SOLIDIFICATION CRACKING IN LOW ALLOY STEEL WELD METALS

by

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Synopsis

A study has been made of the effects of material composition and welding variables, particularly speed, on weld metal solidification cracking. This has involved low alloy steels to specifications En24, SAE4130, ASTM A387B, and several aircraft steels. Existing relationships between composition and cracking have been critically examined, and an improved expression, based on regression analysis, has been developed using data relating to some 80 steels. Practical cracking tests were carried out using the Huxley test, and a comparison of these results with the predictions of the regression analysis confirmed the crack-promoting influences of sulphur, phosphorus and carbon, and isolated the important crack-inhibiting effect of oxygen.

The influence of welding speed was found to be complex and varied with the composition of the steel. In general an increase in speed is detrimental at low speeds, but becomes beneficial when the speed is raised. This phenomenon is explained by a model in which changes of speed are shown to cause opposing effects, which interact to give the characteristic relationship.

The morphology of inclusions was also examined, and crack susceptibility was found to be associated with sulphide films.

The practical significance of the work is also discussed. Apart from exposing important new effects, the results facilitate the assessment of chemical effects on cracking;

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and demonstrate the importance of selecting welding variables appropriate to the particular materials being used.

Two different tests have been used in the work for specific purposes. The results demonstrate the overwhelming importance of selecting a test suitable for the situation under examination.

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Contents

1. Introduction

The development of low alloy quenched and tempered steels has been stimulated very vigorously by the growing demands of the aerospace industry. These demands are for materials offering a high strength to weight ratio combined with good mechanical properties over a wide range of temperature (from sub-zero levels to 500°C and above). A group of steels widely used in such applications is that employing various small additions of nickel, chromium, molybdenum and vanadium, since these materials may be heat treated to give tensile strengths up to 1.9kN/mm^2 (120 TSI). Steels offering strength levels of this order have found their main applications in the manufacture of aircraft undercarriage components, where the weight savings achieved provide for greater payloads and therefore increased commercial benefits. They are also used for rocket motor cases on account of their desirable strength to weight ratio and their ability to withstand high gas pressures. A critical review of the uses of high strength steels has been made by Ineson[1], who also shows that there are many other potential uses to which these materials could be applied. These include bearings, gears, springs, automobile parts, railways, submarine and torpedo hulls, mining equipment and oil installations. The use of high strength steels for ordnance items such as gun barrels and armour is also developing, and other low alloy steels, such as the ferritic creep-resisting steels are widely used for pressure vessels, turbine blades and generating plant.

Among the important groups of steels are the Ni-Cr-Mo and Cr-Mo types e.g. ASTM A387B, En 24 and SAE 4130, the last two being of special interest in the aircraft industry where the quality requirements are most demanding. These materials possess the strength and hardness of a martensitic structure together with toughness and high temperature stability conferred by the alloy elements. ASTM A387B is primarily a pressure vessel steel, designed for service at elevated temperatures.

The usefulness of high strength steels is heavily dependent on good weldability, for a high quality welded joint is essential if the full advantages are to be gained from the advanced properties. The production of sound welds with adequate mechanical properties presents many difficulties, and the weldability problem has done much to hinder the wider use of high strength steels. More detailed study of the weldability of these materials is among the recommendations in Ineson's review.

One of the serious problems in welding is the occurrence of high temperature cracking. This is an insidious defect which may occur in both the weld metal, as solidification cracking, and in the heat affected zone (H.A.Z.) as H.A.Z. hot cracking or liquation cracking. Weld solidification cracking occurs in several forms; it may appear as a continuous centre-line crack, extending for several centimetres, or take the form of isolated fissures, a few millimetres in length, either perpendicular to the fusion boundary or along the centre line.

The damage caused by solidification cracking can be very extensive. It includes the direct effect of large cracks in rendering joints unserviceable, as well as the indirect effect of small fissures in initiating cold cracks, fatigue failure, brittle fracture and stress corrosion. Indeed small cracks may be especially troublesome since they are not easily seen with the naked eye and may require complex or time-consuming techniques, e.g. radiography or microscopy, for their detection. More seriously, they may pass unnoticed until some greater defect which has developed from them causes a major breakdown.

The tendency to weld metal solidification cracking is closely associated with the chemical composition of the material, the impurities sulphur and phosphorus being particularly detrimental. Among a number of alloying elements which also affect crack susceptibility, carbon is particularly significant, a slight rise in the carbon level being reported to give an appreciable increase in cracking. Welding variables are also known to exert a major influence on the occurrence of cracking, but the effects appear to be extremely complicated and very little understood.

Several theories of solidification cracking have been developed, and these explain the general effect in mechanical and metallurgical terms. Moreover, a number of relationships have been suggested to express the crack susceptibility of a material in terms of its chemical composition. These relationships are often specific to a particular application or experiment, however, and there appears to be a lack of general principles which may be applied to other situations. This is particularly true for the effect of process variables on cracking, which has received only scant attention. Furthermore the interaction between the effects of composition and welding variables appears not to have been studied at all. This lack of information often prevents the adoption of a scientific approach to welding fabrication, and techniques are developed on an empirical basis, which is often costly and inconvenient.

It was the aim of the present work, therefore, to quantify the effects of composition and procedural variables, particularly welding speed, on solidification cracking in low alloy steels; and if possible, to present some guidance for reducing the occurrence of the problem. Also an attempt was to be made to explain the variations observed, so that the information obtained could be extended to other situations.

The first stage of the project included a review of the relationship between chemical composition and crack susceptibility under various conditions of welding speed and current. Sulphur, phosphorus and alloy element effects were studied, and the concept of a simple linear relationship between crack susceptibility and material composition was found to be a sound one. It enabled the relative chemical effects to be separated and estimated, and provided a simple means of predicting crack susceptibility with a known degree of confidence; thus any anomolous behaviour could be detected. In this way an important new effect was highlighted, namely the potency of oxygen as a crack inhibitor.

Welding speed was found to exert a complex effect on solidification cracking: a steady increase in welding speed first raised the cracking tendency until a maximum was reached, after which further increase in the speed reversed the effect. The precise form of the cracking-speed relationship varied from steel to steel. A model has been developed to explain these effects, which result from the speed changes causing both beneficial and detrimental effects on cracking, the balance of these determining the exact form of the cracking-speed relationship.

The results of this investigation are therefore of particular significance for welding practice in two ways:

a) The effects on crack susceptibility of changes in the concentrations of individual chemical elements may be easily assessed; and

b) The results demonstrate how welding variables may be manipulated to minimise the solidification cracking tendency in materials of varying alloy and impurity contents.

2. Literature Survey

2.(1) Introduction

Weld high temperature cracking occurs in both ferrous and non ferrous metal joints, in both the weld metal and the heat-affected zone adjacent to the weld. The cracks form either during solidification, as in the case of weld metal solidification cracking, during re-solidification of liquated metal (in the H.A.Z.) or just after solidification is complete.

Cracking takes place to accommodate shrinkage strains which result from the weld thermal cycle, and occurs at planes of weakness and low ductility in the material. These planes of weakness may be caused by thin films of non-metallic segregates, or by segregation of alloy elements, leading to zones of material with a depressed freezing point. Sub-solidus cracking occurs during the "ductility dip" experienced by some materials as the temperature falls after freezing. An attempt to classify the various forms of high temperature cracking has been made by Hemsworth et.al[2] who list six cracking situations: cracking due to non-metallic films in primary weld metal, re-heated weld metal and the H.A.Z., and sub-solidus "ductility dip" cracking in the same three zones of the welded joint. This classification is illustrated in Fig. 1.

Weld high temperature cracking in high strength steels became a severe problem as long ago as the 1930's when these materials were applied in the aircraft industry for their

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attractive mechanical properties. During welding, however, cracks formed in both the H.A.Z. and weld metal, which resulted in a good deal of comment and research, particularly in Germany and the United States of America [3,4,5,6]. Examination of the weldability of many heats of plain carbon and Cr-Mo steels revealed that the cracking tendency was largely dependent on the carbon and impurity contents[3,4]. This early work, and the increasing magnitude of the problem resulting from the greater use of high strength steels, stimulated many other investigators in the study of high temperature cracking, particularly in thin sheet material, where the problem had been most severe[7,8,9], although thicker materials were also studied[10]. Most authors report on the detrimental effects of sulphur and phosphorus, although the relative importance of the two elements has been subject to disagreement, no doubt in part due to the differences in materials examined and in the welding situations. In more recent papers, the effects of the alloying elements have been covered[7,8,11], and these are usually said to act by influencing the behaviour of the main crack promotors, sulphur and phosphorus.

Although the problem of high temperature cracking, particularly solidification cracking, has been treated mainly in chemical terms the importance of welding procedure has been acknowledged by some writers. An early paper by Stout et.al[12] discussed heat input effects, and Terry and Tyler[9] examined the influence of welding speed on the T.I.G. (tungsten inert gas) welding of SAE 4130 steel sheet. The results of these and other works demonstrate that high temperature cracking is a function of welding procedure as well as of material characteristics - a conclusion of great technological importance.

The theoretical treatment of solidification cracking came after the observations and to some extent, after the solution. Although an early attempt was made to relate crack susceptibility to high temperature strength[4], the bulk of the theoretical work was produced after the war, beginning with the problem of hot shortness in aluminium alloy castings. Investigations on Al-Si alloys led to the conclusion that hot shortness was a function of silicon content, and was related to the solidification temperature range of the particular alloy[13,14]. These metallurgical considerations, together with the thermal strain effects, form the basis of most theories on solidification cracking. The strain situation obtaining is a function of the temperature distribution and geometry of the joint, the former being directly dependent on the welding procedure.

Thus the important factors influencing weld solidification cracking are chemical composition, welding variables, metallurgical effects and restraint, and existing knowledge of these is discussed in the literature review.

The final section of the literature review deals with the methods available for assessing the susceptibility of welds to solidification cracking, and on this basis a choice was made of methods to be employed in the investigation.

2.(2) The Classification and Morphology of Solidification Cracks

Solidification cracks are characterised by their intercrystalline appearance[2,4,15,16]. They may occur perpendicular to the welding direction as small fissures a few millimetres long just inside the fusion boundary; or they may take the form of centreline cracks several centimetres long, either intermittent or continuous[3,17,18,19]. There would not appear to be any metallurgical distinction between centreline and transverse cracks, but metallurgical and morphological differences occur in both. Hemsworth et.al[2] have classified the morphological varieties into three groups as follows:-

a) Cracking along cell boundaries,

b) Cracking along the boundaries of blocks of cells i.e. grain boundaries,

c) Interdendritic cracking.

In alloy steel welds, cracks may be associated with intergranular non-metallic inclusions such as sulphides[20] or even carbides[2,15], also with phosphorus - rich segregates[21]. Wilkinson and Cottrell[22] give an illustration of intergranular cracks associated with inclusions, and more definite evidence for solidification cracking associated with segregation is given by the electron probe studies of Randall et.al[23]. Intergranular sulphide films were formed in Ni-Cr-Mo steel welds, and at crack tips sulphur, phosphorus and silicon were found to be heavily segregated. Metallographic studies by Sopher et.al[24] on Ni-Cr-Mo steel

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weld metals and weld simulation specimens also showed intergranular sulphur-rich segregates which were associated with cracks and low ductility in heats high in sulphur and phosphorus. In purer metals, neither cracking nor sulphides were found.

2.(3) Composition Effects

There appears to be universal agreement that high temperature cracking is aggravated by the presence of sulphur, possibly phosphorus, and by high carbon contents. This was noted as long ago as 1914[25] and the significance of the problem was made clear in several pre-war reports[3,4,26,27], concerning the weldability of carbon steels and Cr-Mo aircraft steels such as SAE 4130, in which the problem was particularly severe. The literature review of Spraragen and Claussen describes the pre-war work in this subject. Copper was found to be detrimental in Cr-Mo steels[26,27], but oxygen, hydrogen and nitrogen were not considered important in low concentrations although oxides were thought by some authors to be beneficial and by others to be detrimental[28-31]. Podgaetskii cites oxygen as a crack inhibitor[32,33], both in the steel and when added to the shielding gas, but he later suggests that oxides may be more important than elemental oxygen[34].

Most of the work has been centred on the important effects of sulphur and phosphorus. Reeve[29] describes experiments on cracking in mild steel welding using a simple fillet weld test. Sulphur is reported to have given rise to

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cracking when present in excess of 0.07% but no deleterious effect due to phosphorus was noted. Later American results by Sopher et.al[10,24] on high strength steels (e.g. SAE 4340), using the Lehigh restraint test[12,23] and high temperature tensile tests in which the specimen was tested during cooling to simulate a welding situation, indicate that both sulphur and phosphorus are detrimental to crack resistance, their effect becoming greater with increasing carbon contents. This is in agreement with the earlier work of Muller[3]. In the work of Sopher, agreement was obtained between the restraint cracking tests and the hot tensile tests, inasmuch as the most crack-sensitive materials exhibited the lowest ductilities at temperatures up to 1500°C. Also a possible deleterious effect due to boron was observed[24] as well as an improved crack resistance due to cerium[10]. Jones[35] investigated the solidification cracking of low alloy steel welds using the Murex test[36] and came to the general conclusion that a rise in the manganese to sulphur ratio, $\frac{Mn}{S}$, up to 50 reduced the crack susceptibility, but that nickel was detrimental. Copper also appeared to be harmful in amounts of about 1% but there were indications of a beneficial effect due to cerium in accord with Sopher. Borland[37], working on RS 140 1% Cr-Mo-V steel sheet found little significance in the $\frac{Mn}{S}$ ratio above 50 but confirmed the deleterious effect of phosphorus. This work was carried out using the Huxley[8], patch[38] and Houldcroft[39] tests.

An attempt to quantify the effects of impurities was made by Wilkinson and Cottrell[22], working on 1% Cr-Mo steel

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sheet. They analysed their own work, together with some early German results[3,4] and produced a criterion for freedom from weld cracking:-

$$(S(\%) + P(\%)) < 0.007/C(\%)$$

This work considers both weld metal and H.A.Z. cracking, as indeed do many other investigations since the two forms of defect are very similar in nature and frequently occur together. A more ambitious formula[7], obtained from miniature Houldcroft weld hot crack tests[39] on a number of low alloy sheet steels was given by Wilkinson et.al. thus:-

$$\text{H.C.S.} = \frac{C\left[S + P + \frac{Si}{25} + \frac{Ni}{100}\right] \times 10^3}{3Mn + Cr + Mo + V} < 4$$

where H.C.S. is termed the "Hot Crack Susceptibility," a composition factor which was found to correlate with the test results and has often been used to predict the propensity of a material to hot cracking[20,40]. The levels of the elements are given in weight percentages. The results were said to be in good agreement with production experience, such that steels with an H.C.S. value less than 4 showed good crack resistance. This formula includes the influence of the alloying elements, which is usually small (relative to sulphur and phosphorus) and has been explained in terms of their effect on phosphorus and sulphur behaviour during solidification[8,41]. Austenite-forming elements are cited as crack promoters on account of the lower solubility of phosphorus in that phase and ferrite formers are held to be

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beneficial. The enhancement by austenite formers of the effect of phosphorus on crack susceptibility is illustrated by the formula:-

$$\text{CRACK SUSCEPTIBILITY} \propto P(\%)\,(C(\%) + \frac{Ni(\%)}{30}) \quad \text{(Ref 8)}$$

The beneficial effect of the austenitiser manganese, predicted by Wilkinson's formula is discussed in section 2.6.2.

Solidification cracking formulae have been worked out by Ostrovskaya[11] in the U.S.S.R., in this case three formulae are used, each applicable at a different range of carbon contents, suggesting that the carbon level affects the influence of the other elements:

a. For C between 0.09% and 0.14%:

$$C\ eq = C + 2S + \frac{P}{3} + \frac{Si-0.4}{10} + \frac{Mn-0.8}{12} + \frac{Ni}{12} + \frac{Cu}{15} + \frac{Cr-0.8}{15}$$

b. For C between 0.14% and 0.25%:

$$C\ eq = C + 2S + \frac{P}{3} + \frac{Si-0.4}{7} + \frac{Mn-0.8}{8} + \frac{Ni}{8} + \frac{Cu}{10} + \frac{Cr-0.8}{10}$$

c. For higher carbon contents, the following is suggested:

$$C\ eq = C + 2.5S + \frac{P}{2.5} + \frac{Si-0.4}{5} + \frac{Mn-0.8}{6} + \frac{Ni}{6} + \frac{Cu}{8} + \frac{Cr-0.8}{10}$$

C eq is the so-called "carbon equivalent" value and the element contents are expressed as weight percentages. The crack susceptibility is then a function of the calculated carbon equivalent. These formulae were derived from data on low alloy constructional steels containing up to 0.22%

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carbon, and reflect the innocuous nature of Si, Mn and Cr when present at low levels, although it is not clear which welding processes and tests were used. Another interesting formula, based on the effect of phosphorus, has been a advanced by Cottrell[42]:

$$\text{C.S.F.} \propto P\left[(C + 0.142Ni + 0.282Mn + 0.2Co - 0.14Mo - 0.224V) + 0.195S + 0.00216Cu\right] \times 10^4,$$

where C.S.F. is the so called "Crack Susceptibility Factor". These formulae illustrate very well the main effects, although the coefficients and element interactions, vary from one formula to another. Thus further investigations and a critical comparison are needed before these formulae may be used to predict crack susceptibility in a real situation. It must also be remembered that they have been derived from different tests and materials, and therefore any one formula may not be universally applicable. The significance of these relationships and their suitability as a basis for the research programme are discussed at greater length in section 3.

2.(4) Welding Process Effects

The effect on weld cracking of process variables appears to have received less attention than the chemical effects, although some interesting results have been reported. Following the early note by Bollenrath and Cornelius[4] that crack propensity depended on the welding process and therefore on the heat input, Stout[12] used the Lehigh Restraint Test to examine the influence of heat input on cracking

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during the manual metal arc welding of a 0.3% carbon steel and SAE 4130 (1% Cr-Mo) material. Although the experiments were limited, an increase in the heat input achieved by a reduction in welding speed lowered the cracking propensity in the case of SAE 4130 but little effect was noticed with the 0.3% carbon steel. The effect may have been due to the variation in bead size or welding speed itself, and the non-variation in the mild steel case may simply reflect a lack of severity of the test. Nevertheless, the observation shows an appreciation of the interaction between material and process effects. Terry and Tyler[9] report that a welding travel speed of 8 inches per minute (~ 4 m.m. per second) gives rise to the maximum severity of cracking in the welding of SAE 4130 steel sheet. This effect was observed using the patch weld test[37] and the Pellini test[43] and in production welding. Moreover, the carbon content affected the speed at which sound welds could be made. Borland and Rogerson[38] discuss the dependence of cracking behaviour on the welding speed found using the patch test in terms of heat flow rather than as a matter of variation in the inherent crack susceptibility of the material. Indeed the inherent susceptibility would be expected to be a property of the material itself, with variations in welding conditions acting through their effect on the metallurgical and mechanical conditions. Other comments on process variables include those by Apold[44] that cracking in manual metal arc welding is more serious with heavy gauge electrodes, and by Jones[35] that preheating and weld bead

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size exerted no noticeable effect in the Murex hot cracking test.

American work by Savage et.al[45] was directed towards the influence of welding conditions on solidification behaviour and cracking in HY80 steel. The shape of the weld pool was examined, and this changed from oval to a "teardrop" shape as the welding speed was increased or the current reduced. The centreline cracking tendency in bead on plate tests increased when the "teardrop" puddle was formed, and this effect is explained by the authors in terms of increased centreline segregation.

Thus the influence of welding procedure on cracking has been studied briefly under a variety of conditions. The expertise in this field is less well developed than in the case of composition variations, and the interaction between the two effects has received only scant attention.

2.(5) Metallurgical Factors and Parent Material Effects

The metallurgical features of weld metal are influenced almost entirely by the chemical composition of the base and filler metals and by the process characteristics. It is thus not surprising that there is little literature concerning metallurgy of weld metals as divorced from the other two aspects. Some effects have been observed to result from the metallurgical treatment of the parent material, for example special melting and deoxidising procedures have been found to reduce the crack susceptibility[28,46]. This effect was linked with a higher manganese content[6] and attributed by Eilender and Pribyl[28] to reduced segregation as compared with ordinary melts. The special steels were fine grained as a result of small nucleating particles present during solidification, and this gave a more homogeneous material. Experiments by Eilender and Pribyl[28] on the influence of heat treatment on cracking did not reveal any significant effect except when diffusion occurred and reduced the segregation. Bardenheuer and Bottenberg[31] observed an influence of transformation characteristics of the material but this was subject to some dispute[28]. Some of the early reports are obscure and it is not always clear whether reference is being made to weld metal cracking or H.A.Z. cracking.

