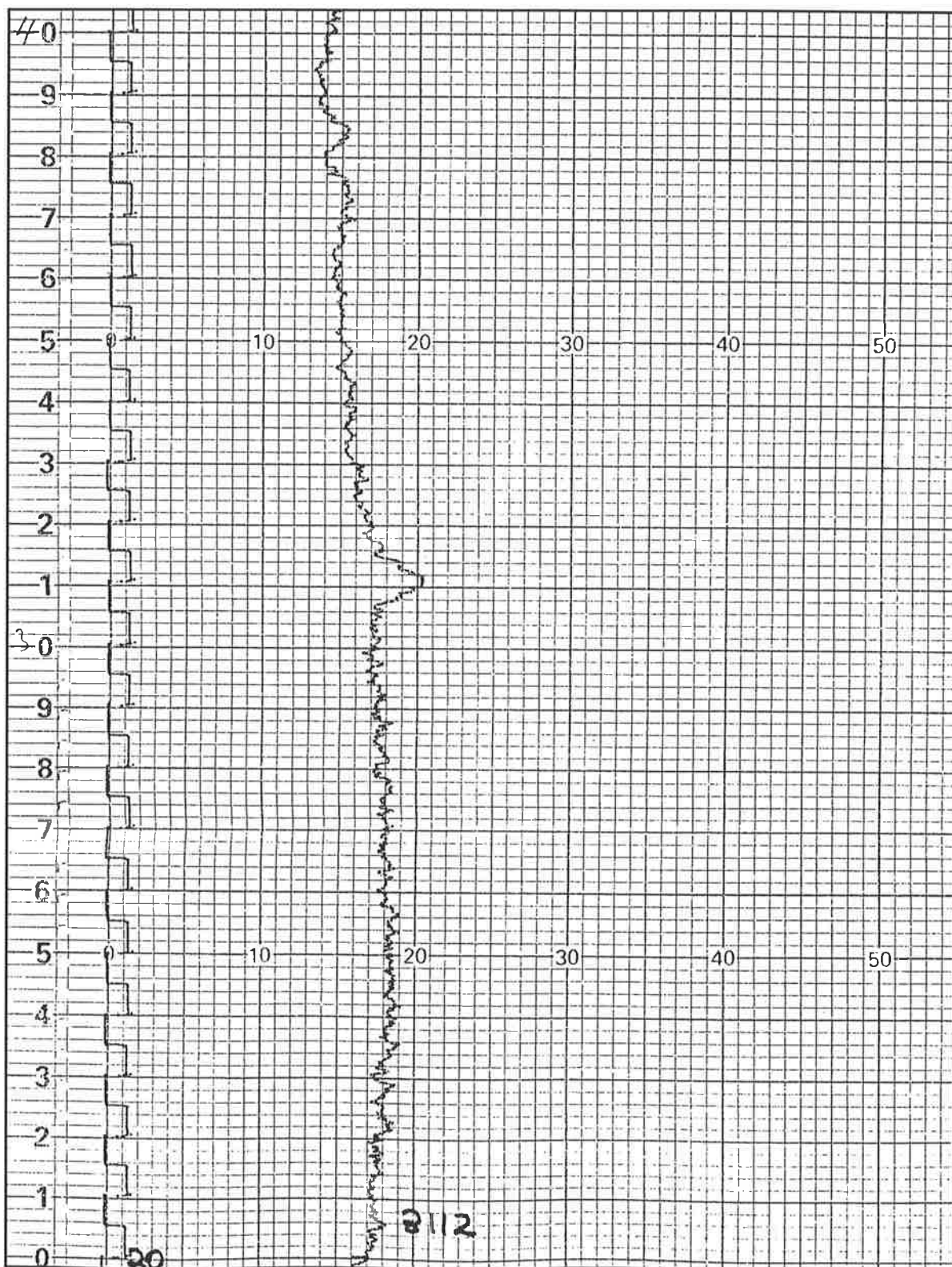


XRPD Trace: Weld 2-1, $2\theta = 65^\circ$ to 85°



Weld 2-2

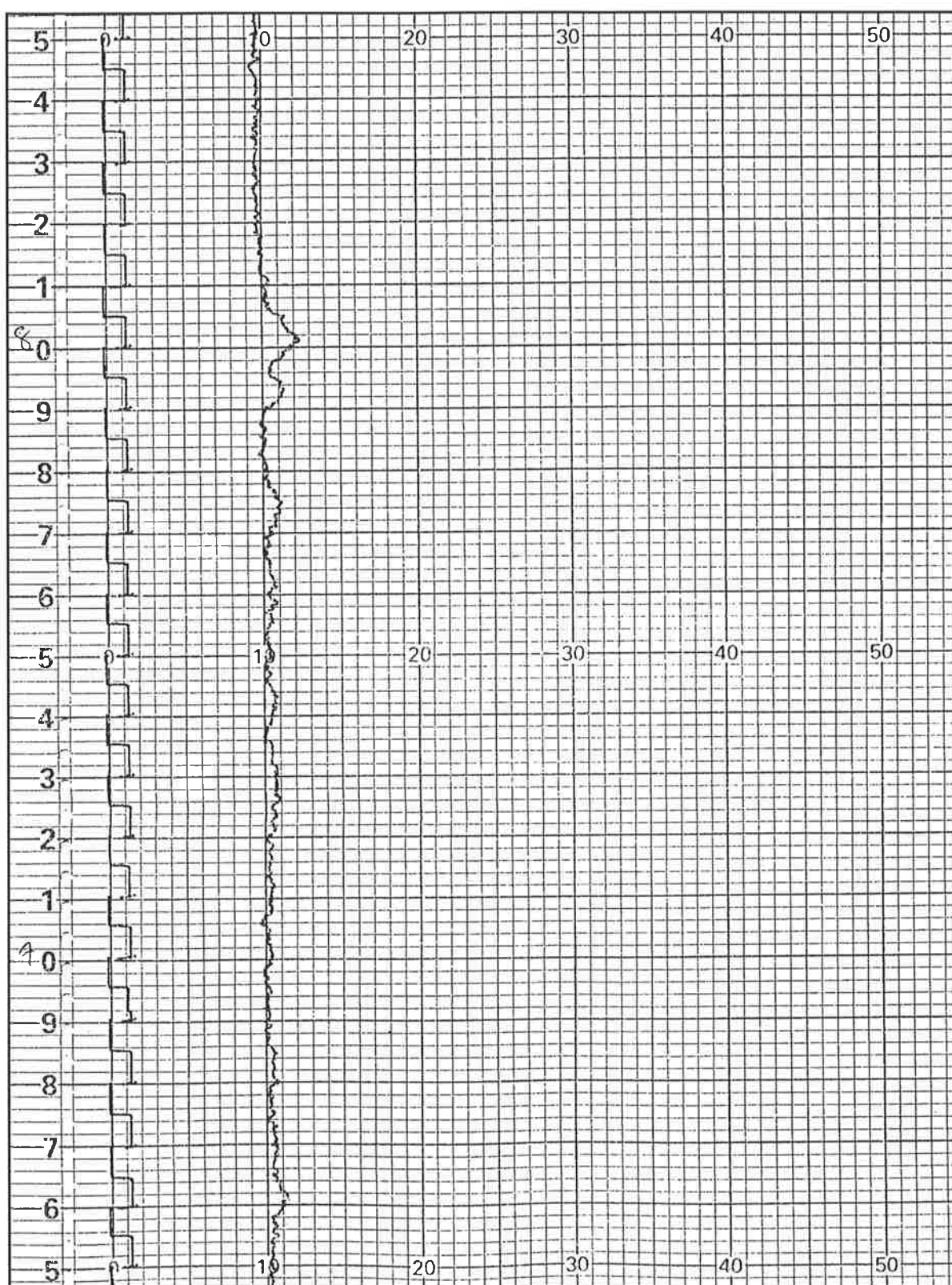
XRPD Trace: Weld 2-2, $2\theta = 20^\circ$ to 40°



XRPD Trace: Weld 2-2, $2\theta = 40^\circ$ to 65°

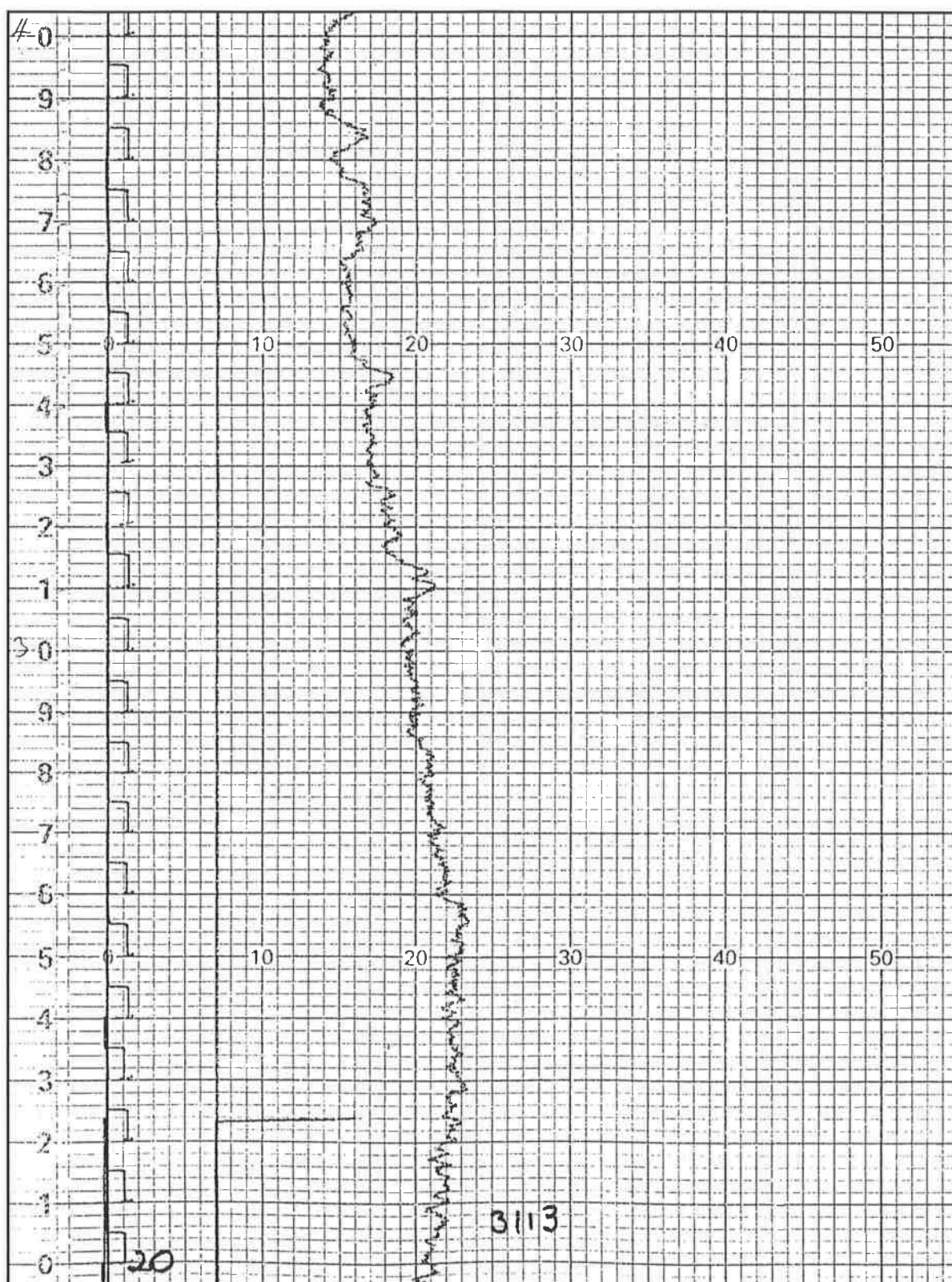


XRPD Trace: Weld 2-2, $2\theta = 65^\circ$ to 85°

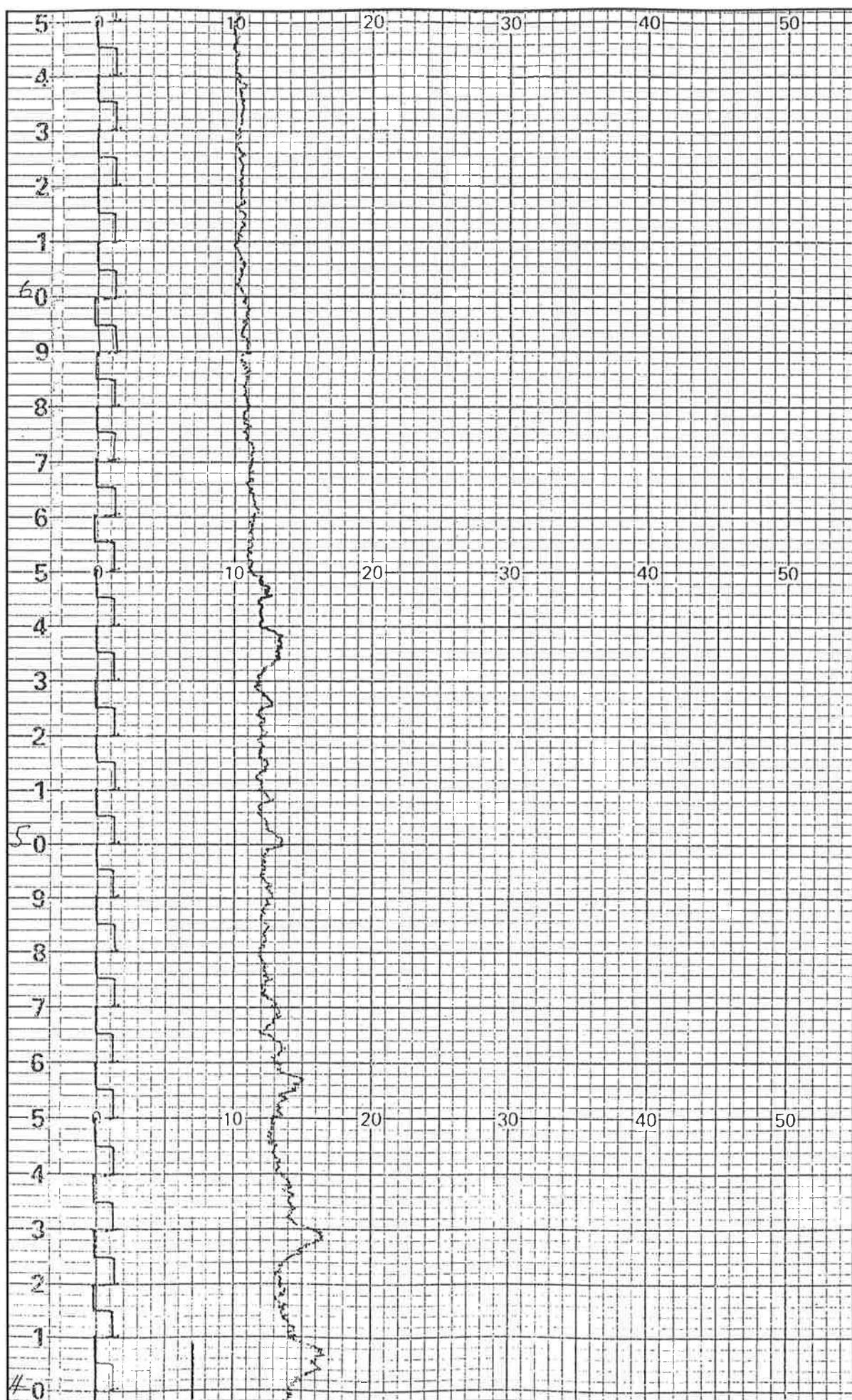


Weld 2-3

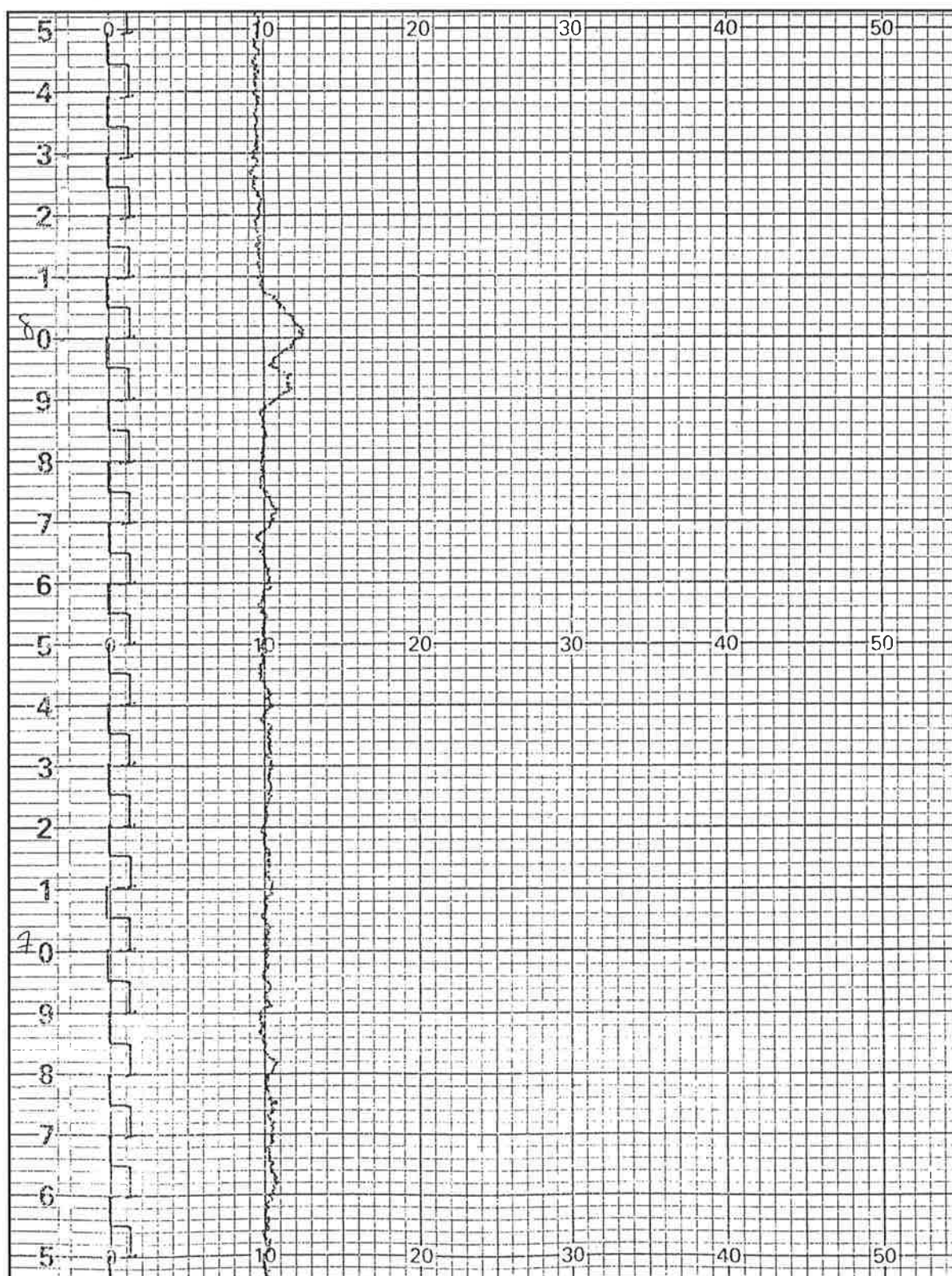
XRPD Trace: Weld 2-3, $2\theta = 20^\circ$ to 40°