More recently, microstructural effects have been discussed in Terry and Tyler's papers[9]: banded ferrite - pearlite materials were found to be particularly susceptible to transverse weld metal cracking, but when normalised, the susceptibility was reduced. Non-banded ferrite-carbide structures became more crack-sensitive when annealed to give a banded microstructure.

The solidification mode in the weld metal has also been found to influence the occurrence of cracking. Bray and Lozano[47] found that they could avoid solidification cracking in Cu–Si–Mn alloys by adjusting the welding variables to give a cellular rather than a dendritic structure. The improvement was attributed to the more favourable distribution of the alloy elements in the cellular structure.

2.(6) Theoretical Aspects

2.(6.1) General Theories

Early theoretical treatment of hot cracking was devoted to the problem of hot shortness in aluminium alloy castings. Investigations into the high temperature properties of Al-Si alloys[13] showed that a range of temperature existed above the solidus in which the material possessed a small finite strength but negligible ductility. This range of temperature increased as the silicon content of the alloy was raised up to 1.8% Si, but was observed thereafter to decrease. In a later paper[14] it was reported that hot shortness was most severe in the alloy containing 0.8% Si and was absent in super-purity aluminium and in aluminium containing 3% or more of silicon. These findings, together with a consideration of the Al-Si phase diagram, led to the conclusion that hot shortness was a function of solidification temperature range (Fig. 2.). The theory was proposed that cracking occurs between the temperature of coherence, i.e. the temperature at which, on cooling, the dendrites interlock and the material begins to behave as a solid rather than as a liquid, and the solidus[48]. This temperature range was termed the "brittle range" and where this range is small cracking does not constitute a problem. In this work, Pumphrey and Jennings state that the inherent cracking tendency is directly proportional to the extent of the brittle range, and to the contraction due to cooling through that range. This theory has been supported and extended by Medovar[49] to include other alloy systems and has become known as the "Shrinkage

Brittleness Theory". Similar in basis although rather different in detail is the Strain Theory of Pellini[50]. Pellini postulates a situation above the solidus in which massive grains of solid material are separated by thin films of still-liquid metal or compounds e.g. sulphides. The thermal shrinkage stresses bring about a parting of the grains at the liquid films as the weld metal solidifies. The "brittle range" of Pumphrey and Jennings is apparently regarded as dangerous by Pellini also.

Borland[51] attempts to rationalise these theories in the form of his "Generalised Theory". He considers the freezing process in four stages as outlined by Portevin and Dannemuller[52];

1. Primary dendrite formation,
2. Dendrite interlocking,
3. Grain boundary development (critical solidification range),
4. Solidification complete.

The temperature at which dendrite interlocking (stage 2.) begins is called the "coherence temperature" (Fig. 3.), and between this and the so-called "critical temperature" any fissures formed due to contraction will be healed by the remaining liquid. Below the critical temperature, in the critical solidification range (stage 3.), the system is dependent on "accommodation", i.e. deformation of the grains to accommodate contraction, to prevent fissuring. The cracks will form if the accommodation available is insufficient for the contraction which occurs, i.e. if the energy required to deform the grains exceeds that required to

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separate them.

The formation of cracks in the critical solidification range is influenced by the distribution of liquid in the grain structure. If liquid covers almost the entire grain faces high stresses are imposed on the thin necks of solid and crack formation occurs. If, on the other hand, the liquid covers only the grain edges and corners, then the solid-solid interfaces are extensive, resulting in lower stresses and less liklihood of cracking. Such factors are influenced by the interfacial energies relevant to the solid and liquid phases in question.

A theory postulating initiation of hot cracks in the solid state has been advanced by Movchan[53]. Called the "Polygonisation Theory", it is applied to alloys solidifying as a single phase face-centred cubic structure e.g. austenite, in which the dendrites form polygonisation boundaries with the aid of shrinkage stresses. Cracks initiate at the intersections of polygonisation boundaries and the regions which are rich in solute i.e. the grain boundaries. The cracks are microcracks, and may propagate under the influence of shrinkage stresses. This theory has found less favour than the solidification range theories, but may explain "ductility dip" cracking.

In all theories of solidification cracking, the fundamental mechanism is the yielding of low-strength materials under conditions of high stress. The importance of high temperature mechanical properties to the cracking problem was realised very early[4], and although hot tensile tests in

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this early work did not give a good indication of cracking properties, more success was claimed in later work[10]. This technique has been used for predicting hot crack susceptibility[54,55], although doubt has been expressed concerning the correlation of such tests with the crack susceptibility observed in practice[56].

2.(6.2) Impurity and Alloying Element Effects

All elements added to iron cause solidification to occur over a range of temperature[57], and therefore give rise to a fundamental crack susceptibility. In practice, however, complex interactions occur, and these may outweigh the fundamental effect of solidification range. It is the importance of these interactions which is discussed in this section.

In low alloy steels, the primary crack-forming elements are sulphur and phosphorus. There appears to be a better understanding of the effect of sulphur than that of phosphorus, the latter element not having been found to form grain boundary films in ferritic materials. It is known to segregate, however[10,21,23], and is thought to form an iron-iron phosphide eutectic. Phosphorus segregation can also be detected metallographically using solute-sensitive etching reagents.

The detrimental influence of sulphur is due primarily to grain boundary films of sulphides[19,23]. The iron-sulphur system is influenced by the deoxidation practice and the addition of alloy elements: in particular the ratio of manganese to sulphur is regarded as important[20,40,56], since manganese exerts a marked influence on the formation of

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sulphides. The iron-sulphur system solidifies in a eutectic manner, the eutectic point occurring at 31% sulphur[58,59]. In order to build up the eutectic composition, necessary for the formation of the sulphide, a binary Fe-S alloy containing 0.03% sulphur would have to be 99% solidified[60]. Since the sulphide cannot come out of solution until solidification is almost complete, it must be located in the solid metal at the grain boundaries and interdendritic spaces, where it may cause separation of the solid masses. The presence of manganese reduces the solubility of sulphur in liquid steel, as reported by Oelsen[61] and by Sherman/Chipman[61], manganese sulphide being formed progressively as the temperature falls. This compound forms while the metal is still liquid, and therefore remains in the metal matrix as harmless spherical inclusions. The phenomenon is explained by Wentrup[62] in terms of the termary Fe-Mn-S equilibrium diagram, Fig. 4. A miscibility gap traverses the ternary system causing the manganese-containing iron-sulphur melt to split into two liquid phases, one rich and one lean in sulphur, the sulphur-rich phase giving rise to the sulphide inclusions. The effect becomes more marked as the manganese content increases. Thus the manganese effectively reduces the sulphur content of the melt.

The inference concerning the effect of the Mn:S ratio in weld hot cracking is as follows:- Manganese removes sulphur from the liquid steel and forms harmless spherical inclusions. This also removes manganese from solution, hence the two elements are rendered inactive. Excess sulphur, if present, will be left to form iron sulphide grain boundary films, and if manganese is in excess, this will remain to act as an alloying element. Thus on the basis of this simplified explanation it is to be expected that manganese would confer cracking resistance by virtue of its affinity for sulphur, until this element is almost entirely removed. Since a given amount of sulphur is accounted for by a finite amount of manganese, benefits from further additions of this element would not be expected after the required level has been reached. This would account for the reported tailing off of the beneficial effect of increasing the Mn:S ratio as this parameter rises beyond about 50[37].

Consistent with the removal of sulphur by manganese and its effect on hot crack susceptibility is the influence of the rare earths. Such elements drastically reduce the solubility of sulphur in liquid steel[60], and have been found beneficial in the fight against hot cracking[10]. Moreover the improvement is observed to result from more desirable types of inclusions in the metal.

The role of oxygen in inclusion formation and the deoxidation process are highly complex in a real steel. With the present knowledge of the mechanisms at work, only a simplified picture of the role of each element can be drawn, to give at best a speculative prediction of the effect of variations in the levels of individual elements.

Deoxidation practice appears to affect the formation of inclusions in two ways, first by affecting the active sulphur

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content of the solidifying melt, and second, by affecting the form and distribution of the inclusions within the metal matrix. It is certainly not to be expected that these effects are entirely independent of one another. As with the iron-sulphur system, the iron-oxygen system forms a eutectic, but in the latter case much of the oxygen is separated from the metal as oxide, by a miscibility gap, before and during the freezing of the metal[58] (Fig. 5). Thus the oxide inclusions form mainly as randomly distributed globules. When oxygen and sulphur are both present the form of the inclusions, whether grain boundary films or globules, depends on the relative concentrations of oxygen and sulphur. If the oxygen to sulphur ratio is high, the oxides flux the sulphur into the globular inclusions. As the sulphur content, relative to oxygen, is increased there is an increasing tendency towards the formation of film-like sulphides. Thus both the active concentration and the distribution of the sulphur is altered by the sulphur to oxygen ratio.

The situation is further complicated in a real system since oxygen, sulphur, manganese, silicon and perhaps other deoxidants such as aluminium and titanium are present together. It is explained by Crafts and Hilty[58] that the use of manganese gives rise to globular oxide inclusions containing varying proportions of sulphides. This results from the existence of both oxide and sulphide miscibility gaps in the metal-oxide-sulphide system which separate oxygen and sulphur from the metal. With aluminium, however, the miscibility gaps do not substantially modify the Fe-FeS binary system,

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with the result that the sulphur concentration of the melt is not limited and the sulphide binary eutectic forms at the grain boundaries after the solidification of the metal is almost complete. Nevertheless there is evidence that aluminium does affect the morphology of sulphides[60,63]. Sims[60] shows that a sufficiently high aluminium concentration may give rise to a massive angular (type III) sulphide. Kiessling et.al[63] report that deoxidation with silicon gives globular (type I) sulphides, but aluminium gives either grain boundary chain (type II) or angular (type III) sulphides. Since aluminium was not detected in any of these sulphides it was concluded that aluminium acts only through its deoxidising power and does not take part in sulphide formation.

The morphology of sulphides has been studied recently by Yarwood et.al[64] who observed that in pure Fe-S-O alloys, a decrease of oxygen concentration brought about more sheet-like sulphide inclusions.

The complexity of the situation is well illustrated in the work of van Vlack et.al[65], who examined the surface energy aspects of inclusion morphology. Oxygen was observed to decrease the penetration of sulphides along the grain boundaries in a simple Fe-O-S alloy, but in a Fe-O-S-Mn-Si alloy, where silicates are present, the surface energy for sulphides and silicates favours increased grain boundary penetration when more oxygen is present. If aluminium is present, it reduces the manganese oxide content and with it the grain boundary penetration of liquid sulphide.

The effects of the alloying elements nickel, chromium, molybdenum and vanadium on solidification cracking are normally explained in terms of their austenitising or ferritising capacities in steel[62,66,67]. Both sulphur and phosphorus are more soluble in ferrite than in austenite, hence steels solidifying as δ-ferrite are more tolerant of these elements since they may be accommodated in solid solution and not rejected as harmful second phase particles. Thus, the austenite former nickel increases the cracking tendency[7,30], and molybdenum and vanadium are generally held to reduce it[7,8,40]. The effect of chromium is less certain and has been cited both as a crack promoter[10] and as a crack inhibitor[7]. This may come about as a result of the apparently opposing carbide forming and ferrite forming tendencies.

2.(6.3) Welding Process Effects

As the experimental treatment of the influence of welding process and procedure on hot cracking is limited, so there is a dearth of theoretical treatment to explain the effects.

Welding procedure is known to influence the solidification mode[45], weld pool length[68], solute segregation[67] and the stress-strain situation[12,20]. These factors, which may all affect cracking tendency, vary together when the welding procedure is changed, and it is not possible to separate them except by the use of special tests[69,70] which may go some way towards this end. Thus the theoretical explanations which do exist are speculative, and all too often tackle only one aspect of the problem.

The work of Savage et.al[45], which was briefly discussed earlier considers only the solidification effects in an explanation of the variation in solidification cracking with welding speed. The "tear drop" pool resulting from fast welding is said to cause excessive centre line segregation, which is responsible for the increased cracking tendency. Evidence of the segregation is not presented, however, and the strain implications of welding procedure are omitted completely.

Pumphrey and Jennings[48] explain how the crack susceptibility is affected by the length of the semi-solid zone behind the weld pool, which is related to the speed of welding. Terry and Tyler[9] use this together with the principles of heat flow[68] to attempt an explanation for their observations that an increase of welding speed may reduce cracking at levels above about 8 i.p.m. The essentials of the argument are as follows:-

The zone of the solidifying metal above the solidus, in which the dendrites are interlocked (i.e. the coherent region) is the crack susceptible zone, since liquid cannot readily flow through the dendrite network to fill the spaces caused by thermal contraction. When the welding speed is increased, the length of this zone is increased, and with it the crack susceptibility of the weld. The stress on the weld, which eventually causes cracking, occurs as a result of the thermal distortion, and Terry and Tyler discuss this in terms of the shape of the peak temperature isotherm which alters with welding parameters. This effect may counteract the increase in the weld pool length. No quantitative estimate of the situation is given, but this account appears to be the nearest approach to a comprehensive argument.

Of less general applicability are the analyses by Rogerson et.al[71] and by Savage and Lundin[72] of the Houldcroft[39] and Varestraint tests[69,70] respectively. Rogerson predicts the influence of welding speed and bead width on crack formation by considering the thermal strains obtaining in a Houldcroft test; Savage and Lundin report an analysis of stresses by considering the thermal expansions of the material around the weld pool. Changes in the stress pattern due to variation in welding conditions, however, do not appear to have been examined.

2.(7) Methods for Assessing Solidification Cracking Susceptibility

2.(7.1) General Considerations

Numerous methods have been developed for assessing the solidification crack susceptibility of weld metal, most of them having been designed to fulfil a specific purpose rather than for general applicability. The underlying principle of all solidification cracking tests is the same, however, namely the deposition under reproducible conditions of a weld or bead on plate in which a crack is induced by an artificially imposed strain. The extent to which cracking occurs in the test specimen, or the strain level at

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which a certain measure of cracking takes place is used as the basis for an assessment of the susceptibility to cracking.

Normally the tests depend for their usefulness on correlation of the results with production experience and may provide either a quantitative assessment of weldability, or a simple "crack-no crack" verdict according to requirements and the design of the test. Tests may be classified broadly by the method used to impose the strain on the solidifying weld metal. On the one hand, the strain may be applied mechanically by an external force, as in the Murex[36], KSLA[73] and Varestraint[69,70] tests, on the otherhand the thermal expansions in a cleverly-designed test piece provide the necessary strain as in the Houldcroft[39,74], Lehigh restraint[12] and Pellini[43] tests. The latter type of test has the advantage of simplicity in operation, normally requiring no more than simple testing apparatus. The manufacture of specimens also presents little difficulty although sometimes it may be time-consuming. Evaluation of the results is achieved either by a straightforward measurement of the crack length or by some other measurement on the specimen indicating in comparative terms the strain level at which cracking either began or finished. The properties of different materials can be readily compared by this means provided that operating conditions remain constant throughout. Tests of this type are limited in their versatility, however, many of them being applicable only to thin sheet material and usable

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with one welding process over a limited range of welding conditions. Moreover, the stress conditions are determined by the geometry of the situation and cannot be varied at will.

Some of these limitations are overcome by the methods employing mechanically induced strain. These techniques, although using more complicated apparatus, often give greater versatility than those previously discussed, in that they may be more readily applied to several welding processes and under different welding conditions. Moreover, the strain levels may be more readily controlled and calculated, enabling more quantitative information to be obtained. A review of some of the more important testing methods is given below.

2.(7.2) Tests Employing Mechanically Induced Strain

2.(7.2)1. The Murex Hot Cracking Test[36]

Designed originally for testing the weldability of manual metal arc electrodes, this technique consists essentially of depositing a fillet weld between two small plates as shown in Fig. 6. One of the plates is rotated such that the angle of the Vee groove is increased as welding progresses. The length of the crack formed in the weld, or the rotation speed necessary to give cracking gives a comparative assessment of the crack susceptibility. The strain imposed can be varied by altering the speed of rotation of the plate to suit the materials being tested. The Murex test is intended for routine examinations but has been widely used for research work[20,40,75]. The test is not readily adaptable for non-manual welding processes and is therefore limited in its applicability. Moreover, the results obtained are subject to operator variations, such as changing speed and manual technique. The complexity of the testing machine renders this method economic only where a large amount of testing is to be carried out, but under these conditions the simplicity of test specimen design and the modest material requirements are a distinct advantage.

2.(7.2)2. The K.S.L.A. Test[73]

This test is similar in principle to the Murex test, and involves the deposition of a weld run in the groove formed between two cylinders mounted side by side with their axes parallel, see Fig. 7. The cylinders are rotated in opposite senses so as to expand the weld and induce a crack. The strain level is altered by varying the speed of rotation. Although fairly recently reported, this test appears to be based on the principle of an old static technique of Kautz[18].

2.(7.2)3. Hemsworth's Tensile Method[76]

Hemsworth has described a test in which a bead on plate run is deposited on a flat specimen, stressed perpendicular to the welding direction to induce centreline cracks, see Fig. 8. Several tests are performed at different stress levels, and the level at which cracking is initiated indicates the crack susceptibility of the weld metal. The test may be used for manual or mechanised techniques.

2.(7.2)4. The Varestraint Test[69,70]

This technique is based on a rather different concept. A weld pass is made along the centre of a bar of dimensions 0.3 m. x 50 m.m. and of known thickness, mounted above a radius former, with the fixed end above top centre (see Fig. 9.). Welding begins at the free end, and when the arc reaches the point above top centre of the former, the bar is pulled rapidly round it by means of a pneumatic cylinder (see Fig. 10.). The resulting strain causes the weld to crack in the newly-solidified region. The strain induced by bending can be calculated and the crack susceptibility is assessed either by the extent to which the weld cracks or by the strain required to induce cracking.

The Varestraint test was operated originally using autogenous T.I.G. welding but has been shown to be applicable to the assessment of welds made by other processes, with or without filler metals. A small scale version of the test has also been developed[72], enabling thin materials to be tested, although it would appear possible to conduct tests on both thick (say 10 m.m.) and thin (less than 5 m.m.) materials using one piece of apparatus, if suitably designed. A technique similar to that of the Varestraint test has been reported by Brockhurst and Muir[77] in which a plate is positioned over a knife-edge fulchrum and a weld run is deposited on the plate surface immediately above and parallel to the fulchrum, (Fig. 11.). On completion of the weld the plate is bent around the knife edge, thus straining the weld in a manner conducive to centre line cracking. A modified form of test, the Transvarestraint test which combines the advantages of both methods has been developed[78,79]. Here the cracking is induced along the centreline by bending about a former with the axis parallel to the welding direction (Fig. 12.).

2.(7.3) Tests Employing Thermally Induced Strain

2.(7.3)1. The Houldcroft Fishbone Test[39]

This is a simple test developed for use with aluminium alloy sheet, although there is no reason why it should not be applied to other materials provided they are in sheet form. The principle is to initiate a crack by starting to weld at the edge of the sheet. As welding progresses, the propagation of the crack is hindered by a change in the shape of the specimen. The test piece is fabricated from a rectangular block 45 x 77 m.m. ($1\frac{3}{4}$ x 3 in), by cutting slots in either side, their depth increasing progressively along the length of the sheet, see Fig. 13. The specimen is placed on a carbon block and a T.I.G. melt run is made along the centreline starting at the end with the shortest slots. The function of the carbon block is to provide a relatively heat-insulating backing which is able to conduct electricity to the test piece. Moreover, the arc may be struck on the carbon so that a true edge start may be achieved on the specimen itself. A crack is initiated at the edge start since the cooling weld metal, which has no support at the trailing edge, has to contract against the expanding parent metal ahead of it. As welding proceeds the stress generated by this expansion is reduced since the increasing depth of the slots reduces the effective width of the plate. The crack ceases to grow when the stress falls below a certain

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level, which depends on the crack sensitivity of the material. The use of a slotted rather than a tapered specimen ensures more uniform thermal conditions along the specimen length. The crack sensitivity is expressed simply as the crack length observed.