XRPD Trace: Weld 2-3, $2\theta = 40^\circ$ to 65°

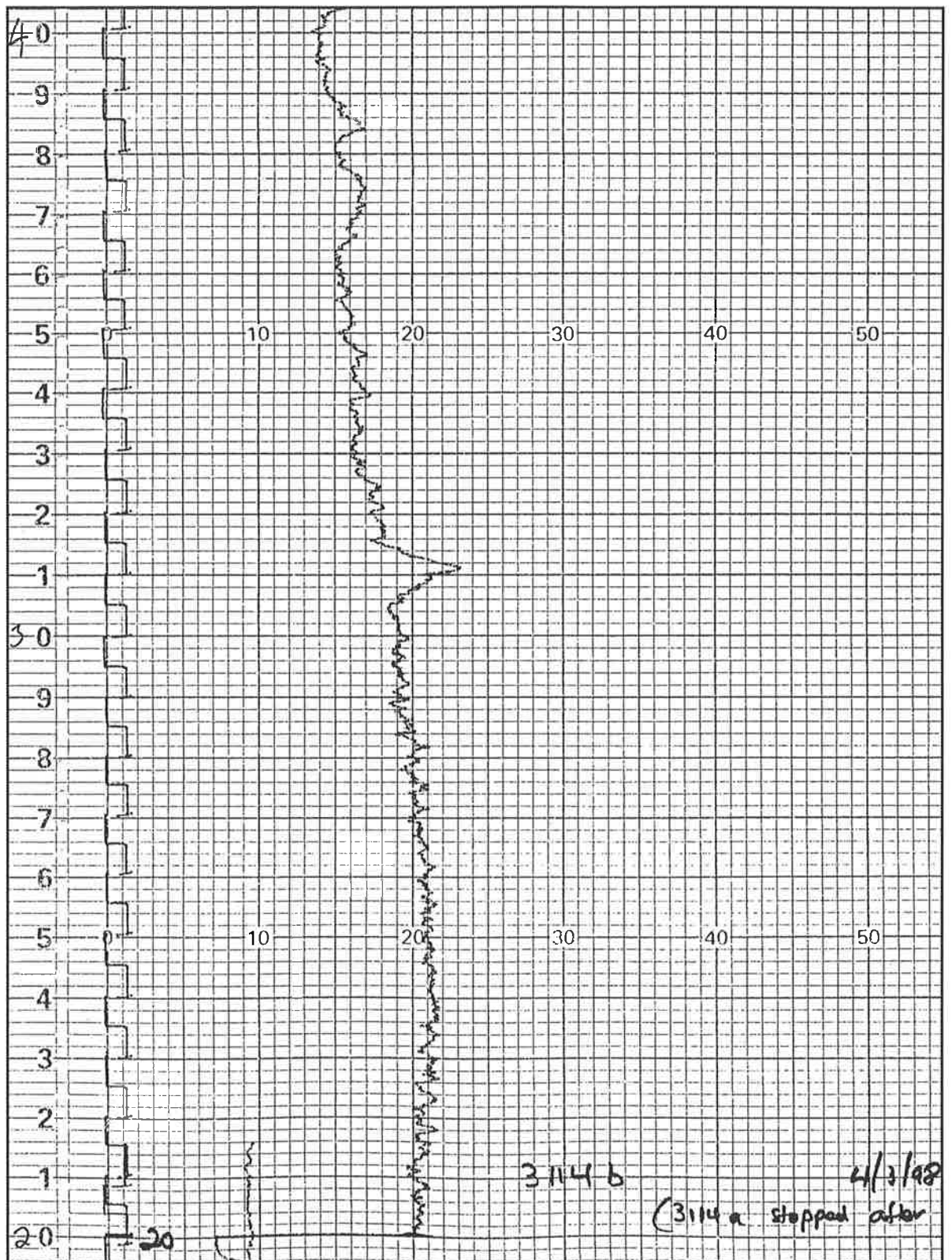


XRPD Trace: Weld 2-3, $2\theta = 65^\circ$ to 85°

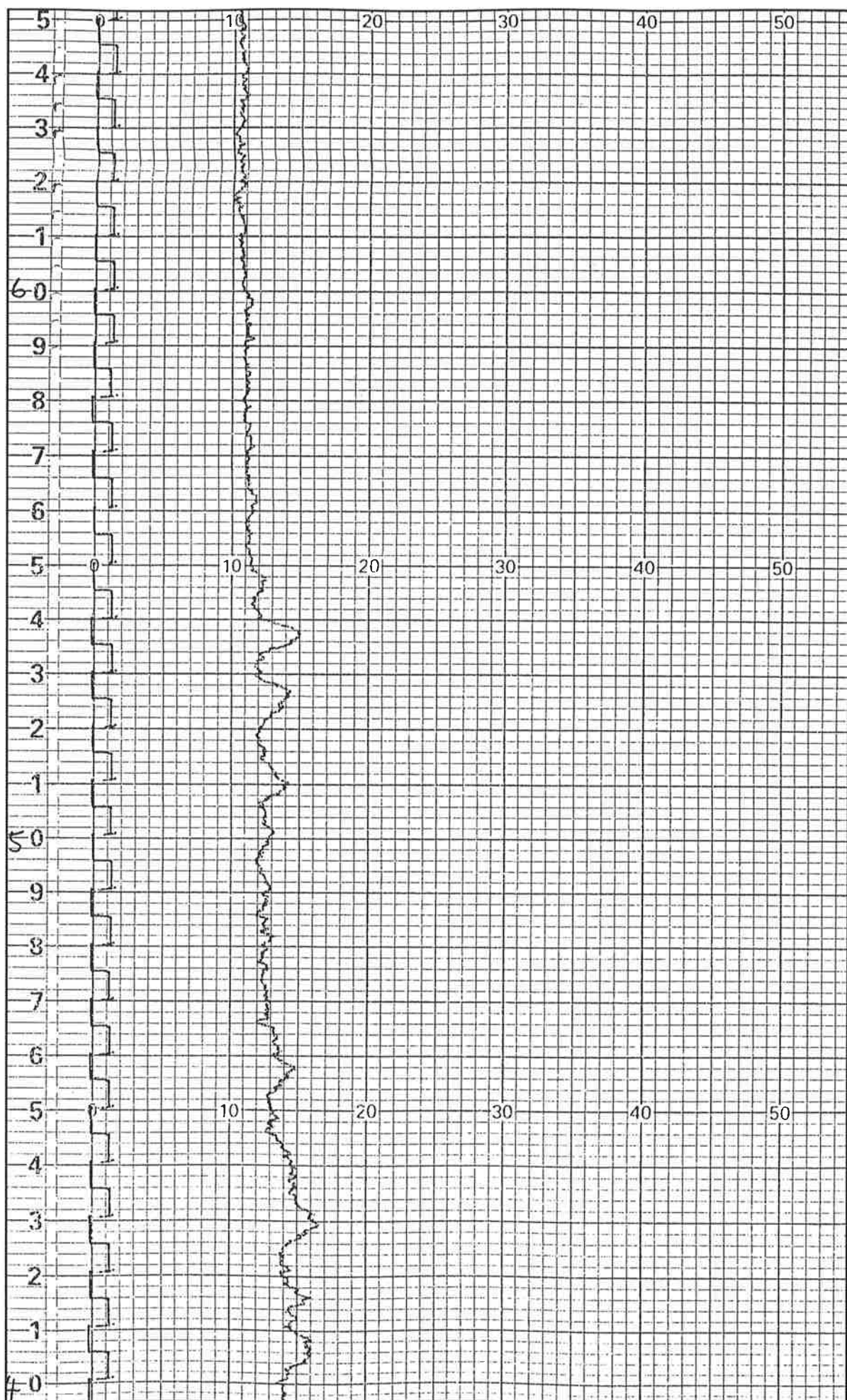


Weld 2-4

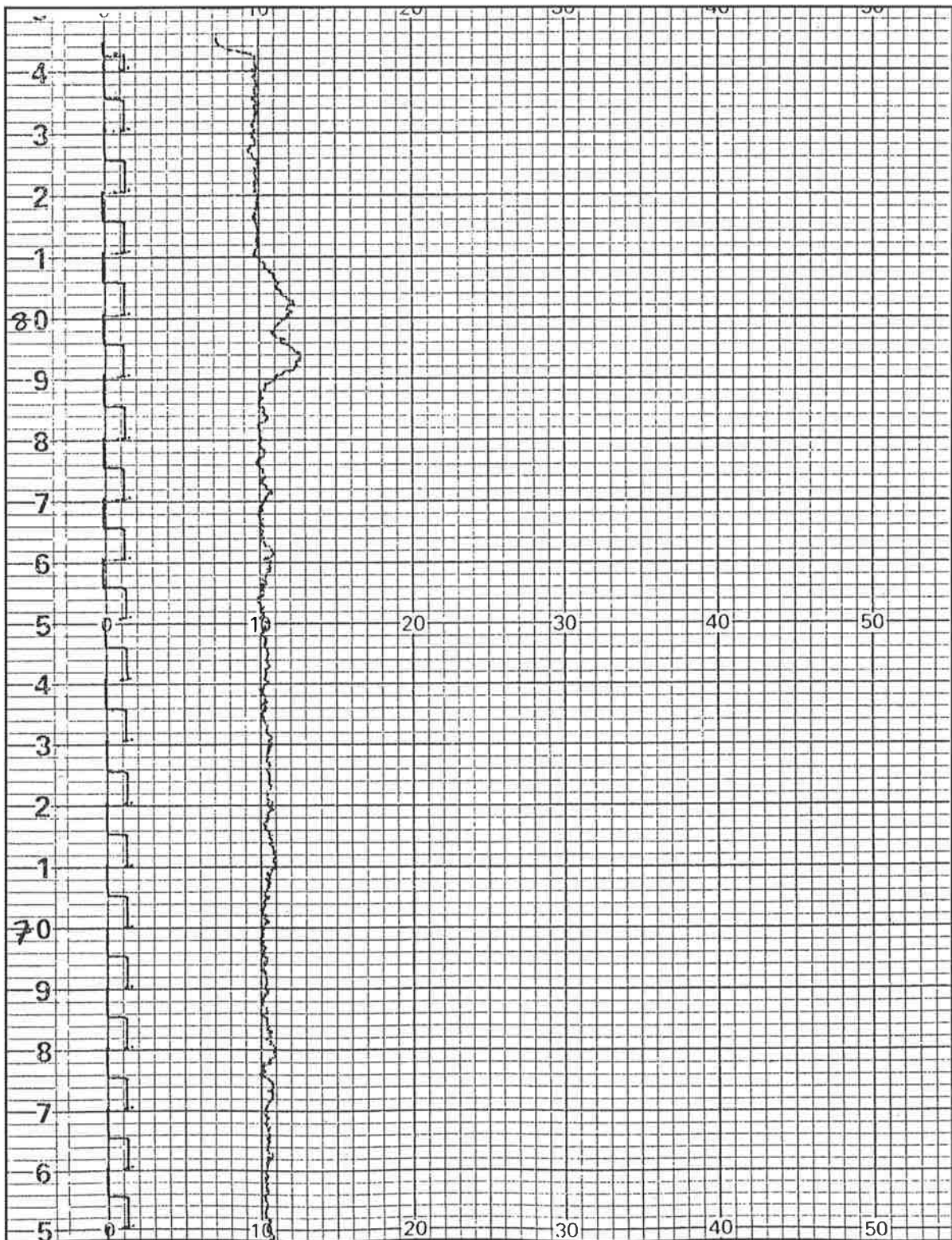
XRPD Trace: Weld 2-4, $2\theta = 20^\circ$ to 40°