A modified form of this test, for which improved reproducibility is claimed, has been described by Garland and Davies[74]. Wilkinson et.al[7] used the Houldcroft test in a miniature form to assess the weldability of high-strength steels and develop their crack susceptibility formula. They used strips of steel 38 m.m. (1½ inches) wide, slotted to form several miniature Houldcroft test specimens joined end to end by a narrow neck. The cracking index was obtained on the basis of the average amount of cracking observed in the several segments.

2.(7.3)2. The Huxley Test[80]

This test was developed specifically for comparing the solidification cracking susceptibility of sheet steels. The specimen is in the form of a strip, 0.4 m. x 38 m.m. x 2m.m. (16 x 1½ x 0.085 inches) slotted at intervals as shown in Fig. 14. to form eight segments each 38 m.m. long, and tabs at either end for start and finish purposes. The slots serve as thermal insulators between the segments and give an edge start effect to initiate a crack. The specimen is mounted in the jig shown (Fig. 14), which consists essentially of two parallel round bars, 12 m.m. in diameter, on which the test strip is loosely held by retaining screws, in line contact only to ensure minimum heat losses. A T.I.G.

/melt

melt run is made along the centre of the strip, starting at one end. As the arc enters a segment a crack is initiated by the edge start effect as in the Houldcroft test. As welding proceeds, and the end of the segment is approached, the amount of metal ahead of the arc, and therefore the stress-forming tendency due to its expansion, is gradually reduced; thus the crack is arrested. Fig. 15. shows the test at the completion of welding. The mean length of the cracks formed in the eight segments is expressed as a percentage of the segment length and used as the index of crack susceptibility, that is the Crack Susceptibility Factor, C.S.F.

The C.S.F. values determined using this test were found by Huxley to be closely related to production experience, and almost no cracking was experienced in the production welding of steels with a C.S.F. of less than 20. The test has been used extensively by Bristol Aerojet Limited for assessing the crack susceptibility of aircraft steels, and by other investigators[37,81]. Its use is limited to sheet materials, but it has the important advantage of modest equipment and material requirements.

2.(7.3)3. The Lehigh Restraint Test[12]

In this test (Fig. 16) the restraint necessary to cause weld metal cracking is measured. The degree of restraint is varied by changing the depth of the slots along the plate edges, and is expressed numerically by the width of the specimen between the inner ends of the slots. The threshold degree of restraint is expressed as the width of specimen which is just sufficient to produce cracking. The disadvantage of this test is that several experiments are required to obtain the threshold restraint level.

2.(7.3)4. The Circular Patch Test[37,38]

The circular patch test makes use of a severely restrained joint configuration which is often encountered in practice. A circular patch is welded into a matching hole in a square sheet as shown in Fig. 17. A sound weld indicates good weldability, and materials may be compared by measuring the extent of cracking and expressing the crack length as a percentage of the total weld length.

Many other forms of test have been devised, some of which are described in the reviews by Kammer et.al[82] and Williams et.al[19]. Most tests have been developed for a specific purpose, thus few, if any, are universally applicable. In choosing a test for solidification cracking susceptibility one must take into account the type and form of the material to be tested, the welding process and the variables to be studied. Also the relation of the results to practical situations, and to the results of tests performed in other laboratories, is a vitally important factor. The amount of testing envisaged will influence the choice between a test using a complex machine with a simple specimen, such as the Varestraint or Murex test, and a test in which the complication lies in the specimen itself, such as the

Houldcroft test. These factors were taken into account in the choice of tests for the present programme.

2.(8) Summary

The main composition factors influencing the occurrence of weld metal solidification cracking appear to be recognised and partially understood. Several empirical relationships between crack susceptibility and composition have been advanced, and these are in general agreement over the main effects. The formulae are often complicated, however, and have been derived under different, and sometimes uncertain, conditions; thus a critical assessment of composition - cracking relationships is essential if these are to be applicable both to welding practice and to the further investigation of cracking phenomena.

The effect of welding variables on cracking has not been rigorously investigated so that the information available is often obscure. Therefore it is inevitable that welding parameters have been selected on an empirical basis. This obvious and deplorable gap in the understanding of process effects must be filled if the optimum welding procedures are to be systematically developed.

Several theories have been proposed to explain the occurrence of solidification cracking in mechanical and metallurgical terms, and these appear adequate to explain the general effects. There seems, however, to be a total lack of quantitative theoretical treatment of the influence of welding parameters on cracking,* and there is virtually

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* See page 38

no information on the interaction of composition factors with welding parameter effects.

* Chihoski has recently published papers on this subject[83,84], but they appeared after the completion of the present research programme.

3. Investigational Work

The programme of investigation was designed to increase the knowledge of the influences of material composition and welding parameters on the phenomenon of weld solidification cracking. The literature survey has shown that a great deal of work has been done, but confusion still exists, particularly over the precise composition effects and the influence of welding parameters. The initial stimulus for the present work has come from the aerospace industry, for which the welding of thin sheet alloy steels presents a particular problem. Thus the emphasis has been placed on these materials, welded using the autogenous tungsten inert gas (T.I.G.) process. This combination of material and welding process gives added advantages for the study of fundamental principles, such as ease of handling and the absence of complicated dilution effects.

The first stage of the investigation covered the detailed analysis of existing composition/cracking data and the derivation of a simple rational relationship between crack susceptibility and material composition. Using this relationship, materials with a range of calculated crack susceptibility factors were chosen for an experimental investigation of the separate and combined effects of composition and welding parameters using the Huxley test. The most important welding parameter was found to be speed, the effects of which are complex, but which are explained in general terms by a model derived from experimental and theoretical considerations.

The investigation is concluded by a metallographic examination of the features influencing solidification cracking.

3.(1) Statistical Examination of Available Data and Formulae Relating Composition to Cracking

Crack susceptibility data had been obtained, for some 80 sheet aircraft steels, by Bristol Aerojet Limited using the Huxley test[42,85]. Using these data, existing published formulae were examined and a new relationship was obtained using the technique of multiple regression analysis. This work was carried out with the aid of Aston University's I.C.L. 1905 computer. The chemical composition range is given in Table 1.

3.(1.1) Examination of Published Formulae

The three most comprehensive formulae found in the literature were examined. These formulae are as follows:

1. The H.C.S. formula of Wilkinson, Cottrell and Huxley[7]:-

$$\text{H.C.S.} = \frac{C\left[S + P + \frac{Si}{25} + \frac{Ni}{100}\right] \times 10^3}{3Mn + Cr + Mo + V}$$

where H.C.S. is termed the Hot Crack Susceptibility. This much-quoted formula, which was partially discussed in the literature review was based on the results of miniature Houldcroft fishbone tests performed on several aircraft steels, and was obtained by manipulating the chemical analysis data to fit a smooth curve when plotted against the cracking test results.

2. Cottrell's Composition Expression[42]:-

$$\text{C.S.F.} \propto \left[P(C + 0.142Ni + 0.282Mn + 0.2Co - 0.14Mo - 0.224V) + \\ + 0.195S + 0.00216Cu\right] \times 10^4$$

where C.S.F. is the crack susceptibility factor as measured by the Huxley test. Although rather complicated, this expression gives a good straight line relationship between composition and the crack susceptibility factor, the correlation coefficient being 0.92 for the data as analysed by Cottrell.

3. Ostrovskaya's Carbon Equivalent Formulae[11]:-

Three equations are used as follows:-

a. for carbon between 0.09% and 0.14%:

$$C\ eq = C + 2S + \frac{P}{3} + \frac{Si-0.4}{10} + \frac{Mn-0.8}{12} + \frac{Ni}{12} + \frac{Cu}{15} + \frac{Cr-0.8}{15};$$

b. for carbon between 0.14% and 0.25%:

$$C\ eq = C + 2S + \frac{P}{3} + \frac{Si-0.4}{7} + \frac{Mn-0.8}{8} + \frac{Ni}{8} + \frac{Cu}{10} + \frac{Cr-0.8}{10};$$

c. for carbon contents greater than 0.25%:

$$C\ eq = C + 2.5S + \frac{P}{2.5} + \frac{Si-0.4}{5} + \frac{Mn-0.8}{6} + \frac{Ni}{6} + \frac{Cu}{8} + \frac{Cr-0.8}{10};$$

and the cracking propensity is said to be a function of the carbon equivalent so calculated. These formulae are based on crack susceptibility data for low alloy constructional steels, but the tests and processes used are not clear.

The three formulae (from Wilkinson, Cottrell and Ostrovskaya) were evaluated for each of the steels with the aid of the computer, and the calculated figures were plotted against the cracking propensity as measured by the Huxley test. The results are shown in figs 18 to 20. Correlation coefficients were calculated for the formulae of Wilkinson

et.al,

et.al, and of Cottrell, but the results due to Ostrovskaya's formula were observed by inspection to show very poor correlation with the observed effects. The calculated correlation coefficients (r) are as follows:

	r
H.C.S. Formula (Wilkinson et.al)	0.4
Cottrell's Expression	0.8

From this examination it is evident that the best of the three relationships considered is, for the present purpose, that statistically derived by Cottrell. The Russian formulae appear to offer no meaningful prediction of crack susceptibility in the case of the aircraft steels under consideration.

In general, three drawbacks are associated with these three formulae:

1. The correlation of the predictions with the Huxley test results is poor. (This applies to formulae 1 and 3).

2. The formulae are complicated and cumbersome to manipulate, and they contain complex terms representing speculative interactions, which prevent an assessment of the effect of each individual element.

3. No estimate is given of the statistical significance of each factor; thus the reliability of the formulae is uncertain.

It was considered essential, both for the present work and for the prediction of crack susceptibility in welding practice, to have a simple linear formula of known significance and good correlation. With such a formula the crack susceptibility could be predicted with known confidence and the relative effects of each chemical constituent could be assessed. In this way a more meaningful assessment of materials could be chosen for further research. It was to this end that the data available were examined by regression analysis.

3.(1.2) The Analysis of Crack Susceptibility Data

3.(1.2)1. Precision of the Huxley Test

In order to determine what is a reasonable precision for the statistical analysis, it was necessary to assess confidence limits for the Huxley test itself. The test consists in laying a T.I.G. melt run along the centre of a strip of steel, $1\frac{1}{2}$ inches wide, divided into eight segments each $1\frac{1}{2}$ inches long by saw cuts penetrating 0.6 inches in from each edge. A segment, two inches long, is included at each end for start and finish purposes. Thus each specimen yields eight cracks, the lengths of which can be measured. Whether each crack may be considered as a separate result, or the mean of eight cracks should be taken before further treatment, is a matter for consideration, since it may be argued that the eight segments of each specimen are not independent tests. Huxley and Shahini[86] have made an assessment on the basis of each crack considered as a separate result and have given the 95% confidence limits for the crack length as ± 0.069 (~ 0.07) inches if eight cracks, i.e. 1 specimen, are measured, and as ± 0.040 inches if 24 cracks, i.e. three specimens, are measured. Thus the

/confidence

confidence limits (95%) for C.S.F. are $\pm 0.07 \times \frac{200}{3} = \pm 5$ and $\pm 0.04 \times \frac{200}{3} = \pm 3$ respectively, since C.S.F. is (crack length) $\times \frac{200}{3}$ for a $1\frac{1}{2}$ inch segment length.

Huxley tests performed at Aston have been analysed in a similar way, on the bases both of each specimen (eight cracks) being taken as one single result, and of each crack considered independently.

The analyses are as follows:-

(i) Thirteen groups of three specimens were analysed, the mean of each group of eight cracks being taken as a single result. Each group had been performed under different welding conditions and/or on a different steel. The analysis technique employed was that outlined in Davies, "Statistical Methods in Research and Production"[87] and the figures are given in table 2.

Now there are two degrees of freedom for each group, and for the variance within the thirteen groups there are 39 - 13 i.e. 26 degrees of freedom.

Thus from column 6 (table 2),

$$\text{Variance } V = \frac{0.0722}{26} = 0.00278,$$

therefore Standard Deviation, SD = $(0.00278)^{\frac{1}{2}} = 0.053$.

For 95% confidence the crack length lies within $\pm 2 \times$ SD of the mean.

Therefore, the 95% confidence limits for the mean crack length of one specimen = $\pm 0.106''$;

but C.S.F. = crack length $\times \frac{200}{3}$,

therefore 95% confidence limits for C.S.F. $= \pm 0.106 \times \frac{200}{3}$

$$= \pm 7.1.$$

If three specimens are tested, the standard error of the mean of the three values is as follows:

$$\begin{matrix}\text{S.E. of mean} \\ \text{of 3 values}\end{matrix} = \frac{\text{S.D.}}{3} = \frac{\text{S.D.}}{1.732}\,.$$

Thus 95% confidence limits for mean crack length $= \pm\frac{0.106''}{1.732} =$ $= \pm 0.061''$, and 95% confidence limits for C.S.F. $= \pm\frac{7.1}{1.732} = \pm 4.1$.

(ii) The thirteen groups of three specimens were considered as thirteen groups of 24 individual measurements.

Mean variance = 0.0085 (Table 2. column 8)

Mean S.D. = 0.092".

Thus, if one specimen is tested, the situation is as follows: It is assumed that all strips within a group are alike, therefore Standard error of mean of 8 cracks $= \frac{0.092}{8} = 0.033''$, therefore 95% confidence limits for mean of 8 crack lengths $= \pm 2 \times 0.033 = \pm 0.066''$,

therefore 95% confidence limits for C.S.F.

$= \pm 0.066 \times \frac{200}{3} = \pm 4.4$,

and if three specimens are tested:-

Standard error of mean of 24 cracks $= \frac{0.092}{24} = 0.018''$,

therefore 95% confidence limits for mean of 24 crack lengths $= \pm 0.036''$,

therefore 95% confidence limits for C.S.F. $= \pm 2.4$.

The confidence data are summarised in table 3.

The figures illustrate the increased estimate of precision arising from the use of each section of the specimen as a separate result. It is considered that since the eight cracks in each specimen represent separate measurements

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but not completely independent experiments, the true precision lies somewhere between the values obtained using the two approaches. It is noteworthy that the excellent agreement between the results of the present work and those published by Huxley and Shahini illustrates the consistency in the precision of the test when performed in different laboratories.

It is not known how many specimens of each steel were used in the collection of the published data on hot crack susceptibility, thus it is uncertain how much variation can be accounted for by chemical variations. However, any variation in the C.S.F. greater than seven should certainly be accountable for, in terms of variations in the steel; and a value between this and the most optimistic value of three, i.e. five would seem reasonable for the 95% confidence limits of the test. The test used would not appear to be capable of discerning finer variations than this with 95% certainty, without the use of more than three specimens. Huxley discusses the precision obtained with more than three specimens and shows that little extra precision is obtained unless substantially more specimens are tested.

3.(1.2)2. The Regression Analysis

By means of the ICL 1905 Computer Statistics Programme Package the regression of Crack Susceptibility Factor (C.S.F.) on the composition was determined. Initially, the chemical elements were considered singly and the computer was programmed to perform three analyses of regression using first the elements significant at the 99% level, i.e. all

/elements,

elements, then those elements significant at the 5% level, and finally those significant at the 1% level. The analyses obtained are shown in tables 4, 5 and 6.

Following this, in order to determine whether extra precision could be achieved, the programme was altered to include certain of the elements taken in pairs, either as sums, products or quotients. Such analyses, including elements significant at the 40% and 1% levels are shown in tables 7 and 8.

3.(1.3) Analysis of Results

3.(1.3)1. The Merit of the Regression Analysis

The analysis of regression of C.S.F. on single chemical elements significant at the 5% level (table 5) yields the following equation:

$$\text{C.S.F.} = 36\text{C} + 12\text{Mn} + 8\text{Si} + 540\text{S} + 812\text{P} + 5\text{Ni} + 3.5\text{Co} - 20\text{V} - 13 \qquad \text{(Equation 1)}$$

where the elements are represented as weight percentages. A plot of observed C.S.F. against the values so calculated is given in Fig. 21.

The equation gives calculated values of C.S.F. which show a correlation of 0.8 with the observed values, and a residual error of 6.5. The residual error is the standard deviation of the distribution of the observations about the regression. For this analysis there were 73 degrees of freedom, thus for 95% confidence, the "t" parameter has the value 2. Hence the 95% confidence limits for the prediction of the C.S.F. value are ± 6.5 x 2, i.e. ± 13.

The inclusion of compound variables (tables 7 and 8) improved the precision of the analysis slightly, e.g. the analysis in table 8 gives a correlation of 0.9 and a residual error of 5.6 (hence 95% confidence limits of ± 11). This expression is more complicated than that incorporating only single elements (equation 1) and it is doubtful whether this extra complication is justified by the small improvement in precision. Most of the compound variables correlate significantly with one or more of the elements, thus this relationship would appear to offer little more than equation (1) in terms of precision of prediction of cracking behaviour or of explanation of the effects.

Considering equation (1), the 95% confidence limits are approximately double those of the test itself. This leaves an error equal to the testing error which must be accounted for by such factors as chemical analysis discrepancy, minor variations in sheet thickness from one material to another, and in testing conditions, and possibly other chemical factors not included in the analysis. From fig. 21 it is evident that much of the scatter results from a small number of anomolous steels, and the reason for this is not apparent from the data available.

The relationship derived appears to be more precise than those of Wilkinson and Ostrovskaya, and as precise as Cottrell's formula; thus, whilst not enabling a perfect prediction of weld cracking behaviour to be made, it comes nearer to that aim than has hitherto been possible. Moreover, the formula is simple, being independent of speculative interaction terms; and it contains only terms for the elements of proven significance.

Thus the requirement stated in 3.(1.1) has been met; the three drawbacks of the published formulae have been overcome, and a sound basis exists for further study.

Several groups of steels have been included in the analysis, thus some of the variance may be due to impurities exerting different effects in the different groups of steels. This possibility could be investigated if sufficient data were available on each class of material.

3.(1.3)2. The Significance of the Chemical Constituents and the Magnitudes of their Effects

The analysis performed in this work describes the effects of most of the alloying elements and impurities found in high-strength steels, and attaches significance levels to them. Considering first the analysis of regression on the elements whose effect is significant at the 1% level i.e. such that there is only a 1% probability that the effect could occur by chance (Table 6), it is seen that only carbon, sulphur, phosphorus, manganese and molybdenum are included, leaving silicon, nickel, chromium, vanadium, copper, cobalt and tungsten aside as insignificant. If the analysis is carried out at the 5% level, silicon, nickel, vanadium and cobalt are also included and molybdenum is removed. (The anomolous effect of replacing vanadium with molybdenum in the 1% analysis arises from the high correlation (0.8) between these two elements in the steels analysed, with the result that their effects cannot be separated statistically). The absence of copper from the formula derived is in contradiction to the formulae of Cottrell[42] and Ostrovskaya[11] but is in agreement with recent work of Huxley[88]. Similarly chromium is included in the formulae of Wilkinson[7] and Ostrovskaya[11], but its effect was regarded as doubtful by Jones[35]. Silicon was shown by Wilkinson to be one of the primary crack forming elements, but in the present analysis it has not appeared significant at the 1% level. Although there is dispute over the significance of some elements, sulphur, phosphorus and carbon remain the most significant crack promoters. A statistical illustration of this is given by the analysis of regression of C.S.F. on the contents of carbon, sulphur and phosphorus only (table 9), which gives a correlation of 0.6 and a residual error of 9. That these three elements alone account for such a large proportion of the variation bears witness to their overwhelming significance in comparison with the other factors.

The absence of $\frac{Mn}{S}$ from the list of significant compounds is noteworthy since this variable is often cited as a factor of great influence on solidification cracking susceptibility. However the Mn:S ratio has been found to be important at low values[37,40], but appears to be less significant at values above about 50[37]. In the present analysis, the Mn:S ratio ranged from 15 to 520, the mean value being about 100. Thus the present lack of significance does not contradict the expected effect.

The magnitudes of the effects of variations in the chemical composition are worthy of some consideration. Steel specifications vary in their tolerance limits, but the changes in C.S.F. arising from chemical variations of the order of the specification tolerances may be calculated as shown in table 10 together with the effects of 0.01% of sulphur and phosphorus. This table provides further evidence of the overriding influence of sulphur and phosphorus, even when the magnitude of likely variations in the contents of the various elements is taken into consideration. However, the additive effect of all the crack-forming elements changing from the lower to the upper end of the specification and vice versa would be large, equivalent to a sizeable variation in sulphur or phosphorus.