XRPD Trace: Weld 2-4, $2\theta = 40^\circ$ to 65°



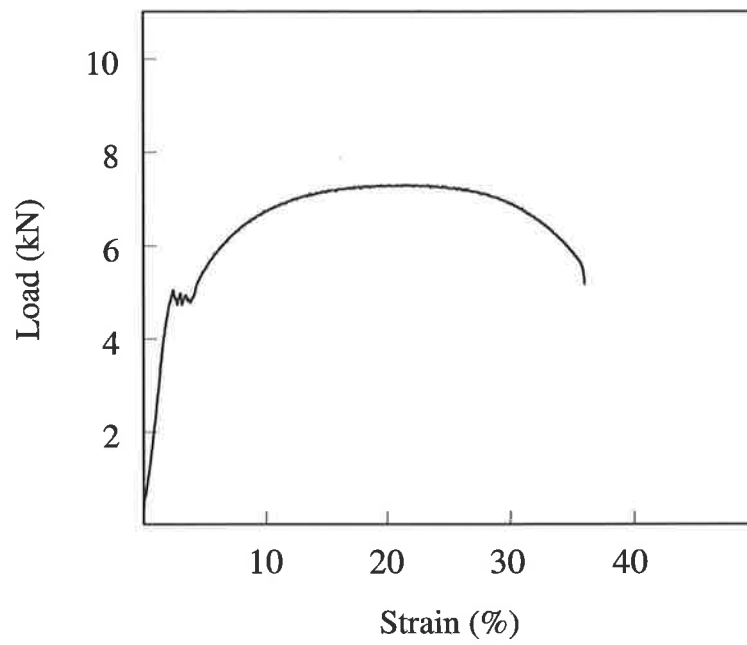
XRPD Trace: Weld 2-4, $2\theta = 65^\circ$ to 85°



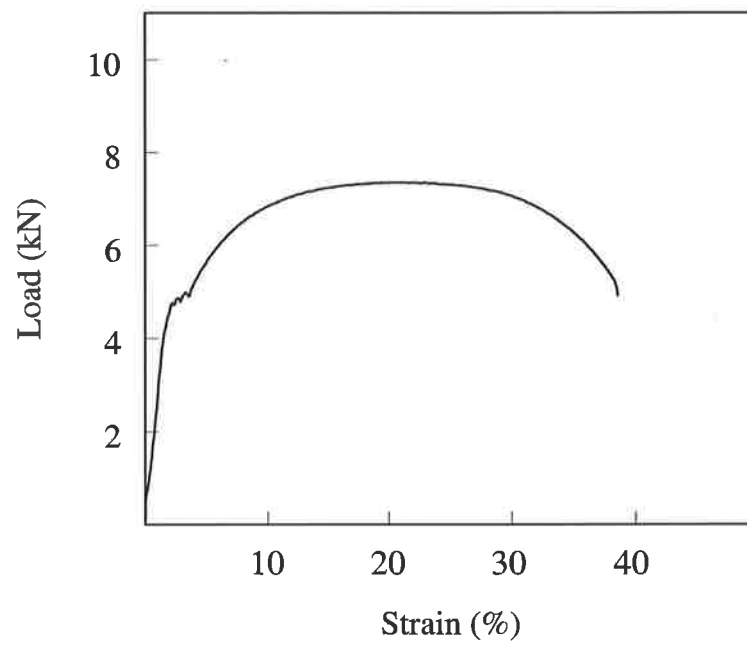
APPENDIX D

Tensile Load - Strain curves.

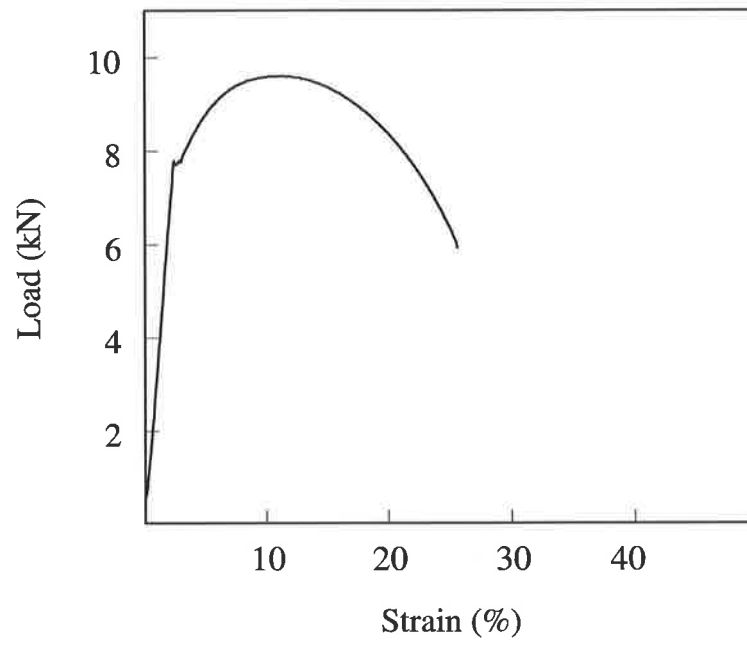
Grade 250 Base Material (a)