Thus, to summarise, the main crack promoters in high strength steels have been shown once again to be phosphorus, sulphur and carbon, but the other elements included in the regression analysis should be taken into account in an attempt to predict the sensitivity of a material to weld metal solidification cracking, without over-estimating the importance of minor variations.

The most important outcome of the analysis is a simple linear relationship, of known significance, between composition and cracking test results for sheet steels under fixed welding conditions. This forms a sound basis for the experimental study of composition and welding parameter effects on solidification cracking, and lays the foundation for a simpler and more reliable assessment of materials for practical welding applications.

3.(2) Experimental Study of the Effect of Composition on Cracking

3.(2.1) Materials

The choice of materials was influenced primarily by the following:

1. The need for a spread of composition to give a variation in crack susceptibility. Carbon, sulphur and phosphorus are particularly important in this respect, thus ranges in the levels of carbon from 0.1% to 0.35% and of sulphur and phosphorus from 0.01% to 0.04% were sought.
2. The requirement for steels of wide commercial and technological interest.
3. The specific interest of the aerospace industry in solidification cracking.

Steels to three specifications, En24, SAE 4130, and ASTM A387B, were therefore selected. Commercial casts were donated by the Redheugh Iron and Steel Co. Ltd., and the required variations in the sulphur and phosphorus levels were obtained by means of specially prepared casts, melted by Messrs. Willans Limited . Since particular interest in this work was expressed by the aircraft industry, three extra aircraft steels were included. These were supplied gratis by Bristol Aerojet Ltd., to their specifications RS 120, RS 130 and RS 140. Samples of all materials were obtained in the form of 2 m.m. sheet suitable for use with the Huxley test.

Chemical analysis figures, which were obtained by "Quantovac" spectrographic analysis and checked by conventional chemistry, are given in Table 11, together with the calculated C.S.F. values. It may be seen that a wide range of C.S.F. values (10 to 50) was included, hence the prime requirement of the selection of steels had been eminently fulfilled. Materials were generally within the specifications, although in steel SD, extra manganese appears to have been substituted for silicon, and in some cases decarburisation of the sheet affected the carbon content. Where this occurred (mainly in the SAE 4130 steels) the analysis samples were remelted so the quoted figures represent the average carbon content through the sheet thickness.

3.(2.2) Tests and Procedure

Crack susceptibility testing was carried out using the Huxley test, which was described in the literature survey, together with the T.I.G. welding process.

A D.C. power source was used, with the electrode negative. Conditions were set as nearly as possible to those used in previous work involving this test, i.e. 11 volts, 95 amps, with a welding speed of 2 m.m/second (5 i.p.m.). Initially three specimens were tested, giving 24 cracks, but in later work only one specimen was used since this left more material available for other experiments.

The cracks in the welded specimens were measured under low power magnification and the Crack Susceptibility Factor (C.S.F.) for each steel was calculated by expressing

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the mean crack length as a percentage of the segment length. Thus for a segment length of 1.5 inch (38.1 m.m.):

$$\text{C.S.F.} = \text{crack length (in inches)} \times \frac{200}{3}.$$

3.(2.3) Results

Results of the crack susceptibility tests are shown in table 12, alongside the values calculated using equation (1), section 3.(1.3)1. This enables the experimental values to be compared with the calculated ones, and hence the merits of the formula may be assessed.

It can be seen that the majority of experimental values fall within the 95% confidence limits of the regression equation, that is ± 13. Four steels gave results outside these limits, however, proving to be less crack susceptible than predicted. Steel SB did not crack at all although it gave a predicted C.S.F. value of 16, but the results giving rise to serious concern were those of steels SC, SD and AA4, where the discrepancy between the observed and calculated values was about 30. In the case of SD, it might be argued that the almost total absence of silicon could give rise to a freak effect, or that the coefficient for phosphorus would not be applicable at a level of 0.049%, which is well above that of the other steels. These composition anomolies are not present in steels SC and AA4, however, and so are unlikely to be wholly responsible for the unpredicted crack resistance. Steels SB and SD contain a higher percentage of aluminium, but once again AA4 is not exceptional in this respect. It is noteworthy that AA4 and AA5, both ASTM A387B steels show no difference in chemical analysis great enough to explain their wide difference in measured C.S.F. value.

It emerged from the literature survey, that deoxidation practice and oxygen content exerted an effect on the sulphur distribution. Thus an examination of the oxygen content of the steels under test appeared to form a logical approach to the solution of the problem.

3.(2.4) The Effect of Oxygen on Solidification Cracking

3.(2.4)1. Oxygen Analysis and a New Regression Equation

Samples of the steels were analysed for oxygen by vacuum fusion analysis using the Balzers Exhalograph apparatus. Nitrogen analyses were obtained at the same time (table 13). The figures for these two elements, together with the aluminium figures obtained from the Quantovac analysis (table 11), were included in the chemical composition data in the regression analysis. Samples of some 25 steels from used Huxley test pieces supplied by Bristol Aerojet Ltd., were also analysed, giving, in all, more than 40 results for regression. The following equation was obtained, including factors significant at the 5% level:

$$\text{C.S.F.} = 42\text{C} + 847\text{S} + 265\text{P} - 10\text{Mo} - 3042\,[\text{O}] + 19 \quad \text{- Equation 2}$$

correlation coefficient = 0.9, 95% confidence limits: ± 11

The C.S.F. figures calculated from this equation for the steels tested at Aston are shown in table 13 alongside the oxygen and nitrogen analyses. The samples from the Bristol Aerojet test pieces contained oxygen in the range 0.002 to 0.006 percent and nitrogen in the range 0.002 to 0.013 percent. Thus equation two would appear to be

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applicable to steels containing up to about 0.015 per cent of both oxygen and nitrogen.

One very important effect is immediately obvious from these results: oxygen acts very powerfully as a solidification crack inhibitor, its coefficient in the regression equation of -3042 being far in excess of those of sulphur and phosphorus in a negative sense. A comparison of steels AA4 and AA5 shows the direct effect of oxygen when added to a particular steel: thus AA4 in which the only great difference from AA5 is the higher oxygen level, exhibits a dramatically lower crack susceptibility.

The crack susceptibility predicted by equation two is, in every steel tested but one, correct within the 95% confidence limits of ± 11, (EB lies just outside these limits with an error of 12). Thus the anomolies arising from equation one have been overcome, even where the oxygen level is not particularly high. This results from the change in the relative crack promoting tendencies of sulphur and phosphorus predicted when oxygen is included in the formula, the smaller effect attributed to phosphorus giving rise to the superior prediction for the high phosphorus steel SD.

3.(2.4)2. Experimental Verification of the Effect of Oxygen

To support the foregoing evidence for the crack inhibiting effect of oxygen, experiments were conducted in which oxygen gas was added to the argon shield during testing. Steel AA5 was chosen for these experiments, to determine whether addition of oxygen would reduce its crack susceptibility to that of steel AA4.

Oxygen additions of 1%, 2% and 5% were made to the argon and the results, shown in table 14, illustrate that the C.S.F. value is progressively reduced as more oxygen is added to the shielding gas. With 2% oxygen, the C.S.F. of steel AA5 approximated to that of AA4 welded with pure argon.

An attempt was made to compare the oxygen contents of the weld metals produced using different shielding gas compositions; but the poor quality of the welds made under argon/oxygen shielding with the T.I.G. process gave too much scatter to enable a comparison to be drawn.

Nevertheless, these experiments support the observation of a powerful crack-inhibiting effect due to oxygen. Thus solidification cracking may be reduced in practice by the presence of oxygen in the steel, or by the addition of this element to the shielding gas in processes where this does not give poor arc characteristics, e.g. the M.I.G. process.

3.(3) The Effect of Process Variables

3.(3.1) Determination of the most Significant Variables using a Factorial Experiment

The effect of process variables has received less coverage than the effect of composition, with the result that it is not clear which variables exert the greatest influence on solidification cracking. These experiments were designed to highlight the major effects, and to indicate the areas needing further study.

The experiments were confined to the T.I.G. welding of sheet materials, and the crack susceptibility under the various conditions was measured by the Huxley test.

The variables which may be altered are numerous, and include not only welding speed, heat input, voltage, current, bead size and preheat, but also gas flow rate, electrode size and stickout, backing conditions, material heat treatment condition and many others.

For the present purpose three variables were selected, namely welding speed, bead size and material condition. Speed and bead size were chosen as the most likely variables to exert significant effects. Material condition was included to determine whether strength changes resulting from heat treatment affected the crack susceptibility because of changes in deformation characteristics. Any such influence could complicate the composition effects. Many of the welding variables are interdependent, thus, for example, bead size depends on the combination of voltage, speed and current. In these experiments, voltage was held constant by maintaining a constant arc length, and bead size at constant speed was adjusted via the current. Similarly, when welding speed was altered, the current was adjusted, where necessary to maintain a constant bead size. The three variables were investigated by means of a factorial experiment, the results of which were treated by analysis of variance.

3.(3.1)1. Design of Factorial Experiment and Selection of Testing Conditions

The technique of factorial design is fully described by Davies,[89], and was adopted in order to derive the greatest benefit from the experimental work. The investigation took the form of a $2^2 \times 3^2$ factorial experiment; that is; each of the three Bristol Aerojet Steels RS120, RS130 and RS140 were tested in three heat treatment conditions using two welding speeds and two different weld bead sizes. This required $3 \times 3 \times 2 \times 2 = 36$ tests in all.

The low and high levels of welding speed were set at 1.3 m.m/second and 3.8 m.m/second (3 i.p.m. and 9 i.p.m.) respectively since previous workers have observed considerable variation in cracking behaviour between welds made at these speeds. Weld bead sizes were the minimum and maximum which could be used with the Huxley test with full penetration and without fusion of the material around the slots. These levels were found by trial to be 5 m.m. and 6 m.m. respectively. For variation in heat treatment condition, samples were tested in the hardened, softened, and as received (quenched and tempered) states to give three strength levels.

The test results were examined by analysis of variance using the ICL 1905 computer statistics package.

3.(3.1)2. Results and Implications for Further Work

The Huxley test results are given in table 15. and the analysis of variance, as output by the computer, together with the significance tests is set out in tables 16 A and B.

These results show that the heat treatment condition of the sheet does not significantly affect the test results, but that both welding speed and bead size are highly significant. An increase of bead size, and a rise of welding speed from 1.3 m.m/second to 3.8 m.m/second both increased the degree of cracking.

On the basis of these results, and broader considerations, plans were made for further investigations. It was deemed essential that further investigations be conducted into the effect of welding speed, for not only has this been found highly significant, but it is capable of very wide variation in practice. Moreover, its effects have been found to be complicated[9], and so some very useful results, both technically and theoretically, could be obtained from a study of the effect of welding speed on cracking in steels of different compositions. The effect of weld bead size was highly significant, and its importance is not to be underestimated. The scope for varying bead size in the welding of thin sheet is limited, however, the values used in the factorial experiment being the maximum and minimum practicable. Moreover, the effect observed may have been exaggerated in the Huxley test since the approximation to an edge start becomes less true as the bead width is reduced. Thus it is doubtful whether much additional benefit would be forthcoming from further study of this variable. A study of bead size effects would be more appropriate to the welding of heavy sections, where the largest weld beads are used for

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economic reasons, to reduce the number of passes required.

The lack of significance of heat treatment condition enabled this factor to be discounted from the consideration of steel to steel variations and no further study of it in this programme was deemed necessary.

The main work on welding process effects was therefore concentrated on the influence of welding speed under conditions giving a constant weld bead width.

3.(3.2) The Effect of Welding Speed

3.(3.2)1. Crack Susceptibility Testing

All steels listed in table II were usedfor this part of the programme. Huxley tests were performed at speeds ranging from 1.3 m.m/second to 6.4 m.m/second (3 i.p.m. to 15 i.p.m.), the bead size being kept constant at 6 m.m. width. The welding current values required to fulfil this condition for the various speed levels are shown in table 17. Pure argon was used as the shielding gas, although with steel AA5 some additional tests were performed at selected speeds using argon + 5% oxygen. This was designed to reveal the effect of adding oxygen at different welding speeds.

3.(3.2)2. Results

The Huxley test results are shown graphically in figures 22 - 38. From these graphs it may be seen that the general effect is complex, thus an increase in welding speed first causes an increase in the crack susceptibility factor, and when increased further, it reduces it. The precise form of the C.S.F. - speed curve varies from one steel to another, as manifest by the varying position and height of the peak. Thus the steels of high crack susceptibility (as assessed at the reference speed of 2 m.m/second) show their maximum C.S.F. values at lower speeds than those of low crack susceptibility, and these maxima are higher. In extreme cases (e.g. steel EC) the peak occurs below 1 m.m/second and the relationship appears simply as one of decreasing C.S.F. with increasing speed. Conversely, in steels AA4 and AC, for example, the peak occurs close to the 6 m.m/second point, with the result that the rising portion of the curve is dominant. The relationships between the C.S.F. at 2 m.m/second and the position and height of the peak are shown in figures 39 and 40 respectively.

When oxygen was added to the shielding gas in the testing of steel AA5, the C.S.F. was reduced at all speeds, but the general form of the curve was preserved (figure 41). Thus the relationship for this steel was modified, and became similar to that for AA4, which was richer in oxygen than AA5 but otherwise similar.

3.(4) The Mechanism of the Speed Effect

3.(4.1) Introduction

It has been shown in the foregoing (section 3.(3.2)) that the effect of welding speed on weld centreline solidification cracking is complex. The experiments in this investigation have been confined to a single laboratory test, but the same general effect has been noted previously by workers using two different tests, as well observations on production welding, as a basis for their conclusions. Moreover, in the present work, the precise form of the cracking-speed relationship was found to vary from steel to steel although the same test was used throughout. In the light of these facts it can be stated with confidence that the relationships found are real ones, and not solely the result of welding test piece geometry. Thus the explanation of the effect which is about to be proposed is formulated on normal welds made in large plates, and not on a Huxley test piece. The welding conditions and sheet thickness selected are those used in the welding tests, but the explanation is intended to be broadly applicable to a fundamental effect, and not confined to the example observed in this investigation.

The significance of welding speed differs fundamentally from that of the composition factors, for welding speed affects not only the properties of the solidifying metal itself, but also the external forces acting upon it.

The speed of welding determines the thermal situation around the solidifying pool, and this in turn decides the thermal and metallurgical characteristics of the solidifying weld bead and the pattern of thermal strains in the surrounding material. These secondary effects are many and complex and may act either to promote or to inhibit the occurrence of cracking.

From the theories of solidification cracking, it is understood that the defect occurs as a result of stresses acting on a weak, so-called "crack susceptible" zone of the

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solidifying metal, which is present over a range of temperature characteristic of the particular alloy. It follows that cracking would tend to increase if either the length of the "crack susceptible" zone, or the stress acting upon it, or both, were increased.

The present observations are that the degree of cracking is first increased, then decreased, by a rise in the welding speed. The form of this relationship varies according to the composition of the steel, as explained in section 3.(3.2)2. and illustrated in Fig. 42. Curve 1. occurs with the steels richest in crack promoting elements e.g. EC, EA4, and EA5. The materials containing the lowest levels of crack promoting elements (or the highest levels of crack inhibitors) e.g. AC, SB and AA4, give rise to a form of relationship illustrated by curve 3. The intermediate materials e.g. RS120, AA5, EB, SD etc., are characterised by curves between these two forms, illustrated schematically by curve 2. These relationships cannot arise from a simple effect, but must result from at least two opposing forces interacting. The relative strengths of these forces change with the welding speed and alloy composition to give the curves illustrated.

The variation in crack susceptibility with speed is treated with reference to the effects of welding speed on the strains around the solidifying weld bead, and on the extent of the "crack susceptible" zone. This is in line with the existing theories, which are considered adequate to explain the basic defect. The effect of composition on the form of the cracking-speed relationship is explained by

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considering the relative importance of the interacting forces in the different alloys.

3.(4.2) Formulation of a Model to Explain the Cracking-Speed Relationship

3.(4.2)1. The Effect of Welding Speed on the Crack-Promoting Forces

The forces acting on the solidifying weld bead are considered in terms of the temperature distribution around the weld centreline. It is assumed that the material exhibits constant thermal properties throughout the temperature range, hence the expansion or contraction tendencies, which give rise to the compressive or tensile forces, are proportional to the rise or fall in temperature.

The thermal distributions of the zone within 30 m.m. either side of the weld centre line was obtained using a computer programme (WELDMODEL)[90] based on Rosenthal's two dimensional heat flow equation[68,91]. This equation is quoted in section 3.(4.2)2. In order to make the calculation possible, several simplifying assumptions have to be made as follows:

1. Heat losses through the plate surface are negligible.
2. The heat source is a point source.
3. The temperature at a point remote from the heat source remains unchanged.
4. The physical properties of the metal remain constant.
5. There is no change of state.

These assumptions may lead to inaccuracy, since radiation losses do occur and the physical properties of the material

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are not absolutely constant over a wide temperature range. Assumption 5. prohibits the use of the formula in the fused zone. The validity of the assumptions has been discussed by Tall[92], who used Rosenthal's equation to predict residual stresses. The theoretical treatment gave the pattern and magnitude of stresses of the right order, compared with measured values, indicating that the assumptions were justified. A theoretical analysis of the thermal situation around a weld is therefore made possible by this means.

Calculations were carried out for the welding of 2 m.m. thick steel sheet at speeds ranging from 1.3 m.m./second to 6.4 m.m./second, using current levels which had been found in practice to give a constant bead width (see Table 17). The change in shape of the temperature distribution with welding speed is shown in fig. 43. a.i and b.i, in which the 400°C isotherms for welds made at 1.3 m.m./second and 6.4 m.m./second are plotted. This simple illustration brings forth the first fundamental and very important observation that the thermal distribution during fast welding is drawn out to the rear of the heat source far more than during slow welding.

To investigate the temperature distribution further, graphs were plotted, for each welding speed, of temperature against distance from the centre line at several positions in front of and behind the heat source. These graphs, examples of which are shown in fig. 43. a.ii and b.ii, showed the lateral temperature distributions for the various stages in

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the thermal cycle of the work piece. The area under a curve, given by $\int_{0}^{30} T\,dy$, where T is the temperature and y the lateral distance from the centreline, represents the LINEAR HEAT CONTENT of a plane perpendicular to the welding direction. (In practice the lower limit of the integral was set at 4 m.m. and not zero, because the temperature of the fusion zone was not obtained from the programme. The upper limit of 30 m.m. was chosen as being the limit of temperature rise above about 200°C). The linear heat content so calculated is proportional to the expansion of a small element of material perpendicular to the welding direction.

The longitudinal distribution of linear heat content was obtained by plotting this against the longitudinal distance from the heat source. These graphs, for 1.3 m.m/second and 6.4 m.m/second welds are shown in fig. 43. a.iii and b.iii. The total heating of the parent metal, or AREA HEAT CONTENT is obtained from the area under these curves, i.e. $\int (\int T\,dy)dx$ where x is the longitudinal distance from the heat source.

It is from these graphs that the effect of welding speed on the crack-promoting forces may be observed. They enable the area heat content to be obtained not only for the whole heated area, but also for any small zone either ahead of or behind the heat source. The longitudinal distribution of linear heat content may be easily compared for the different welding speeds; in particular the peak of the graph shows where overall cooling, hence contraction, of the parent metal around the weld begins. Thus the zones of expansion and

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contraction relative to the position of the heat source can be observed.

Inspection of the graphs reveals the following features:

1. The area heat content of the zone extending from the cold metal ahead of the heat source to a position 40 m.m. behind it decreases as the welding speed is raised. Quantitive measurement showed that the area heat content was reduced from 4×10^5 m.m^2 °C to 2×10^5 m.m^2 °C when the speed was increased from 1.3 to 6.4 m.m/second.
2. The area heat content ahead of the heat source is greatly reduced by an increase in the welding speed. At 6.4 m.m/second the figure was 4,000 m.m^2 °C compared with 64,000 m.m^2 °C at 1.3 m.m/second.
3. The peak linear heat content occurs further behind the heat source as the speed of welding is raised. Thus the parent metal around the weld is still heating, and therefore expanding, at a distance of 40 m.m. behind the heat source in the 6.4 m.m/second case, whereas cooling, and therefore contraction, begin almost immediately behind the heat source in the case of welding at 1.3 m.m/second.