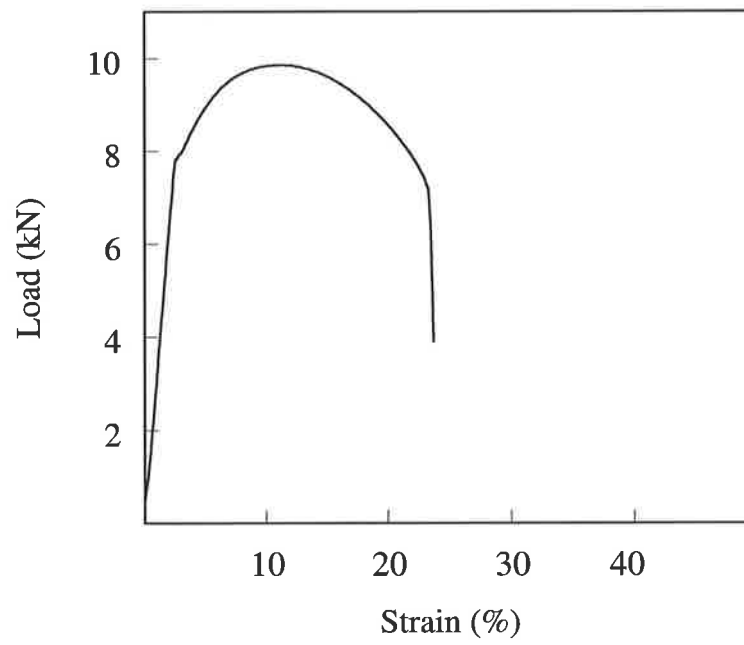
Grade 250 Base Material (b)