Observation 1. indicates that there is more expansion of the parent metal during slow welding than during fast welding. This suggests a greater stress forming potential in the slow welding case, but the absolute level of distortion is considered to be of less significance than its distribution relative to the heat source and solidifying, crack

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sensitive weld metal, expressed in observations 2. and 3. Areas which are expanding are in compression, and areas which are contracting are in tension. Thus the zones of the weld metal subject to external tension and compression may be expected to vary as the welding speed is altered. The situations obtaining in the cases of welds made at 1.3 m.m/second and 6.4 m.m/second are considered in turn.

(i) Welding at 1.3 m.m/second

In this case there is a large degree of heating in the parent metal ahead of the heat source, hence this material is in compression. In the fusion zone this compressive force is relieved, since the expansion is occurring into the molten pool. It is seen from observation 3. above that cooling of the parent metal begins almost immediately the heat source has passed. Thus a tensile force is set up which acts on the solidifying weld metal. This tensile force arises from two sources: first the contraction of the parent material which is immediately surrounding the relevant zone of weld metal, and which has lost the compressive force acting on it because the centreline zone has been melted; and second, the reaction to the compressive force set up in the expanding metal ahead of the heat source. Thus in this case, the solidifying metal, which is subject to its own solidification and shrinkage strains, is also acted upon by tensile forces from the surroundings which contribute to the crack-promoting forces.

(ii) Welding at 6.4 m.m/second

From observation 2. above, it is seen that there is far less heating in the parent metal ahead of the heat source during fast welding than during slow welding. Moreover, from observation 3., the parent metal is heating well behind the heat source, long after the weld metal is solidified. Thus, in contrast to the case of welding at 1.3 m.m/second the force acting on the solidifying metal is compressive, since although the lateral forces are reduced to zero at the fusion zone, expansion is occurring after this has passed, leading to the renewal of compression. The tensile reaction, which in the 1.3 m.m/second case acted on the solidifying metal, acts in the 6.4 m.m/second case on the already solid metal well to the rear. Furthermore, since the expansion is less in this case (observation 1.), the tensile reaction must also be reduced.

It should not be overlooked that the total compressive force in the fast welding case is less than in the slow welding case. However, as has just been shown, during slow welding this large compressive force is distributed to the detriment of the solidifying weld pool, whereas during fast welding the force, albeit smaller, acts to mitigate the shrinkage strains in the crack susceptible freezing weld metal.

It has thus been predicted theoretically that the crack-promoting forces acting on the solidifying weld pool decrease as the welding speed is increased. The approach adopted has been a simplified one: many complicating factors have been omitted in order to enable a comprehensible picture to be

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built up. Ideally this theoretical treatment of the distribution of distortions and forces should be supported by experimental observations, thus attempts were made to do this. Experimental observations of the distortions around a weld bead are, however, hampered by a number of difficulties, for it is essential that the observations be made while the weld is being laid down. It is relatively easy to scribe marks on a sheet and take measurements before and after welding, and such measurements may give very valuable information on the residual strains and final distortion after welding. It may even be possible to speculate on the dynamic distortion situation from the static measurements, but the true picture is not obtainable in this way. The difficulties of taking measurements during welding arise from the high temperature of the region around the heat source, and the smallness of the scale of the distortions to be measured. The use of strain gauges is normally prohibited by the high temperature, especially as a zone of material is fused. The other technique which may be tried is the use of photography, and experiments were tried in this investigation by this means. Grids were scribed on the surface of the sheets, along which melt runs were to be made under the required welding conditions. Such a scribed sheet was placed on a stand above the welding table with the scribed side downwards, so that the melt run could be made on the reverse side, and the grid photographed during welding. To avoid the danger in placing a camera under the test sheet, the underside (scribed side), suitably illuminated, was reflected by a mirror placed at 45° so that the camera could be set up to photograph the image horizontally. The sheet was photographed both before and during welding using a 5 inch x 4 inch sheet film camera. It was intended to examine the distortion patterns around the weld pool under varying welding conditions.

Useful information was not obtained from these experiments however, mainly, it is thought, for the following reasons:

(a) The distortion expected ahead of the heat source for sheet steel is very small. Assuming a coefficient of linear expansion of 15×10^{-6} m.m/m.m/^{o}C, an average rise of temperature of 1000°C over 2 m.m. would give a free expansion of 0.3 m.m. The expansion in constrained sheet could be very much less than this.

(b) The technique employed was not able to yield measurements with sufficient accuracy. Thus apart from small movements and buckling of the sheet during welding, the distortions by the optical equipment and lack of dimensional stability of the film would have given errors great enough to swamp the actual effect.

(c) The thermal glow and oxidation in the region of the weld pool reduced the clarity of the scribe marks.

It was not thought that the difficulties entailed in obtaining such measurements would be insuperable, but the time required was judged to be unjustifiable. Thus the experiments were abandoned, leaving only the theoretical pre-

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predictions to form this part of the model of the cracking-speed relationship.

3.(4.2)2. The Effect of Welding Speed on the Length of the Crack-Susceptible Zone

It was shown in section 3.(4.2)1. that as welding speed is increased, the thermal distribution is extended further to the rear. It follows from this that the length of the weld pool and of the solidifying zone behind it also increase as the welding speed is increased. The analysis given in that section does not cover the fusion zone, however, so the following analysis, also based on Rosenthal's heat flow equation, is presented to predict the change in length of an arbitrary zone, that between 1500°C and 1000°C, behind the molten weld pool with change in welding speed.

The Analysis

Consider two isothermals T_1 and T_2 in the region of the solidus along the weld centreline. It is assumed that the thermal conductivity, density, specific heat and thermal diffusivity of the material are constant with temperature.

Rosenthal's heat flow equation states:

$$T_1 - T_o = \frac{Q}{2\pi Kg} e^{-\lambda v \xi} Ko(\lambda VR),$$

where T_1 is the Temperature of a Point, T_o is the Initial Temperature,

Q is the Heat Input per Unit Time,

K is the Thermal Conductivity,

g is the Sheet Thickness,

$\frac{1}{2}\lambda$ is the Thermal Diffusivity,

V is the Welding Speed,
$-\xi$ is the Distance of the Point Behind the Heat Source,
R is the Polar Coordinate of the Point,
$Ko(x)$ is the Bessel Function (zero order second kind) of the quantity x.

The coordinates are illustrated in Fig. 44.

Using the approximation:

$$Ko(\lambda VR) = e^{-\lambda VR}\left(\frac{\pi}{2\ \lambda VR}\right)^{\frac{1}{2}}:$$

$$T_1 - T_o = \frac{Q}{2\pi Kg} \cdot e^{-\lambda v\xi}\, e^{-\lambda VR} \cdot \left[\frac{\pi}{2\lambda VR}\right]^{\frac{1}{2}} .$$

To the rear of the weld pool, for any isothermal at the centreline,

$$\xi = -R .$$

Therefore $$\xi = \frac{-\pi V}{2\lambda} \cdot \frac{Q}{V} \cdot \left[\frac{1}{2\pi Kg\ (T_1 - T_o)}\right]^2 .$$

Hence the distance, Y, between two isothermals T_1 and T_2 ($T_1 > T_2$) is given by:

$$Y = \xi_1 - \xi_2 = \frac{\pi V}{2\lambda} \left[\frac{Q}{V}\right]^2 \left[\frac{1}{2\pi Kg}\right]^2 \left[\frac{1}{(T_2 - T_o)^2} - \frac{1}{(T_1 - T_o)^2}\right]$$

$$= R_2 - R_1 .$$

If it is not assumed that $\frac{Q}{V}$ is constant with speed:

$$Y = \frac{A^1\ Q^2}{V}$$

where $$A^1 = \frac{\pi}{2\lambda} \frac{1}{(2\pi Kg)^2} \left[\frac{1}{(T_2 - T_0)^2} - \frac{1}{(T_1 - T_0)^2}\right]$$

In practice, the current and speed values (see Table 17) were related approximately in the following way:

$$I = 40 + 240\,V$$

where I is the Current (Amps)
V is the Speed (c.m/sec.)

Now $$Q = \frac{\eta\, E\, I}{4.2} \text{ cals per second}$$

where η = Arc Efficiency
E = Welding Voltage

Putting $\eta = 0.4$(93), $E = 10$ volts,
$Q = I$

Hence $$\begin{aligned} Q &= 40 + 240V \\ &= 40(1 + 6V) \end{aligned}$$

$$\text{c f} \quad Q = 8KTg\left(\frac{1}{5} + \frac{Vd}{4\alpha}\right)$$ 94,95

where d is the Fused Width, and
α is the Thermal Diffusivity

Hence $$Y = \frac{1600\,A^1\,(1 + 12V + 36V^2)}{V}$$

Putting $\lambda = 5$ $K = 0.1$ cal/cm/sec, $g = 0.2$ cm
$T_1 = 1500^\circ C$ $T_2 = 1000^\circ C$ $T_0 = 0^\circ C$

$$A^1 = 1.1 \times 10^{-5}$$

$$Y = \frac{0.02\,(1 + 12V + 36V^2)}{V}$$

A plot of Y against V is given in Fig. 45. It shows that Y increases with V in the speed range of interest, and that the relationship becomes effectively linear as the speed value becomes large.

Thus it is predicted theoretically that a zone bound by two isothermals, behind the molten pool, increases in length as the welding speed is increased. The crack susceptible zone is such a zone. Its boundary temperatures may be expected to vary with the material composition, so the precise relationship between welding speed and the extent of the crack susceptible zone may vary from one steel to another. However, the length of the crack susceptible zone depends directly on the temperature distribution for a given material, therefore the theoretical prediction that the crack susceptible zone length increases with welding speed is generally applicable to any welding situation.

The experimental investigation of this effect proved to be more satisfactory than that of the distortion of the parent material. In order to measure the length of the crack susceptible zone for varying welding speeds on samples of the steels used in the crack susceptibility experiments, use was made of the Varestraint test. This test, which was described in the literature review, involves the deposition of a fusion run along the centre of a bar which is held over a former. When the welding torch reaches the point over the top centre of the former, the bar is bent round the former by a pneumatic ram so as to strain the weld longitudinally (see Figs 9 and 10). Under these conditions the solidifying weld metal behind the molten pool is subjected to a very large strain and cracks are formed in the crack susceptible zone. The length of the zone containing cracks is measured under low power magnification.

Samples from special casts of the En 24, SAE 4130 and ASTM A387B steels used for the Huxley tests were obtained in 13 m.m. thickness for the Varestraint tests, aswell as commercial samples to these specifications. After initial trials to ascertain suitable welding and testing conditions, tests were performed at welding speeds ranging from 1 m.m/sec. to 10 m.m/sec. Helium shielding was used, since it is essential at the higher welding speeds to prevent "humping", i.e. uneven distribution of weld metal forming a humped and perhaps hollow weld bead.

The test results are shown in Figs 46 - 54, in which the crack susceptible zone length is plotted against welding speed. No graphs were obtained for steel AC and the commercial grade of ASTM A387B, since no cracking was observed in these materials. It may be seen from the graphs that, except in steels showing very low crack susceptibility, e.g. SB, AD, the length of the crack susceptible zone increased as the welding speed was raised. The length of the zone and the steepness of the relationship varied from one steel to another as expected, but the general pattern predicted by the theoretical analysis has been borne out by the experimental results.

3.(4.2)3. The Interaction of the Effects on the Crack-Promoting Forces and on the Crack-Susceptible Zone Length

The foregoing analyses have predicted two conflicting effects of welding speed on the potential solidification crack susceptibility. On the one hand an increase in welding speed reduces the tensile force on the solidifying weld bead due to the parent metal distortion, and on the otherhand, it increases the length of the zone of material susceptible to cracking.

It is beyond the scope of the present analysis to predict the precise relationship between welding speed and any of the individual factors contributing to cracking. Thus the simplified approach does not reveal whether a particular relationship is linear, or one in which the effect of speed is strong at one end of the scale and weaker at the other. The increase in length of the crack susceptible zone with increase of welding speed was seen theoretically to become steeper and approach linearity as the welding speed was raised, but there is no reason to suppose that the crack promoting tendency is directly proportional to the length of this zone. Similarly there is no knowledge of the exact relationship between the speed and the crack promoting forces beyond the general conclusion that an increase of speed reduces them.

In order to explain the observed general effect of welding speed on cracking, that is an increase followed by a decrease as the speed is increased, the detrimental influence must be dominant at low speeds while the beneficial influence is dominant at higher speeds. Such a situation is not difficult to visualise. Taking first the effect of the crack susceptible zone length, this is due largely to the increased difficulty of healing a crack with molten metal from the pool when the zone is long. There is almost certainly a critical length of this zone in which an increase is most severe: when the solidifying zone is very short an increase in length would be hardly detrimental since healing would present no problem. When the length is such that healing begins to become difficult, an increase would raise this difficulty severely. Once healing is so difficult as to be almost impossible, a further increase in solidification zone length could not do much more damage. The position of this critical zone would, moreover, affect the overall cracking-speed relationship.

Considering now the effect of speed on the crack promoting forces, it is postulated once again that there is a range of speed over which the effect is critical. It is recalled that during slow welding there is a large compressive force ahead of the crack susceptible zone due to the parent metal expansion, and a tensile force around and behind that zone. As the welding speed is increased, the compressive force moves to the rear. While the compressive force is still well ahead of the crack susceptible zone, this movement would not be expected to exert a great beneficial effect. Similarly once the compressive force is firmly acting on the crack susceptible zone, further movement to the rear due to increase of speed would not be expected to give a great deal more benefit. The maximum reduction in the cracking by this effect would be expected when the force on the crack susceptible zone is changing from tensile to compressive.

These two effects are summarised in fig. 55, in which each is plotted out together with the resultant, which is obtained by summing the individual effects. It is seen from this diagram that this resultant effect resembles very closely the general relationship between welding speed and crack susceptibility observed during crack susceptibility testing.

3.(4.3) <u>The Influence of Composition on the Cracking-Speed Relationship</u>

The experimental results were reported in section 3.(3.2)2., in which it was shown that steels of high crack susceptibility (as measured by the Huxley test at 2 m.m/sec.) gave maximum C.S.F. values at lower speeds than those of lower crack susceptibility. Moreover, these maxima were higher, see figs 39 and 40.

This effect of composition on the cracking-speed relationship is explained by a change in the relative effects of the parent metal distortion and solidification zone length factors. Since the changes in composition involved were very small and therefore would not cause significant changes in the thermal properties, the distortion effect would be expected to be constant throughout the composition range.

The crack susceptible zone length, however, would certainly not remain constant throughout the whole range. Most alloying elements extend the solidification temperature range, and if low melting point grain boundary segregates (e.g. sulphides) are formed, the crack susceptible zone length may be elongated substantially. Thus in the steels higher in the alloy and impurity elements, the critical solidification zone length, postulated in section 3.(4.2)3. above, occurs at lower welding speeds, thus pushing the peak crack susceptibility also to lower speed levels. Since the parent metal distortion effect is greater at these lower speeds, the peak crack susceptibility is higher. The effect of composition on the cracking-speed relationship is shown schematically in figure 56.

3.(5) Metallography

It was shown in section 3.(4) how the speed of welding affects the cracking behaviour of different steels through its influence on the distortion stresses and solidifying zone length. The solidifying zone is made crack susceptible, however, by the presence of low melting point segregates, whose form and distribution may vary from steel to steel and be affected by the welding conditions.

The object of the metallographic examination was to determine the influence of welding speed on the segregation of the impurities sulphur and phosphorus, and to ascertain the nature of the non-metallic inclusions in weld metals of the different steels examined. The techniques of optical microscopy, scanning electron microscopy and electron probe microanalysis were used.

3.(5.1) Segregation Studies: Phosphorus

The distribution of phosphorus was examined in welds made at 1.3 m.m/sec., 3.8 m.m/sec., and 6.4 m.m/sec. in steels SD and AD, both containing the higher level of phosphorus (see Table 11).

Sections were taken parallel to the sheet surface, polished and etched in 2% nital followed by an aqueous solution of picric acid containing a wetting agent. Picric acid darkens phosphorus-rich areas[2,96], thus the use of the two reagents in combination shows the distribution of phosphorus in relation to the grain structure. A typical area is shown in Fig. 57, in which the grain boundaries are delineated by dark-etching phosphorus-rich patches, showing that phosphorus has segregated to the grain boundaries. Fig. 58 shows the position of the centreline solidification cracks in relation to the microstructure, and it may be seen that the cracks follow the segregation pattern.

The distribution of phosphorus on a macro scale was fairly even throughout the whole weld zone, and there did not appear to be any concentration towards the centreline at

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any welding speed, in either of the two steels examined.

To follow up the microscopic examination, the electron probe microanalyser was used to obtain point analyses for phosphorus at positions in the weld from the fusion boundary to the centreline. No difference in the phosphorus level from one part of the weld to another was detected in any of the specimens examined.

Sulphur

Welds in steels AD and EC were selected for the examination of the sulphur distribution. Sulphur prints were taken from surfaces prepared parallel and perpendicular to the sheet surface, and point analyses were obtained using the electron probe microanalyser. No conclusive evidence was obtained by either of these methods for the segregation of sulphur to the centreline in welds made at 1.3, 3.8 or 6.4m.m/s.

3.(5.2) The Nature of the Inclusions

The inclusions at the weld centreline, that is, those in the region of the solidification cracks, were examined using the scanning electron microscope. Samples were taken from welds made at several speeds in all the special steels used, together with steels AA4 and AA5. Since the high temperature crack surfaces were oxidised, a satisfactory examination of them was not possible directly. Thus specimens were prepared by extending the solidification cracks obtained during crack susceptibility testing, the test pieces being fractured at room temperature. The fracture surface adjacent to the solidification crack was then examined.

If a high temperature crack has occurred along a plane of weakness containing second phase particles, these would be observable on the extended fracture face. An examination of this would therefore be expected to reveal the form of the inclusions associated with the solidification crack, and so contribute to an explanation of the variation in the crack susceptibility of different steels. The fracture mode would of course be different at room temperature from that near the solidus, hence the room temperature fracture characteristics themselves would bear little relation to the solidification cracking characteristics. Attention was therefore concentrated, in this study, on the form and distribution of the inclusions.

The observations were as follows:

ASTM A387B Steels

The outstanding feature of the welds in steel AD (high S, high P) was the fine, film-like appearance of the centre line fracture face (Fig. 59) which was observed at higher magnification (Fig. 60) to consist of very fine dimples around small globular inclusions. The fracture surface appeared faceted, showing that fracture had occurred at the grain boundaries, to which the inclusions had segregated. The same feature was observed in the AA4 and AA5 welds, although less inclusions appeared in the higher oxygen-containing AA4 welds than in those of AA5. The welds in steel AC (low S low P) contained markedly fewer inclusions, and the faceted grain boundary fracture appearance was absent.

En 24 Steels

These steels exhibited quite different features from the ASTM A387B materials. The fracture faces of welds in steel EC (high S low P) contained fewer and larger inclusions, and areas covered by what appeared to be a continuous film were seen. Also areas of sulphide eutectic, characterised by its palmate form, were observed (Fig. 61). The fracture faces of welds in EB appeared to be of the same form as those in EC, but no sulphide eutectic was apparent, and the fractures did not have the same film-like appearance (Fig. 62).

SAE 4130 Steels

The welds in the steels high in sulphur (SC and SE) contained a large number of inclusions and gave fractures similar to those in the ASTM A387B welds, (Fig. 63). The dimpled areas were less extensive, however, and the inclusions were often larger and lenticular in shape (Fig. 64). Welds in the low sulphur steels (SB and SD) contained very few inclusions by comparison with the high sulphur materials.

Within each steel sample, the welds exhibited similar features at all the welding speeds examined, although there appeared to be less inclusions in the welds made at higher speeds. It was not possible to confirm this observation, however, owing to the irregular distribution of the inclusions.

Following the examination using the scanning electron microscope, polished and etched surfaces parallel to the sheet surface were examined under the optical microscope, and the inclusions observed were analysed using the electron probe microanalyser.

The observations under the optical microscope supported the findings of the scanning electron microscope examination:

The ASTM A387B welds were found to contain chains of very fine inclusions embedded in areas of continuous ferrite (Fig. 65). These were similar inclusions to those observed in the scanning electron microscope, the optical micrograph being a two dimensional section through the surface shown in the scanning electron micrograph (Fig. 59). These chains of inclusions were far more numerous in the AD welds (high S) than in the AC welds (low S), and were present not only at the weld centreline, but throughout the whole weld bead width. This observation supports the lack of centreline segregation found in these weld metals.

The inclusions in the En24 weld metals (high S) were elongated, appearing as a cross section through the films observed in the scanning electron microscope (Fig. 66). The low sulphur En24 weld metals contained less inclusions than the high sulphur materials, consistent with the previous findings.

The inclusions in the weld metal of the SAE 4130 steel SE appeared in form similar to those in En24 weld metals, although they were more sparsely distributed (Fig. 67). In steel SC the inclusions were numerous, although did not appear to be at all elongated or film-like (Fig. 68).

Samples of the inclusions in welds of steels AD and EC were examined in the electron probe microanalyser. The very small size of the inclusions prevented an accurate analysis, but in both cases they were found to be rich in manganese and sulphur, indicating that they were manganese sulphide.

An attempt was made to analyse the inclusions in situ on the fracture faces using a scanning electron microscope with a microprobe attachment, but the time available did not permit such confirmatory information to be obtained.

4. Discussion

The investigational work just described has given a great deal of information on solidification cracking in high strength steels. A complete re-assessment of the cracking-composition relationship has led to a simplified crack susceptibility formula. This formula demonstrates more clearly than previous relationships the overriding importance of sulphur and phosphorus as crack promoters, and includes for the first time the powerful influence of oxygen as a crack inhibitor.

The remarkable and complex effect of welding speed on cracking has been shown to be heavily dependent on composition, making the present work the first systematic study of the interaction between chemical and process effects.

Cracking is attributed to the presence of grain-boundary films of non-metallic segregates. In the present investigations, the occurrence and form of the inclusions have been found to be related to the composition and crack susceptibility of the materials.

In this section these results are discussed, with particular reference to their importance, merit and limitations. The experiments have been conducted using two laboratory cracking tests, the Huxley test and the Varestraint test. These tests exhibit widely different characteristics which have been used to advantage in this work. The strong dependence of crack susceptibility determinations on the test method and procedure, however, highlights the importance of a thorough understanding of testing techniques. The application of cracking tests is therefore discussed at some length in this section. The discussion ends with a consideration of the practical implications of the results.

4.(1) The Effect of Composition on Solidification Cracking

4.(1.1) The Relationship between Crack Susceptibility and Composition

The purpose of crack susceptibility formulae was seen in the present work to be twofold. First, the formulae are needed for practical situations to aid material specification and assess weldability characteristics. Second, they are required to characterise materials to make possible a meaningful study of the interaction of composition and welding procedure effects. In order to perform these functions efficiently it is considered that the formulae must fulfil a number of conditions:

(a) The predicted behaviour of a material must correlate well with the observed behaviour,

(b) The level of significance of the elements included in the formulae must be known,

(c) The formulae must be simple enough to be used conveniently.

Examination of the most important published formulae revealed that none of these fulfilled the requirements. In only one case was the correlation between observed and calculated crack susceptibility satisfactory for the data under examination, and no information on the significance of the elements was available for any of the formulae. Moreover, each of the relationships was complicated, involving speculative interactions between elements. Thus the formulae were difficult to use since the effect of any one element could not readily be assessed. These shortcomings have been surmounted in the present work by means of a statistical treatment. The technique of regression analysis has been used, and it has been assumed that the effect of each element is simple and linear. Thus successive additions of a particular alloying element or impurity are assumed to have the same effect on the crack susceptibility, and to be independent of the other constituents present. This represents a radical departure from traditional techniques, in which product and quotient terms have been employed to account for the complex chemical interactions. A wealth of old data, supplemented by fresh results, has been completely reassessed by the regression technique, leading to equation (2) as follows:-

$$\text{C.S.F.} = 42\text{C} + 847\text{S} + 265\text{P} - 10\text{Mo} - 3042(\text{O}) + 19$$

This equation has the added advantage that all elements included are significant at the 5% level. The assumption that the effects are linear is justified by the high correlation coefficient (0.9) obtained. This is higher than that achieved using any of the previously published formulae.

The influence of composition on cracking is certainly not simple in practice, and it is not to be expected that

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the effect of each element is completely linear. Carbon, for instance, has been reported to exert varying effects according to its concentration, the effects changing at some critical level[32,35,66,97]. Huxley[97] suggested that carbon was beneficial in amounts up to 0.14% and detrimental at greater levels. The actual effects appear to be in some doubt, the level of 0.14% having been chosen arbitrarily as the point for separating the low-carbon and high-carbon steels. Jones[35] reported similar observations but Borland[66] attributed much of the variation to manganese and sulphur effects. The precise effects are therefore in dispute and indeed appear to be so complex that any attempt to account for the interactions in composition relationships cannot improve the reliability of the predictions. The simple linear relationship (equation 2) thus appears to represent the best available.

The value of regression analysis for highlighting new effects has been amply demonstrated. A number of the experimental steels gave C.S.F. values which were well outside the 95% confidence limits of regression equation 1. In particular steel AA4 was dramatically less crack susceptible than predicted although AA5, which appeared to be almost identical to it, showed no such discrepancy. It was known from the confidence limits that this effect was most unlikely to have occurred by chance, thus further investigations were initiated with confidence. The result of these was the discovery of the powerful effect of oxygen in inhibiting cracking. When oxygen was included in the composition data for the regression analysis the new formula (equation 2) was obtained.

This gave the higher coefficient of correlation (0.9 as opposed to 0.8) between the observed and calculated C.S.F. and gave more accurate predictions for the steels falsely assessed by equation 1. In order to test equation 2. rigorously, however, further data would be required, since the experimental data obtained in this programme were used in the analysis.

The inclusion of oxygen in the regression analysis has improved the precision, but another marked difference between equations 1. and 2. contributes to the improved assessment of steel SD. That is the reduced coefficient for phosphorus. In equation 2. the phosphorus coefficient is 265 compared with 812 in equation 1. It is thus apparent that phosphorus is less potent than indicated by equation 1. This discrepancy, which becomes more apparent when the level is high, as in SD, may have resulted from the omission of oxygen from the analysis, especially if there was a high negative correlation between the phosphorus and oxygen contents in the data. There was no means of checking this, since oxygen figures were not obtainable for the majority of the steels from which the original data were taken.

The precision obtainable with the regression analysis is generally sufficient to rate the materials correctly and indicate the main chemical effects. The predictions for steels SC and SE, however, are both correct within the 95% confidence limits but they have been rated in the reverse order to the observed effects. The existence of a further factor, still not included in the analysis, may therefore be suspected, although the precision of equation 2. is not sufficient to discriminate between the two steels. Additional data obtained from different steels are therefore required to improve the precision. In this way further factors influencing solidification cracking may be successively discovered, and an ever-increasing reliability of prediction may be achieved for production welding.

The merit of regression equations for expressing crack susceptibility has therefore been demonstrated. An equation of known precision was obtained using available data. Experimental results were then compared with the predictions of this formula, and the discrepancies used to discover new effects. A fresh relationship was then developed incorporating the new factors. This process may be repeated as more data become available, to reveal all significant factors, and produce the very best relationship which may accurately predict crack susceptibility in the practical welding situation.

4.(1.2) The Effect of Oxygen

The most spectacular example of crack inhibition by oxygen was observed by the comparison of the two heats of A.S.T.M. A387B steel, AA4 and AA5. The two steels were very similar in composition except for their oxygen contents (Tables 11 and 13) and both gave rise to high C.S.F. predictions using equation 1. In practice, AA4 showed very little cracking and significantly contained three times the oxygen content of AA5. This apparently very strong crack-inhibiting effect due to oxygen is illustrated by the high negative coefficient for that element in equation 2.

Although there appears to have been little attention paid to the role of oxygen in high temperature weld cracking, the general concensus of opinion supports the view that oxygen reduces the problem[19,32,67]. Much of the evidence, however, has been unconvincing. For example Podgaetskii[32] gives chemical analysis figures for a number of crack sensitive and crack insensitive steels, and links the crack resistant character of materials with an increased oxygen content. These steels were generally lower in the crack-promoting elements carbon, sulphur and phosphorus than the crack sensitive steels, making the evidence for oxygen inconclusive. In contrast, the evidence presented in the present work is very strong. Moreover it is supported by the observation that oxygen also reduces cracking when added to the shielding gas during welding. Thus the addition of 2% of oxygen to the argon reduced the crack susceptibility of the sensitive AA5 steel to that of the crack resistant AA4.

The mechanism of the oxygen effect has not been studied in detail in the present work. The result is, however, consistent with the known action of oxygen in fluxing sulphur out of the solidifying metal to form harmless spherical inclusions[58]. Solidification cracking is attributed to the presence of low melting point films, particularly sulphides, thus any element which reduces the tendency to film formation is a potential crack inhibitor. In steel AA5, the inclusions were not films, but very fine particles. They were, however, densely distributed along the grain boundaries in film-like array, thus they may have significantly weakened the solidifying metal. The crack resistant steel AA4 did not appear to contain these inclusions to the same extent. Thus it seems likely that they were responsible for cracking, and that oxygen reduced cracking by removing these. In order to prove the mechanism conclusively, further investigations need to be carried out using steels of carefully controlled composition and varying oxygen contents. In this way the effect of oxygen on solidification cracking in steels of different compositions could be clarified, and an examination of the inclusion distributions could reveal the method by which crack susceptibility is affected.

4.(2) The Effect of Welding Speed on Solidification Cracking

The speed of welding has been found to exert a very strong and complex effect on the solidification cracking tendency of low alloy steel welds. The general form of the relationship between speed and cracking is such that the maximum level of cracking occurs at a particular speed, above and below which the cracking tendency is reduced. The precise shape of the cracking-speed curve is severely dependent on the chemical composition of the material, the nature of this dependence being illustrated in Fig. 42. The materials rich in the crack promoting elements, particularly carbon, sulphur and phosphorus, give rise to a relationship illustrated by curve 1. The peak cracking level is high, and occurs at a low welding speed, perhaps even lower than can be reasonably employed, e.g. Steels EC, EA4 and EA5 (Figs 30-32). When the levels of the crack promoting

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elements are reduced, the peak in the curve drops and moves to higher speed levels (curves 2 and 3, Fig. 42). Practical examples of these two forms of relationship are given by steels AA5 and AC (Figs 28 and 25) respectively.

This is the first extensive and systematic study of the interaction of welding procedure and composition effects on cracking, previous works have usually considered only one of these factors. The general effect of welding speed was reported by Terry and Tyler, however. Their work was particularly useful, since they obtained a similar relationship using two different tests as well as in production welding. Thus any fears that the observation is merely a characteristic of the test are dispelled. Moreover, any effect which resulted from peculiarities of the test would appear the same in every steel, and would not show the dramatic variation with composition which was in fact observed.

The welding speed effect can be explained by a model in which two conflicting forces have been postulated, namely the effects of welding speed on the length of the crack susceptible zone and on the crack promoting forces. This is necessarily a simplified approach, since the thermal and metallurgical situation in and around a molten weld pool is so complex that each individual effect could not be identified, let alone analysed. Apart from the two effects considered, the system is influenced by the absolute cooling rate of the weld metal, the direction of grain growth in the solidifying metal, and segregation effects. These factors may all vary with the changes in welding speed and therefore,

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along with others, influence the cracking propensity of the weld. The fundamental cause of solidification cracking, however, is the action of the thermal stresses on a crack-sensitive zone in the weld metal. Thus, in order to make an analysis possible, the treatment was restricted to these two basic factors. The influence of welding speed on the crack promoting forces is particularly complex. It results from the differential thermal expansions around the weld which induce stresses which vary in magnitude and relative position according to the welding speed. In practice the stresses occur both parallel and perpendicular to the welding direction and cause bending as well as straightforward expansion and contraction tendencies. A comprehensive analysis would have to include the elastic and plastic deformations, which depend on the precise variation of the physical properties of the material with temperature. Once again the model was only made possible by simplifying the analysis to consider only the transverse stresses, assuming constant physical properties. The outcome of this analysis was that the thermal stresses around the weld centreline remain compressive further to the rear as the welding speed is raised, thereby reducing the crack promoting tendency (Fig. 43). This analysis has been supported by a similar treatment by Chihoski[83,84], which was published after the completion of the present work. Chihoski analysed the situation obtaining in the welding of aluminium sheet at speeds of 6 and 20 i.p.m. By treating the workpiece as a series of strips, parallel

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and perpendicular to the welding direction, he showed how the crack-promoting tensile forces occur further to the rear during welding at the higher speed. Chihoski presents some experimental evidence to support the analysis, but his papers did not include a treatment of the extent of the crack-susceptible zone. The variation in the length of this zone was demonstrated in the present work both theoretically and experimentally. It was shown conclusively that the length is increased by a rise in the welding speed (Figs 43 and 45). Thus the two main effects due to welding speed have been isolated and analysed. These exert opposing influences on solidification cracking, to give the characteristic overall relationship.

The influence of composition on the form of the relationship is explained in terms of its effects on the crack susceptible zone length. The compositions of the steels did not differ sufficiently to cause large variations in the thermal properties, hence the crack-promoting forces would not be expected to differ significantly from steel to steel. The crack-susceptible zone, on the other hand, is heavily dependent on the composition, as shown by the Varestraint test results (Figs 46-54). The concept of a critical crack susceptible zone length was discussed in section 3.(4.2)3. The critical length, and the welding speed at which it occurs remain speculative. However there can be little doubt that it occurs at lower speeds in the materials where the length is generally greater, since the maximum

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length at which healing can occur would be similar in every steel, if not identical. Indeed, in the highly crack susceptible materials, the crack susceptible zone length may be above the critical level even at the slowest welding speed; hence the lack of any peak in the overall cracking speed relationship, Thus although there is no direct experimental evidence to support the concept of a critical crack susceptible zone length, it is consistent with the observed effects.

4.(3) The Effect of Inclusions

Since solidification cracking is attributed to the presence of non-metallic films at the grain boundaries, the morphology of the inclusions in the experimental weld metals was clearly significant. Of particular importance in this respect were the En24 and ASTM A387B steels. In general, En24 was found to be highly crack sensitive, but ASTM A387B was far more readily weldable. These characteristics were reflected in the inclusion morphologies. In the En24 steel EC, areas of apparently continuous film were seen as well as the palmate sulphide eutectic (Fig. 61). In the ASTM A387B steels, on the other hand, fine spherical inclusions were predominant (Fig. 59). These characteristic types of inclusion were seen under the optical microscope (Figs 65 and 66) as well as in the scanning electron microscope. Thus these observations provide strong evidence to support the concept that cracking is enhanced by film formation.

The metallographic studies also reflected the variation in crack susceptibility between different heats to the same specification. Thus steel EB, of lower crack susceptibility, did not contain the films present in EC, and the least crack susceptible heats of ASTM A387B contained fewer of the fine inclusions. These results not only reinforce the film concept, but also show that spherical sulphide inclusions are not harmless, but simply the lesser of two evils. It is therefore desirable that sulphides be removed altogether, or absorbed into the deoxidation products. Oxygen is held to absorb sulphur by this means, and the mechanism is supported by the observation of less fine inclusions in the high oxygen AA4 weld metal than in AA5.

Very few inclusions were observed in the SAE 4130 steels low in sulphur, (SB and SD). In steels SC and SE the inclusions were more plentiful and mainly spherical, although in SE some elongated and lenticular inclusions were seen (Figs 64 and 67). These results are consistent with those of the cracking tests, although the reason for the difference between SC and SE is not entirely clear. Phosphorus would not appear to promote spherical inclusions, since it is noted as a crack promotor and not a crack inhibitor. SC did contain 0.017% of aluminium, and it is possible that this affected the sulphide morphology. Further work is needed to examine this point.

4.(4) The Application of Cracking Tests

In the present project all the results have been obtained using cracking tests. It must therefore be asked whether the results obtained are applicable to practical

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situations or just to the laboratory test set-up. Moreover, is it justifiable to conduct a project of this length on laboratory tests alone? Numerous cracking tests have been developed over a long period with the aims of achieving reproducible conditions for the assessment of weldability and obtaining an index of this property. It is extremely difficult, if not impossible, to achieve these aims without a special test. The applicability of test results to practical situations must be tested by long term experience, and it is this issue, which is linked with the justification and indeed the necessity for cracking tests, which is discussed in the first of the following two sections (4.(4.1)).

In this work, experience has been gained with two very different forms of cracking test, the Huxley test and the Varestraint test. They are totally different not only in the method of operation but also in the information given by the results. The difference is vividly demonstrated in the present work, and this highlights the importance of selecting a test very carefully to ensure that the information it provides is most relevant to the real situation. The differences between the two tests and their important implications, are discussed in section 4.(4.2).

4.(4.1) The Need for a Cracking Test

In practice solidification cracking is an intermittent defect. There is no sharp crack - no crack demarkation about which a weld metal cracks either obviously and continuously or not at all. On the contrary, if a large number of welds are produced some may be completely sound while others contain varying numbers of cracks. Over a period it is possible to notice a trend, however, so that certain materials and welding conditions tend to give rise to a more frequent occurrence of cracking than others. It is the strength of this tendency which is known as the crack susceptibility. Because of the intermittent nature of the defect, it is not normally possible to assess the crack susceptibility of a particular welding situation without making a large number of welds, since the first weld may not be typical of the batch.

The purpose of a cracking test is to exaggerate the factors which cause cracking so that an assessment of weldability can be made quickly and easily. This may be done in two ways:

1. By means of a crack- no crack test, in which freedom from cracking in the test indicates that the material is non-crack susceptible.
2. By means of a test giving a numerical assessment of crack susceptibility.

The first form of test is only satisfactory if the crack- no crack demarkation of the test coincides with a suitable acceptance level in practice. Otherwise satisfactory materials or welding procedures may be discarded or unsatisfactory ones accepted, causing unnecessary waste of resources. The second type of test has the advantage that the test results may be correlated with practical experience so that the gradation in weldability properties may be indicated by the numerical test results. Acceptance or rejection may then be based on the performance in the test in relation to the determined acceptance level.

The Huxley test is of the second kind, and has been used successfully to rate materials for aerospace components. It was designed for the assessment of materials under fixed welding conditions, and Huxley suggested that materials giving a C.S.F. value less than 20 should be readily weldable. In the present work the test has been used not only to assess materials under fixed welding conditions, but also to study the influence of welding speed on cracking. Some interesting relationships were obtained, the general form of which has been confirmed by earlier observations. Would the precise relationships between welding speed and cracking propensity be the same in production welding as in the test? Also would the acceptance level of 20 for the C.S.F. value be applicable at all welding speeds? Experiments were conducted to verify the observed relationships using normal weld joints. Initially straight butt welds were made, and later circular "patch type" welds were performed to give greater severity. Cracks were observed, but limitations of time and material prevented large numbers of welds being produced. As a result, no clear pattern emerged - isolated cracks were observed but no convincing observations were made, even when "good" and "bad" combinations of material and speed were compared. Attempts were made to increase the severity of the conditions, but such procedure departs from the practical situation and results merely in the development of another test!

These experiences support the necessity for suitable tests for measuring crack susceptibility. Before being accepted, the results of tests must be shown to represent the real situation, by comparison with practical experience over a long period. The results of the present investigations are in general agreement with production experience as observed by Terry and Tyler, and there is no reason to suppose that the more detailed effects are not representative of the practical situation. This needs to be verified by experiments in production, which are beyond the scope of a work of this nature. The results presented therefore form a sound basis of laboratory work on which longer term shop work can be based to achieve the optimum weldability properties of all types of material.

4.(4.2) The Importance of Selecting the Correct Test

Cracking tests with particular characteristics have been developed to suit specific requirements, and the characteristics of individual tests have been used to advantage in the present work. In the Huxley test, the stresses acting on the weld bead are thermally induced, just as they are in practice. They therefore change when the welding conditions change so that the test assesses not only the weld bead itself but the whole welding procedure. The Varestraint test, on the other hand, operates under a mechanically applied force, which may be externally controlled, quite independent of the welding procedure. This characteristic proved invaluable for isolating the effect of welding speed on the weld pool characteristics, but the test does not assess the welding situation as a whole. Apart from the verification of a theory, as in the present investigations, what are the applications of such a test? As with the Huxley test, the result depends on the welding conditions, but unlike the Huxley test, the Varestraint test overrides the important effect of thermal distortions. Situations can be visualised where this might be advantageous. In complicated structures, particularly in thick sections, stresses may be greater or more complex that those in the simple "Huxley test" situation; and these may not be reproducible in any simple test relying on thermal distortions. If the stresses and strains can be calculated, they may be applied to a weld in the correct magnitude by means of a test such as the Varestraint test. The controlled strains may be applied to welds made at any speed, and under any conditions. Thus almost any process may be used and any combination of parent material and filler metal may be tested. The Huxley test, on the other hand, is restricted in its application to autogenous T.I.G. welding.