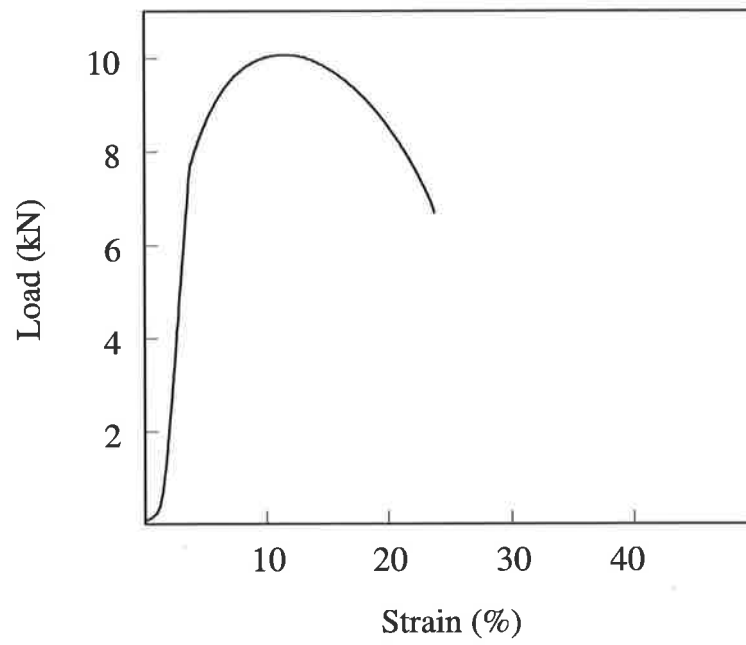
Weld 1-1a



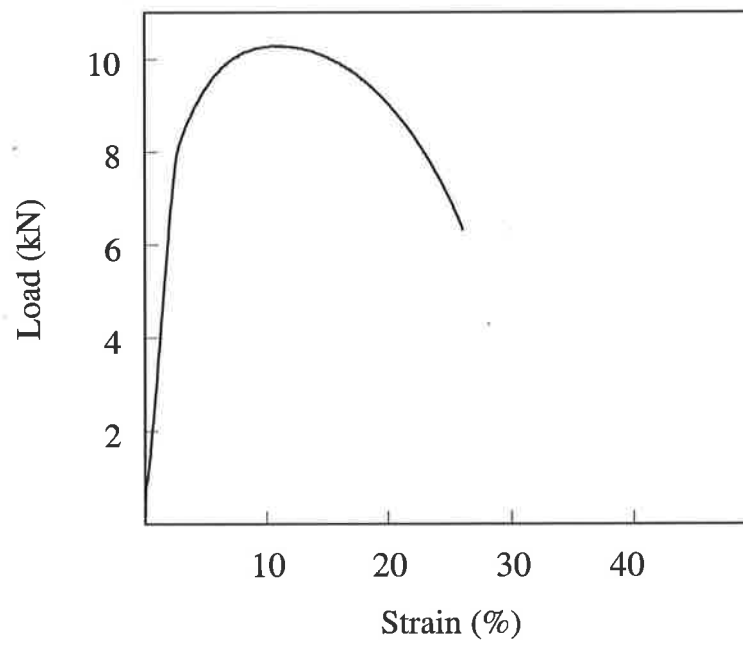
Weld 1-1b



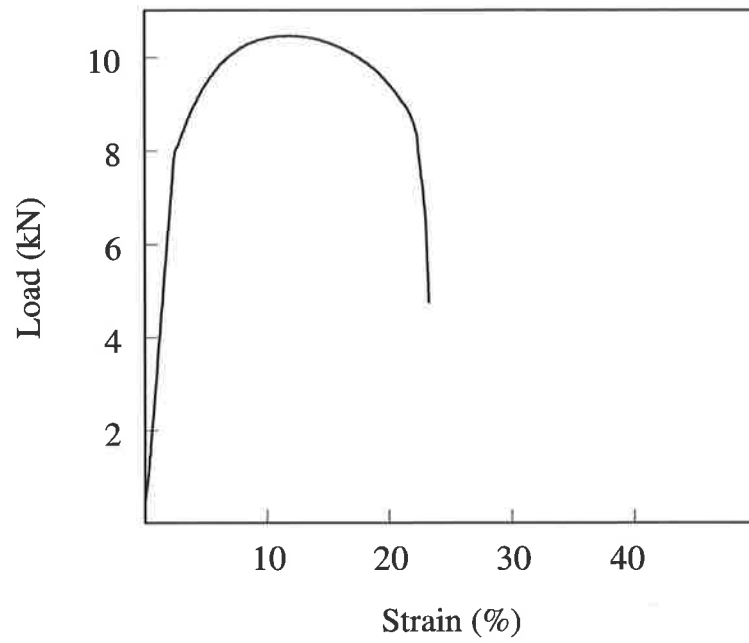
Weld 1-2a



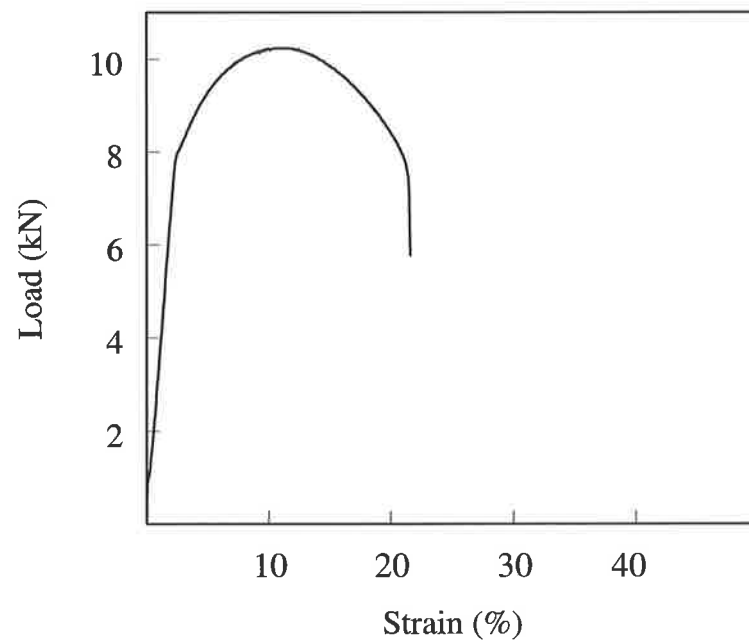
Weld 1-2b



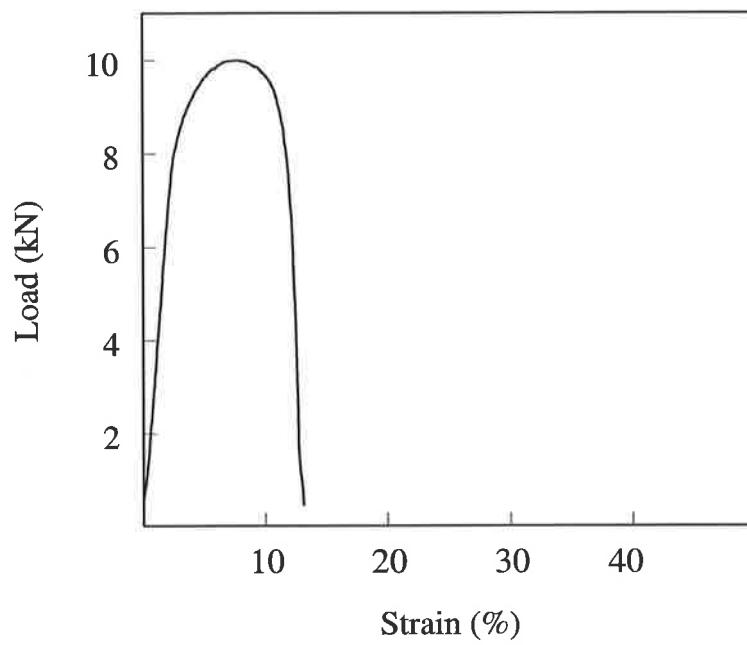