The strain in the Varestraint test is applied at high speed. This is quite unlike any real welding situation and may detract from the advantages of varying the strain at will. The main potential of the Varestraint test for application to production welding would appear to be in the routine assessment of parent and filler metals under constant straining conditions. The relative crack susceptibility of materials may alter when the welding conditions are changed, and this may be detected and predicted by the test. It would not appear possible, however, to determine the change in cracking behaviour of a particular material as the welding conditions are altered, by means of the Varestraint test.

The information provided by the two tests is therefore quite different: the Huxley test assesses a welding situation, whereas the Varestraint test assesses a solidifying weld pool in isolation. A consideration of these two tests, and their respective applications in the present work, provide a very pertinent example of the importance of selecting the correct welding test for a particular application.

4.(5) Practical Implications of the Results

The wealth of results obtained in the investigation constitutes a rare accumulation of data embracing composition, welding procedure and test characteristics. The practical implications of the work are therefore far-reaching.

The importance of chemical composition in the problem of solidification cracking has long been recognised and this has led to many relationships of the type discussed in this work. Unnecessary and fruitless complication, however, has detracted from their usefulness. The formula developed in the present investigation is simple and this gives three outstanding practical advantages:

1. The relative importance of the individual chemical elements is immediately apparent. Thus, for example, the severity of sulphur and phosphorus in comparison with the other elements is placed clearly in perspective.

2. The simple statement of the chemical effects aids the specification of materials for welding applications, since the effect of variations may be easily predicted.
3. The formula provides a simple means of checking the weldability of batches of material on the basis of routine chemical analysis.

The powerful influence of oxygen on the occurrence of solidification cracking has important implications both for steel making and for welding practice. Cleaner steels with reduced oxide inclusion contents have been sought to meet increased requirements for mechanical properties. Thus special melting techniques such as vacuum melting have been used to give good toughness and fatigue properties, and to prevent lamellar tearing. It would appear, however, that from the solidification cracking standpoint, excessive deoxidation might be a bad thing. In view of the importance of oxygen in this respect, it is apparent that the specification and control of oxygen content is as important as that of, say, sulphur. In addition, the introduction of oxygen via the shielding gas has been shown to exert a marked influence on solidification cracking. Argon/oxygen shielding gas is not a practical proposition for T.I.G. welding of steels since it gives rise to an unstable arc and a poor quality weld. However, in other welding processes, e.g. M.I.G. welding or electroslag welding, an oxidising shielding gas or flux might well be used to advantage. Indeed for consumable electrode metal arc welding of steels, an oxidising atmosphere, e.g. CO_2 or argon + oxygen, is essential to give a stable arc and smooth metal transfer. Thus in situations where the composition of the base materials may not be easily or economically modified, the selection of appropriate welding process and consumables may represent a vital dimension in the elimination of solidification cracking.

The influence of welding speed is particularly important since this parameter gives a potentially rapid means of overcoming difficulties. The model used to explain the speed effects increases the understanding of the problem, and therefore, suitably modified or extended if necessary, it can enable the best advantage to be gained from welding procedure modifications. The model as formulated is directly applicable to the welding of thin sheet, where the thermal distribution is essentially two dimensional and the joint is completed in a single pass. There is no reason, in principle, why it should not also be applied to thicker sections if a single pass technique is used, or to the first pass of a multipass joint. For subsequent passes, where the relative movement of the plates is prevented by the root run, and where parent metal contraction effectively acts in three dimensions, the situation is substantially modified. In such a case the three dimensional heat flow would have to be analysed and the resulting stresses would be extremely complex. For example, in the two dimensional case the contraction of the parent metal around the weld bead causes tensile stresses to act on the solidifying metal. In the three dimensional case these may be partially balanced by the

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contraction of the solid metal underneath the weld bead. In theory the situation could be analysed in a similar way to the two dimensional case considered in this work, but the complexity of the stresses involved raises doubts about the feasibility of such an exercise in practice.

The model has been used to explain the observed effects, including that of composition on the form of the cracking-speed relationship. At present this is the limit of its capabilities, although it would be desirable to predict the cracking-speed relationship in terms of composition. Relationships between the C.S.F. at 2 m.m/second and the height and speed of the cracking peak have been plotted (Figs 39 and 40), and with more data these could be used for prediction in a similar way to the basic cracking-composition formula. In theory it should be possible to use the model to predict the form of the cracking-speed relationship in terms of the crack susceptible zone length. A further programme of work would be necessary to investigate the feasibility of this.

In welding practice, the empirical cracking-speed relationship and its variation with composition are of great importance in themselves. For the avoidance of solidification cracking, great emphasis is rightly placed on the material composition. If the specifications and welding procedure are correctly formulated and adhered to, welding is usually trouble-free. Minor variations in composition or the introduction of new materials may however lead to welding problems which appear to defy solution. A sound knowledge of the effects of welding processes on weldability, and an

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understanding of how these effects depend on the material itself, can save a great deal of time and frustration. It may even make the difference between a batch of material being usable or scrap, and does enable the best advantage to be gained from high strength materials which are notoriously difficult to weld.

It is not uncommon for cracking problems to occur without warning during welding. Such occurrences may cause serious delays, since welding is one of the many operations leading to the finished product, and must often be fitted into a tight production schedule. Problems of this kind often result from a change in the materials, and whilst these may take weeks or months to replace, the correct modification of the welding procedure may solve the problem without delay. Thus as well as being of considerable scientific interest, the findings described contribute to the knowledge and understanding of welding procedures and are thus of great economic significance.

The significance of cracking tests has been discussed above with reference to the two tests used in this project. The importance of understanding the methods used and the results obtained is universally applicable. The experience gained in this work may therefore be extended to other tests, and is particularly important if new testing procedures are being developed. This work has shown that the crack susceptibility indicated depends not only on the test itself, but also on the testing conditions. It is thus vitally important that test conditions reflect the situation being simulated, and that the results give a meaningful estimate of the true crack susceptibility. Unless care is taken in this respect, materials and procedures may be erroniously accepted or rejected, leading to severe economic consequences.

5. Conclusions

1. The occurrence of weld metal solidification cracking in low alloy steels is affected by both material composition and welding parameters, welding speed being particularly important.

2. Under conditions of fixed welding parameters, the cracking propensity may be related to the material composition by a linear regression equation. By this means the crack susceptibility of a material may be predicted, and unknown effects may be brought to light.

3. This relationship clearly demonstrates the relative importance of the chemical elements; in particular, it confirms the crack-promoting effects of sulphur and phosphorus.

4. Oxygen has been found to be a potent crack inhibiting agent, both when present in the material and when added to the shielding gas during welding.

5. The effect of welding speed on cracking in thin sheet low alloy steels is complex. The general effect of an increase in speed is to first increase, then decrease, the crack susceptibility.

6. The form of the cracking-speed relationship varies from steel to steel. High carbon and impurity contents tend to increase the maximum level of cracking, and reduce the speed at which it occurs.

7. The cracking-speed relationship is explained by a model based on the concept of two opposing effects due to an increase in the welding speed:

a. The length of the crack susceptible solidifying zone is increased, and

b. The tensile forces acting on this zone are reduced.

The balance of these effects determines the cracking behaviour of the weld.

8. Solidification crack susceptibility is related to the form of the grain boundary sulphide inclusions in the weld metal. Continuous film-like inclusions give rise to greater crack susceptibility than do fine globular particles.

9. Specially designed tests are essential for assessing susceptibility to solidification cracking. It is vital that the tests are well understood, so that crack susceptibility predictions may bear the best relationship to the practical situation.

6. Recommendations for Further Work

It is considered that the principle areas requiring further work are as follows:-

1. Long term production welding experiments are required to follow up the laboratory investigations. Furthermore, experiments are required on heavier-gauge materials to ascertain the extent to which the principles established may be applied, and what modifications are required in such circumstances.

2. Further tests on fresh materials are required to determine whether the cracking-composition formula requires modification to take account of further chemical effects.

3. The powerful effects of sulphur and phosphorus have been confirmed, but the relative potencies of these elements require further investigation.

4. Investigation of the effects of oxygen in steels of different compositions is required. The relation between oxygen, impurities and inclusion morphology in the context of cracking requires further study.

7. Acknowledgements

The author wishes to thank Dr. M. F. Jordan for his supervision and lively interest throughout this investigation. Thanks are also due to Professor W. O. Alexander, Head of the Department of Metallurgy for providing laboratory facilities. The author is indebted to the Ministry of Defence, to whom the work was carried out under contract; also to Bristol Aerojet Limited, especially Mr. H. V. Huxley, for supply of materials and information. The donation of steels by the Redheugh Iron and Steel Company Limited is gratefully acknowledged. Finally the author wishes to thank the academic and technical staff of the University for their help throughout the project.

8. Appendix - Publications

(a) Welding Institute Autumn Meeting, 1970.
"Weldability of Structural and Pressure Vessel Steels".
Contribution to discussion (Proceedings Pp 220 - 221).

Mr. E. J. Morgan-Warren and Dr. M. F. Jordan (University of Aston in Birmingham). In their paper, Dr. Baker and his co-workers have suggested some suitable methods for assessing weldability, and emphasis has been laid on the importance of being able to determine susceptibility to welding defects in terms of both material composition and fabrication procedure, e.g. welding process, material thickness etc. In the case of one important weld defect, that of solidification cracking, several formulae relating the crack susceptibility to the material composition have been advanced for the welding of low alloy sheet steel:

1. $$\text{H.C.S.} = \frac{C\left[S + P + \frac{Si}{25} + \frac{Ni}{100}\right]}{3Mn + Cr + Mo + V}$$

Wilkinson et.al[1].

2. (a) For C content between 0.09 and 0.14%:

$$CEq = C + 2S + \frac{P}{3} + \frac{Si - 0.4}{10} + \frac{Mn - 0.8}{12} + \frac{Ni}{12} + \frac{Cu}{15} + \frac{Cr - 0.8}{15};$$

where CEq is CE value.

(b) For C content between 0.14 and 0.25%:

$$CEq = C + 2S + \frac{P}{3} + \frac{Si - 0.4}{7} + \frac{Mn - 0.8}{8} + \frac{Ni}{8} + \frac{Cu}{10} + \frac{Cr - 0.8}{10};$$

(c) For materials containing more than 0.25%:

$$CEq = C + 2.5S + \frac{P}{2.5} + \frac{Si-0.4}{5} + \frac{Mn-0.8}{6} + \frac{Ni}{6} + \frac{Cu}{8} + \frac{Cr-0.8}{8}$$

Ostrovskaya[2].

3. $C.S.F. \propto \left[P(C + 0.142Ni + 0.282Mn + 0.2Co - 0.14Mo - 0.224V) + 0.195S + 0.00216Cu\right] \times 10^4$;

Cottrell[3].

However, these formulae are very complicated, incorporating complex and speculative chemical interactions, with the result that their usefulness for prediction and for assessment of the effects of individual elements has been limited.

In the course of recent research at Aston University, supported by the Department of Trade and Industry, formerly the Ministry of Technology, solidification cracking test data on more than eighty thin sheet aircraft steels, obtained using the Huxley test[4] have been examined by the technique of multiple regression. This analysis has produced a simple equation relating the measured parameter, the crack susceptibility factor (CSF) to the levels of the main chemical elements present, i.e. those whose effects were found to be significant at the 5% level:

$$C.S.F. = 36C + 12Mn + 8Si + 540S + 812P + 5Ni + 3.5Co - 20V - 13$$

The steels examined contained chemical elements in the following ranges (up to %):

C	0.4	Mn	1.5	Si	1.5
Ni	2	Cr	5	Mo	2
V	0.5	Co	6	S and P	0.03

When this relationship is used to calculate the C.S.F. the plot of observed value against calculated value is as shown in Fig.A, which includes the scatter band for 80% confidence.

A comparison of this formula with those quoted earlier shows it to be as accurate as the best of these, with the added advantage of simplicity. It facilitates a ready assessment of the performance of these materials in the Huxley test, which has been proved efficient for predicting the crack susceptibility of sheet aircraft steels during T.I.G. welding.

Comparative Effects

Element	Change in concentration, %	Change in CSF
Carbon	0.1	4
Manganese	0.2	2
Silicon	0.2	1
Sulphur	0.01	5
Phosphorus	0.01	8
Nickel	0.5	2
Vanadium	0.1	-2

TABLE A

Moreover, since the relationship is a simple linear sum, the effects of individual elements can be separated. Table A. shows the effect on the C.S.F. value of changes in individual elements which would be permitted by most specifications. Here the effects are put into perspective, taking into account the variations that are likely to occur between batches within a specification.

Thus, a simple formula has been obtained which enables hot crack susceptibility of a steel to be predicted in terms of performance in a proven test. The formula has been derived for one given situation, namely the T.I.G. welding of sheet aircraft steels using constant welding parameters. There would appear in principle to be no reason why similar relationships could not be derived for all other welding situations, e.g. submerged-arc welding of thick plate using any specified parameters. The necessary prerequisite is the availability of a versatile test such as the Varestraint or Transvarestraint test which has been thoroughly proven and correlated with representative welding situations.

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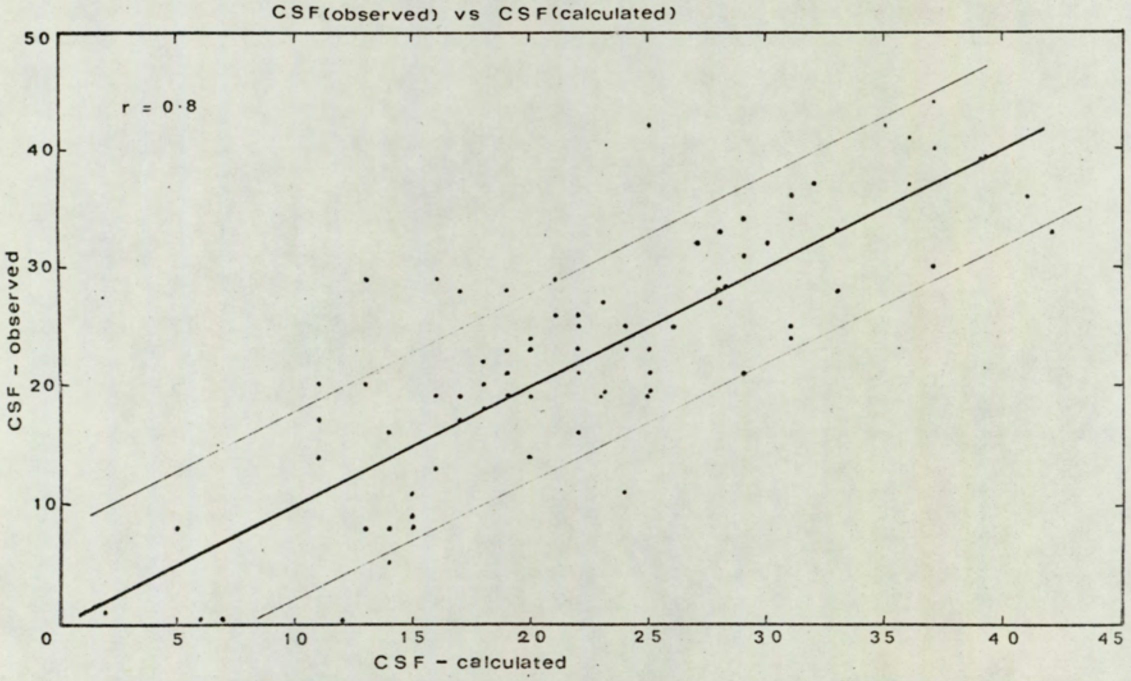
CSF(observed) vs CSF(calculated)
r = 0·8
CSF - observed
CSF - calculated
0
5
10
15
20
25
30
35
40
45
10
20
30
40
50

Fig. A.

(b) Technical note to be published in Journal of the Iron and Steel Institute:

"The Role of Oxygen in Weld Solidification Cracking in Low Alloy Steels".

by E. J. Morgan-Warren, B.Sc., and
M. F. Jordan, B.Sc., Ph.D., A.I.M., M. Weld. I.

In the welding of high strength steels, a major problem is the occurrence of solidification cracking, often termed "hot cracking", in the weld metal and a great deal of effort has been devoted to its solution, as Kammer, Masubuchi and Monroe show in their review.(1) It is generally agreed that cracking is due to the fracture of liquid films at grain boundaries during the last stages of solidification. The effect has been attributed to the impurities sulphur and phosphorus which tend to form low melting point segregates, the distribution of which is influenced by the alloying elements. However, recent work concerned with the effects of composition variables on weld solidification cracking in thin sheet low alloy steels has shown that the oxygen content of the steels is also a major factor.

Several investigators [2,3,4] have shown that quantitative expressions can be developed to relate crack susceptibility to chemical composition. In particular the present authors[5] demonstrated that the following simple linear formula adequately represented a large number of available crack susceptibility - chemical composition data:

$$\text{C.S.F.} = 36\text{C} + 12\text{Mn} + 8\text{Si} + 540\text{S} + 812\text{P} + 5\text{Ni} + 3.5\text{Co} - 20\text{V} - 13 \qquad \ldots\ldots(1)$$

In this expression C.S.F. is the Crack Susceptibility Factor determined in a weld cracking test developed by Huxley[6] and the concentrations of chemical constituents are expressed as weight percentages. This formula was obtained by regression analysis of C.S.F. on composition for tests on some 80 steels in thin sheet form, welded by the autogenous T.I.G. process; and it includes elements found to be significant at the 5% level.

In the present work this expression has been used to predict the crack susceptibility of a number of additional low alloy steel samples which were subsequently tested experimentally. While agreement between theoretical predictions and the test results was in general most acceptable, certain individual steels gave C.S.F. values well below those predicted by equation 1. The discrepancies were too great to be explained by errors in chemical analysis or by variation in testing conditions. Since oxygen plays an important role in the chemical reactions occurring during solidification and in particular on the distribution of sulphur[7,8], it was considered that the oxygen content of the steels should be investigated.

Vacuum fusion analysis of the steels exhibiting anomalously low crack susceptibility showed that their oxygen content was in general higher than the others. A total of 25 steels were then analysed for oxygen and a further regression

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analysis of the composition - cracking data was carried out, leading to the following revised expression:

$$\text{C.S.F.} = 42\text{C} + 847\text{S} + 265\text{P} - 10\text{Mo} - 3042\,(\text{O}) + 19 \quad . . . \; 2$$

This expression predicts a very powerful crack inhibiting effect of oxygen, which is quite consistent with the known tendency of oxygen to flux sulphur out of liquid steel to form harmless spherical inclusions[9].

The most spectacular example of crack inhibition by oxygen observed in these experiments emerges from a comparison of the behaviour of two heats of A.S.T.M. A387B steel, see Table B. Both heats are very similar in analysis except for their oxygen contents and give rise to very similar high C.S.F. predictions using equation 1. In practice Heat B showed very little cracking and significantly contained three times the oxygen content of Heat A. Use of equation 2 for crack susceptibility prediction eliminated this anomaly and gave good agreement with the experimental result.

In the light of these results it was apparent that it should be possible to reduce the level of cracking in a susceptible steel by increasing the oxygen content of the weld metal. Experiments were therefore conducted in which oxygen was added to the argon shielding gas in the T.I.G. welding of a crack susceptible steel (Heat A, A.S.T.M. A387B). Significantly there was a marked reduction in the cracking obtained in the Huxley cracking test. The results of these welding tests therefore confirmed those of the regression analysis on the composition - cracking data.

A great deal more work is required to elucidate the effect on cracking of oxygen, both individually and in combination with other elements. It is, however, apparent that oxygen is a powerful inhibitor of solidification cracking and, if present in a suitable concentration may minimise or eradicate the harmful effect of sulphur. Thus control of oxygen content would seem to be as important as the control of sulphur content in specifying the composition of weldable steels. Furthermore, since the weld metal oxygen content is a function of the welding process employed and particularly the oxidising potential of the shielding gas and/or slag, there would appear to be scope for control of cracking susceptibility by appropriate choice of welding process and consumables.

Acknowledgements

The authors thank Professor W. O. Alexander of the Department of Metallurgy, University of Aston, for provision of laboratory facilities. They also thank Bristol Aerojet Ltd., and the Redheugh Iron and Steel Co. Ltd., for the supply of steels. The investigation is being carried out under contract to the Ministry of Defence (Materials Division).