Weld 1-3a



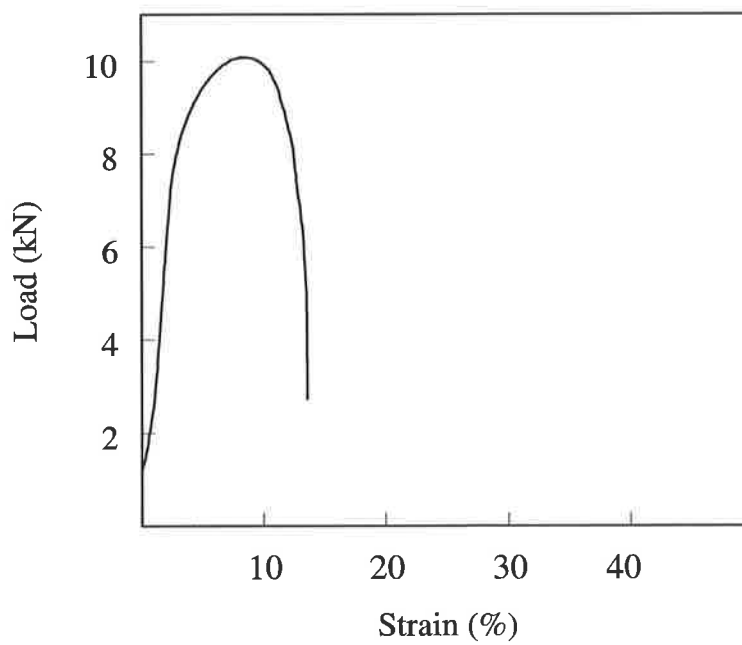
Weld 1-3b



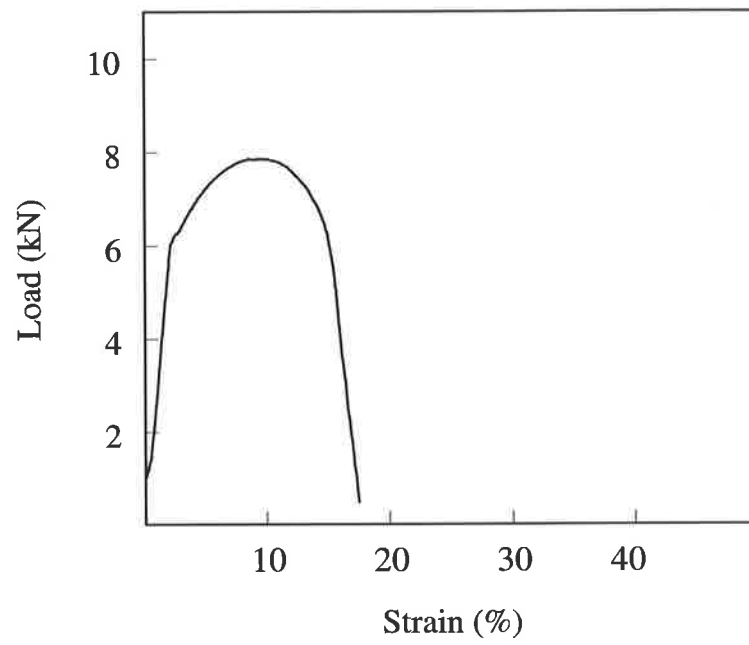
Weld 1-4a



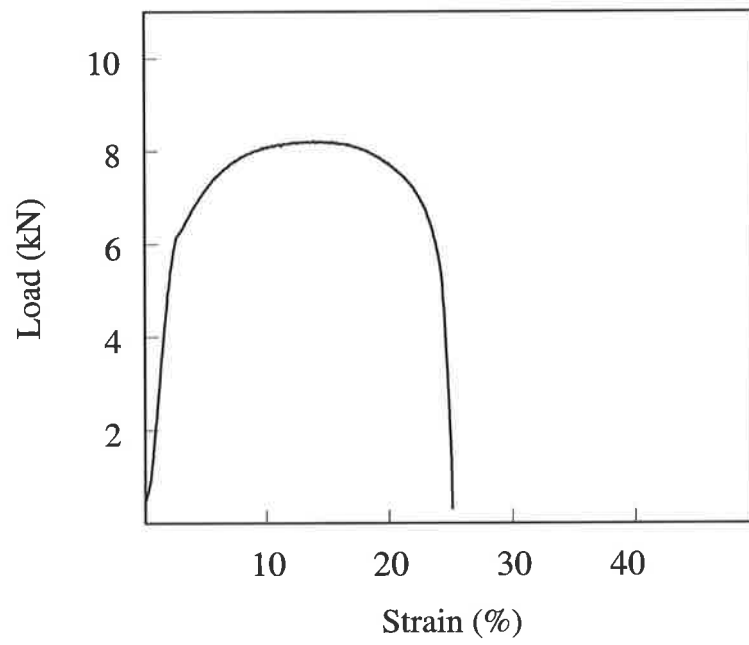
Weld 1-4b



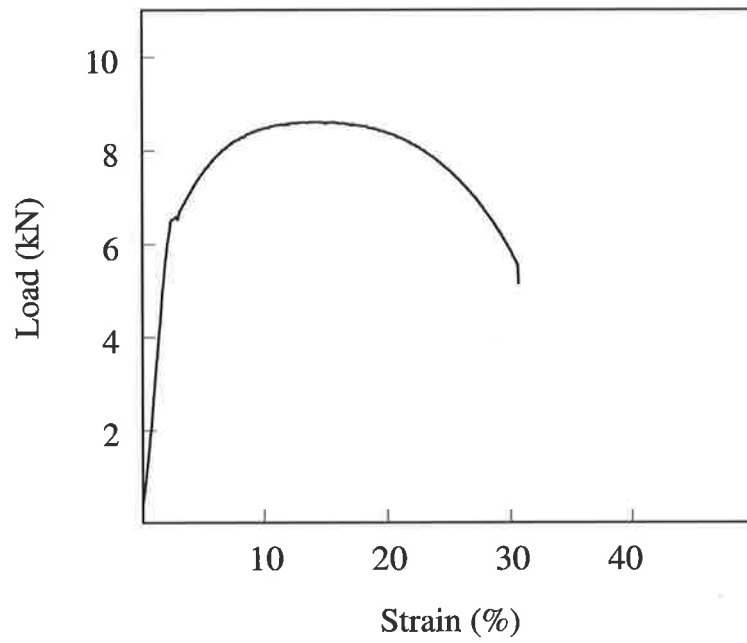
Weld 2-1a



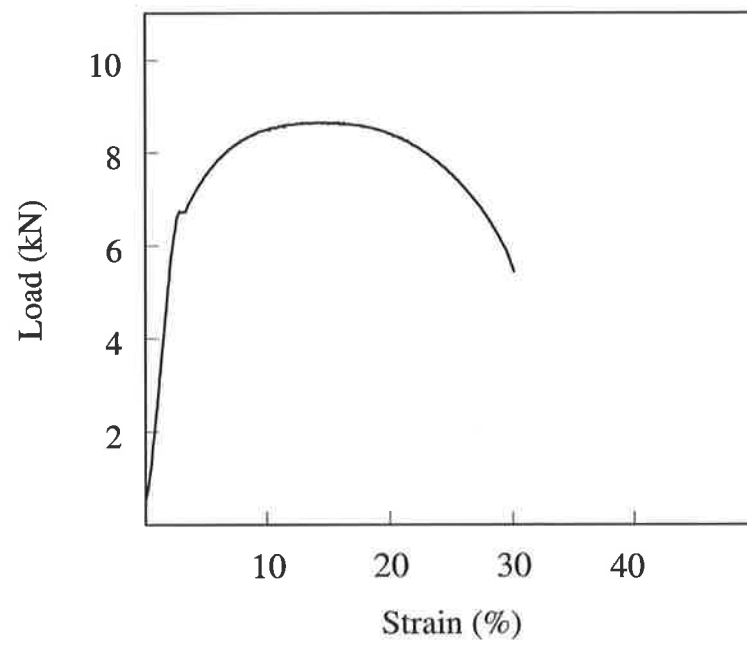
Weld 2-1b



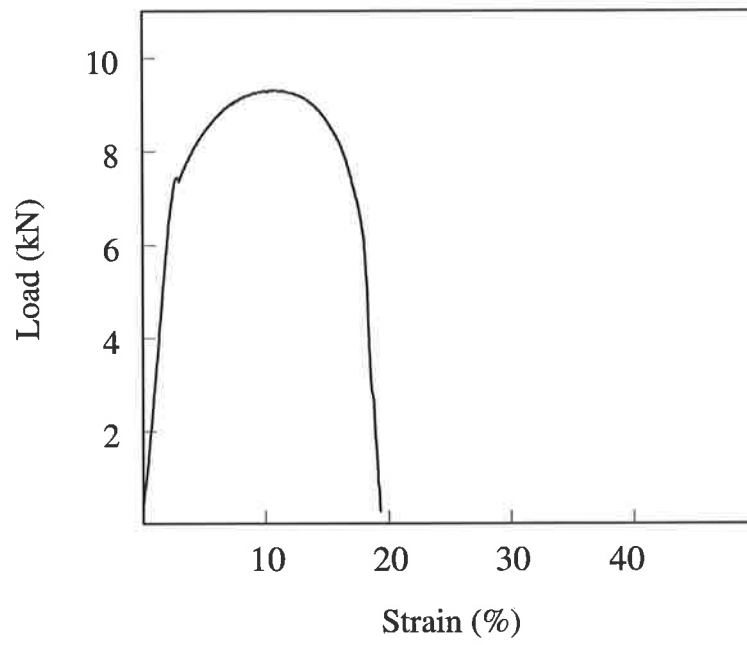
Weld 2-2a



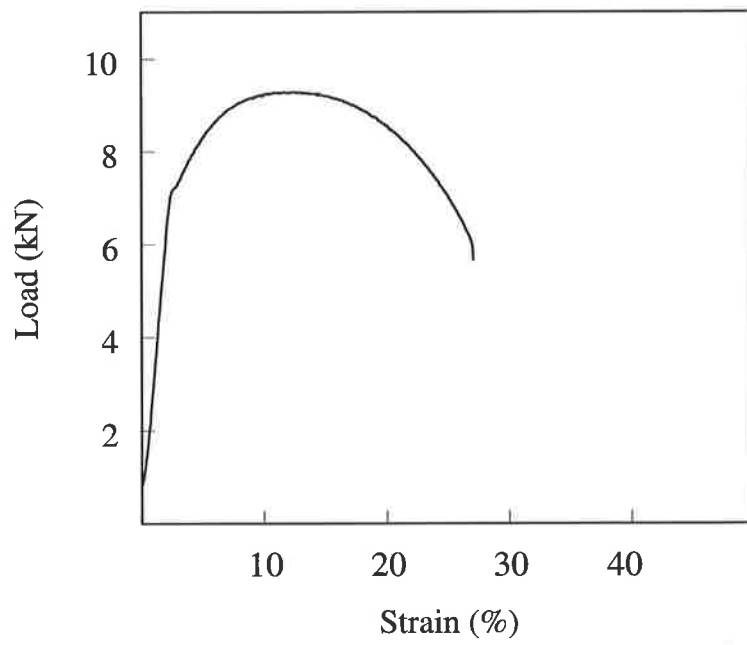
Weld 2-2b



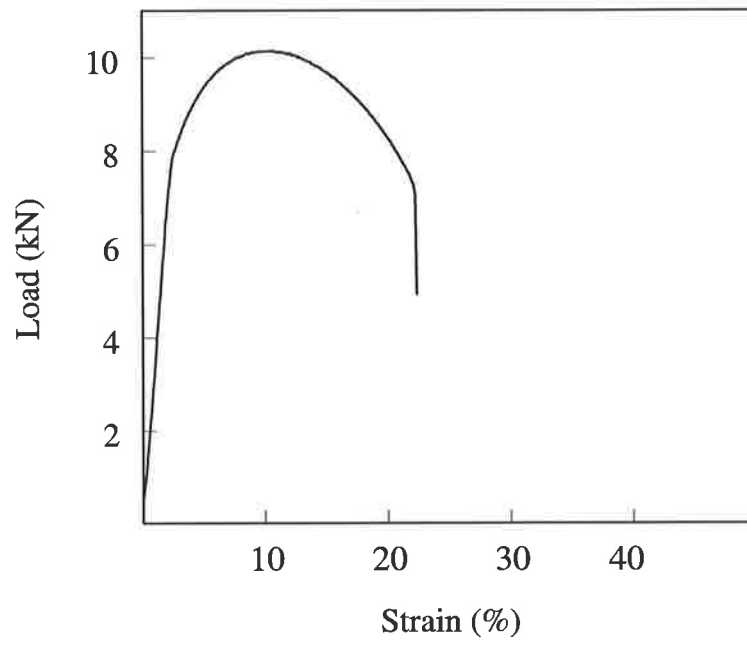
Weld 2-3a



Weld 2-3b



Weld 2-4a



Weld 2-4b