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TABLE B

Compositions and Crack Susceptibilities of ASTM A387B High Strength Steel

Steel Sample	Composition % wt										Crack Susceptibility		
	C	S	P	Si	Mn	Ni	Cr	Mo	Al	O	CSF.1	CSF.2	CSF. (obs)
Heat A	0.08	0.029	0.035	0.28	0.54	0.24	1.09	0.49	<0.005	0.005	44	36	37
Heat B	0.13	0.035	0.025	0.27	0.59	0.13	0.92	0.46	0.005	0.015	41	11	5

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TABLE 1. The Range of Chemical Composition Covered by the Statistical Analysis

Element	Range (up to %)
C	0.4
Mn	1.5
Si	1.5
Ni	2.0
Cr	5.0
Mo	2.0
V	0.5
Co	6.0
S	0.03
P	0.03
Cu	2.1
W	6.7

1	2	3	4	5	6 $(= (5) - (4))$	$7 = \frac{(6)}{2}$	8
Specimen Group	S Sum of Means	S^2	$S^2/_3$	$\sum x^2$	$\Sigma (x - \bar{x})^2$	Variance of Means	Variance of Individual Measurements
1	1.05	1.1025	0.3675	0.3677	0.0002	0.0001	0.0014
2	1.63	2.6569	0.8856	0.8875	0.0019	0.0010	0.0025
3	1.60	2.5600	0.8533	0.8542	0.0009	0.0004	0.0013
4	0.64	0.4096	0.1365	0.1504	0.0139	0.0069	0.0189
5	0.20	0.0400	0.0133	0.0176	0.0043	0.0022	0.0093
6	0.11	0.0121	0.0040	0.0045	0.0005	0.0002	0.0041
7	0.92	0.8464	0.2821	0.2888	0.0067	0.0034	0.0128
8	1.51	2.2801	0.7600	0.7605	0.0005	0.0002	0.0023
9	1.21	1.4641	0.4880	0.5059	0.0179	0.0090	0.0261
10	0.90	0.8100	0.2700	0.2846	0.0146	0.0073	0.0118
11	1.59	2.5281	0.8427	0.8435	0.0008	0.0004	0.0017
12	0.38	0.1444	0.0481	0.0570	0.0089	0.0044	0.0097
13	1.21	1.4641	0.4880	0.4891	0.0011	0.0006	0.0086
Total	12.95	-	5.4391	5.5113	0.0722	0.0361	0.1105

TABLE 2 Computation of Variance for the Huxley Test

Mean = 0.0085

TABLE 3 Confidence Data for the Huxley Test

Source of Result	Basis of Analysis	Number of Specimens	95% Confidence Limits	
			Crack Length	C.S.F.
Huxley and Shahini (Ref[86])	Single Cracks	1	0.069"	4.6
		3	0.040"	2.7
Present Work	Single Cracks	1	0.066"	4.4
		3	0.036"	2.4
Present Work	Complete Specimens (Mean of 8 Cracks)	1	0.106"	7.1
		3	0.061"	4.1

TABLE 4 Analysis of Regression of C.S.F. on Independent Variables Significant at the 99% Level

Independent Variable	Regression Coefficient	Standard Error	T Statistic	Partial Correlation
Carbon	39.31	9.41	4.18	0.46
Manganese	12.54	4.26	2.94	0.34
Silicon	3.64	2.74	1.33	0.16
Sulphur	545.40	105.10	5.19	0.54
Phosphorus	815.50	124.30	6.56	0.63
Nickel	4.39	1.52	2.89	0.34
Chromium	0.05	0.48	0.11	0.01
Molybdenum	- 0.47	2.31	0.21	- 0.03
Vanadium	-15.26	6.15	2.48	- 0.29
Copper	0.64	2.80	0.23	0.03
Cobalt	3.50	1.11	3.14	0.36
Tungsten	- 2.23	1.84	1.21	- 0.15

Intercept term: -15.42 Correlation: 0.83

Residual error: 6.63 Degrees of Freedom: 66

TABLE 5 Analysis of Regression of C.S.F. on Independent Variables Significant at the 5% Level

Independent Variable	Regression Coefficient	Standard Error	T Statistic	Partial Correlation
Carbon	35.55	7.37	4.82	0.49
Manganese	11.99	4.00	3.00	0.33
Silicon	4.80	1.35	3.57	0.39
Sulphur	540.56	96.65	5.59	0.55
Phosphorus	811.94	109.92	7.39	0.65
Nickel	4.91	1.30	3.77	0.40
Vanadium	-19.82	3.33	5.95	-0.57
Cobalt	3.52	1.04	3.38	0.37

Intercept term: - 13.47 Correlation: 0.82

Residual error: 6.50 Degrees of Freedom: 73

TABLE 6 Analysis of Regression of C.S.F. on Independent Variables Signfificant at the 1% Level

Independent Variable	Regression Coefficient	Standard Error	T Statistic	Partial Correlation
Carbon	28.18	8.29	3.40	0.36
Manganese	12.45	4.23	2.94	0.32
Sulphur	539.84	111.20	4.85	0.49
Phosphorus	705.22	125.53	5.62	0.54
Molybdenum	- 7.32	1.53	4.77	- 0.48

Intercept term: - 3.09 Correlation: 0.72

Residual error: 7.70 Degrees of Freedom: 76

TABLE 7 Analysis of Regression of C.S.F on Single and Compound Variables Significant at the 40% Level

Independent Variable	Regression Coefficient
Phosphorus	1323.16
Chromium	0.55
Cobalt	- 3.82
Tungsten	-12.23
C * Mn	99.34
Mn * S	-502.53
Mn * P	-681.54
Si * S	469.23
Si * Ni	- 19.31
Si * V	21.98
Ni * P	444.28
C / S	- 0.11
C / P	0.18
Ni/C	1.06
Mn/Si	- 0.37
Mn/P	- 0.09
Si/S	-00.04
V /Si	1.58
Co/S	0.01
Mo/S	0.12
Mo * S	1035.17
V + Mo	-28.97

Intercept term: - 0.75 Correlation 0.92

Residual error: 5.02 Degrees of freedom: 59

TABLE 8 Analysis of Regression of C.S.F. on Single and Compound Variables Significant at the 1% Level

Independent Variable	Regression Coefficient
Phosphorus	746.83
Tungsten	- 6.22
C * Mn	60.81
S * V	13.44
Ni * P	381.69
C / S	- 0.11
C / P	0.12
Mn / P	- 0.06
Mo / S	0.08
Mo * S	699.35
V + Mo	-19.72

Intercept: 6.95 Correlation: 0.88

Residual error: 5.59 Degrees of Freedom: 70

TABLE 9 Analysis of Regression of C.S.F. on the Percentage Contents of Carbon, Sulphur and Phosphorus Only

Independent Variable	Regression Coefficient
Carbon	13.95
Sulphur	511.37
Phosphorus	714.81

Intercept: 3.94 Correlation: 0.61

Residual error: 8.72 Degrees of Freedom: 78

TABLE 10 Comparative Effects of Alloy Elements and Impurities on the Crack Susceptibility Factor

Element	Change in % Concentration	Corresponding Variation in C.S.F.
Carbon	0.1	4
Manganese	0.2	2
Silicon	0.2	1
Sulphur	0.01	5
Phosphorus	0.01	8
Nickel	0.5	2
Vanadium	0.1	- 2

Relationship: $$\text{C.S.F.} = 36\,C + 12\,Mn + 8\,Si + 540\,S + 812\,P + 5\,Ni + 3.5\,Co - 20\,V - 13.$$

TABLE 11 Chemical Analysis and Predicted Crack Susceptibility Data for Experimental Steels

	STEEL	C	S	P	Si	Mn	Ni	Cr	Mo	Sn	Cu	Al	CSF*
		Chemical Analysis (Weight perCent)											
AIRCRAFT STEELS	RS120	0.35	0.012	0.023	0.30	0.53	0.10	0.93	0.24	0.018	0.12	0.005	33
	RS130	0.32	0.007	0.007	0.27	0.49	0.08	1.00	0.19	0.014	0.09	0.005	15
	RS140	0.46	0.011	0.018	0.27	0.68	0.17	3.30	1.02	0.018	0.10	0.005 V:0.25	29
A.S.T.M. A387B	AC^{+}	0.08	0.010	0.008	0.30	0.50	0.01	0.91	0.37	0.008	0,01	0.008	10
	AD^{+}	0.08	0.036	0.031	0.36	0.51	0.01	0.88	0.36	0.008	0.01	0.008	44
	AA4^{x}	0.13	0.035	0.025	0.27	0.59	0.13	0.92	0.46	0.014	0.09	0.005	41
	AA5^{x}	0.08	0.029	0.035	0.28	0.54	0.24	1.09	0.49	0.014	0.13	<0.005	44
En 24	EB^{+}	0.35	0.010	0.010	0.53	0.63	1.37	1.07	0.24	0.010	0.02	0.025	32
	EC^{+}	0.30	0.036	0.008	0.58	0.64	1.38	1.08	0.30	0.010	0.02	0.028	43
	EA4^{x}	0.36	0.041	0.014	0.32	0.70	1.54	1.23	0.24	0.025	0.20	0.010	51
	EA5^{x}	0.35	0.026	0.013	0.29	0.62	1.36	1.09	0.26	0.039	0.23	0.009	41
S.A.E. 4130	SB^{+}	0.18	0.010	0.008	0.34	0.70	0.02	0.99	0.21	0.009	0.02	0.060	16
	SC^{+}	0.18	0.034	0.037	0.37	0.69	0.01	0.99	0.22	0.008	0.02	0.017	53
	SD^{+}	0.19	0.010	0.049	0.06	0.85	0.01	0.92	0.27	0.005	0.01	0.063	50
	SE^{+}	0.21	0.033	0.010	0.34	0.70	0.01	1.02	0.23	0.006	0.02	<0.005	32
	SA4^{x}	0.22	0.026	0.028	0.24	0.67	0.16	0.93	0.20	0.012	0.13	0.005	42
	SA5^{x}	0.28	0.025	0.030	0.17	0.67	0.13	0.92	0.19	0.015	0.13	0.005	45

*C.S.F. $= 36C + 12Mn + 8Si + 540S + 812P + 5Ni + 3.5Co - 20V - 13$

$^{+}$ Special Casts

^{x} Commercial Casts

TABLE 12 Results of Crack Susceptibility Tests

	Steel	C.S.F. (calculated)*	C.S.F. (observed)
Aircraft Steels (Bristol Aerojet)	RS 120	33	35
	RS 130	15	14
	RS 140	29	23
A.S.T.M. A 387 B	AC	10	0
	AD	44	38
	AA4	41	5
	AA5	44	37
En 24	EB	32	36
	EC	43	55
	EA4	51	49
	EA5	41	53
S.A.E. 4130	SB	16	0
	SC	53	20
	SD	50	23
	SE	32	30
	SA4	42	37
	SA5	45	42

$$* \text{ C.S.F.} = 36C + 12Mn + 8Si + 540S + 812P + 5Ni + 3.5Co - 20V - 13$$

TABLE 13 Oxygen, Nitrogen and C.S.F. Data for Experimental Steels

Steel	O (%)	N (%)	C.S.F. (obs)	C.S.F. (calculated)
RS 120	0.005	0.009	35	32
RS 130	0.006	0.009	14	20
RS 140	0.006	0.008	23	24
AC	0.011	0.005	0	0
AD	0.008	0.006	38	33
AA4	0.015	0.004	5	11
AA5	0.005	0.005	37	36
EB	0.006	0.009	36	24
EC	0.004	0.006	55	49
EA4	0.004	0.007	49	58
EA5	0.003	0.005	53	47
SB	0.009	0.007	0	8
SC	0.011	0.006	20	30
SD	0.007	0.005	23	25
SE	0.011	0.005	30	23
SA4	0.004	0.007	37	43
SA5	0.005	0.006	42	43

$$\text{C.S.F.} = 42C + 847S + 265P - 10Mo - 3042(O) + 19$$

TABLE 14 C.S.F. Data for Steel AA5 using Argon + Oxygen Shielding Gas

Steel	Shielding Gas	Observed C.S.F.
AA5	Argon (pure)	37
AA5	Argon + 1% O_2	15
AA5	Argon + 2% O_2	4
AA5	Argon + 5% O_2	2
AA4 (comparison)	Argon (pure)	5

TABLE 15 Results of Huxley Tests performed under varying conditions of Welding Speed Bead size and Material Condition

Steel	Condition	Bead Size (Current)	C.S.F. at 1.3 m.m/s	C.S.F. at 3.8 m.m/s
RS 120	A.R.	Small	13	26
		Large	34	35
	S.A.	Small	16	15
		Large	25	34
	O.Q.	Small	13	19
		Large	22	37
RS 130	A.R.	Small	0	2
		Large	13	26
	S.A.	Small	0	0
		Large	7	29
	O.Q.	Small	0	0
		Large	5	26
RS 140	A.R.	Small	3	14
		Large	16	33
	S.A.	Small	3	11
		Large	23	35
	O.Q.	Small	1	13
		Large	23	33

```
                    17/10/22    06/10/70      ICL  1900
                    ANALYSIS OF VARIANCE          MAT1

ANALYSIS OF VARIANCE          MAT1  LP  F TEST

MODEL Y=A(I)+B(J)+C(K)+D(L)+AB(IJ)+AC(IK)+AD(IL)+BC(JK)+
      BD(JL)+CD(KL)+ABC(IJK)+ABD(IJL)+ACD(IKL)+BCD(JKL)+
      ABCD(IJKL)+E

NUMBER OF LEVELS OF FACTOR SUBSCRIPTS  ( 4 )
                  A-I  ...   3
                  B-J  ...   3
                  C-K  ...   2
                  D-L  ...   2

            I IS STEEL  J IS CONDITION K IS SPEED  L IS CURRENT

                  MODEL TERM     SUM  OF    DEGREES     VARIANCE
                  **********     SQUARES   OF FREEDOM   ********

                      A(I).307017E  0         2        0.15351

                      B(J).571667E- 2         2        0.00286

                      C(K).182044E  0         1        0.18204

                      D(L).582678E  0         1        0.58268

                     AB(IJ).101667E- 1        4        0.00254

                     AC(IK).667222E- 2        2        0.00334

                     AD(IL).973889E- 2        2        0.00487

                     BC(JK).183889E- 2        2        0.00092

                     BD(JL).872222E- 3        2        0.00044

                     CD(KL).300444E- 1        1        0.03004

                   ABC(IJK).574444E- 2        4        0.00144

                   ABD(IJL).601111E- 2        4        0.00150

                   ACD(IKL).257056E- 1        2        0.01285

                   BCD(JKL).990556E- 2        2        0.00495

                 ABCD(IJKL).133444E- 1        4        0.00334

                   RESIDUAL.000000E  0        0

                     TOTALS.119750E  1       35        0.03421
```

Analysis of Variance for the Results shown in Table 15

as output by the Computer

TABLE 16a

```
                  17/10/25     06/10/70      ICL  1900
                  ANALYSIS OF VARIANCE          MAT1

                 -------- RESIDUAL -------                 SIGNIFI-
MODEL TERM       SUM  OF     D.OF                VARIANCE  CANCE AT
**********       SQUARES      F.    VARIANCE        RATIO  % LEVELS
                                                             5   1

           A(I).1200E  0     29      0.004       37.084    YES YES

           B(J).1200E  0     29      0.004        0.691

           C(K).1200E  0     29      0.004       43.978    YES YES

           D(L).1200E  0     29      0.004      140.762    YES YES

         AB(IJ).6071E- 1     16      0.004        0.670

         AC(IK).6071E- 1     16      0.004        0.879

         AD(IL).6071E- 1     16      0.004        1.283

         BC(JK).6071E- 1     16      0.004        0.242

         BD(JL).6071E- 1     16      0.004        0.115

         CD(KL).6071E- 1     16      0.004      100.000    YES

       ABC(IJK).1334E- 1      4      0.003        0.430

       ABD(IJL).1334E- 1      4      0.003        0.450

       ACD(IKL).1334E- 1      4      0.003        3.853

       BCD(JKL).1334E- 1      4      0.003        1.485

    ABCD(IJKL).0000E  0      0
```

Significance Tests for the Results shown in Table 15

as output by the Computer

TABLE 16b

TABLE 17 Welding Speeds and Corresponding Current Levels Required to Give a 6 m.m. Bead Width

Speed		Current (Amp)
I.P.M.	m.m/sec.	
3	1.3	65
5	2.0	90
7	3.0	110
9	3.8	135
11	4.7	155
13	5.5	175
15	6.4	190

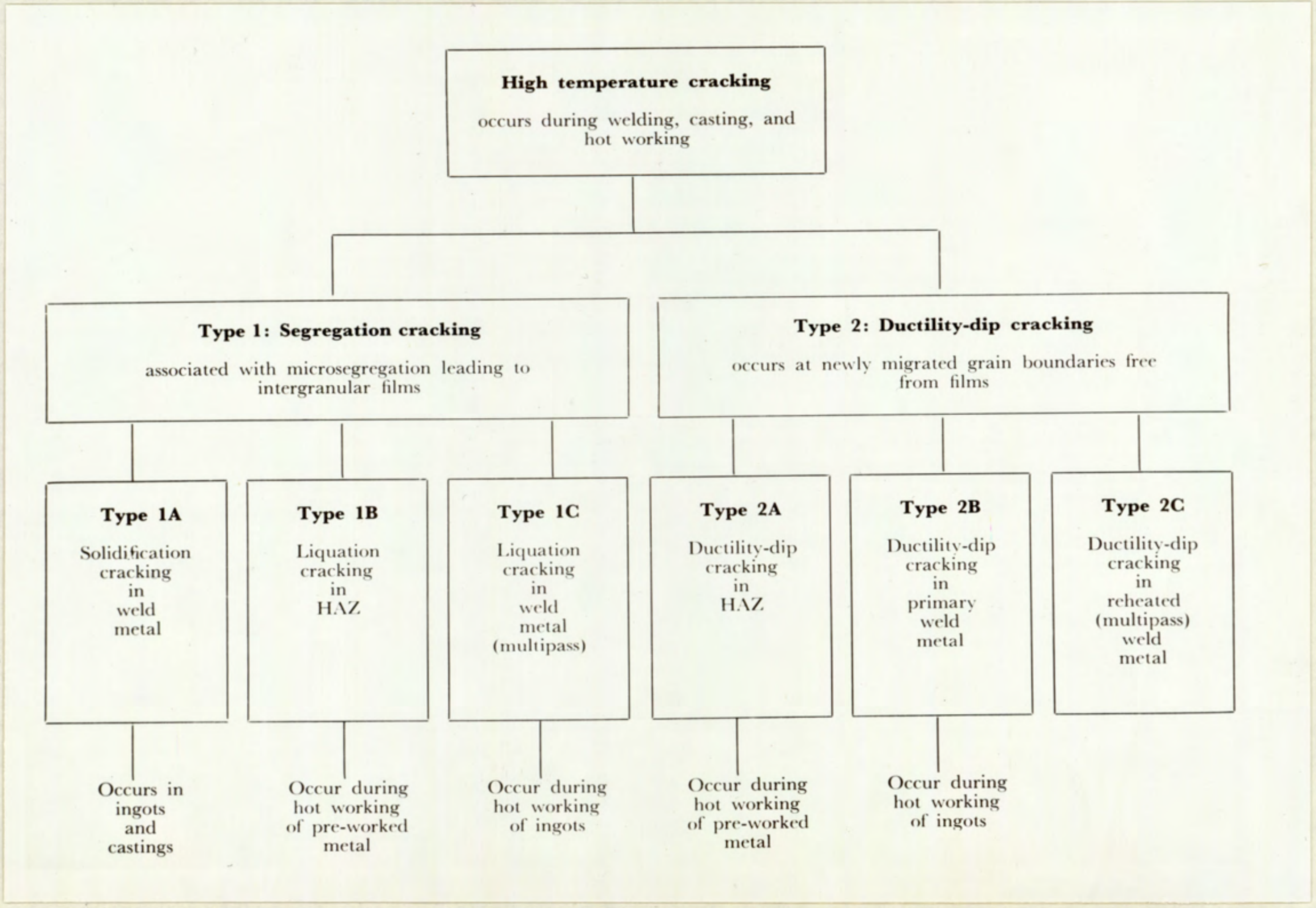


Fig.1. The Classification of High Temperature Welding Cracks.
(After Hemsworth et. al.[2])

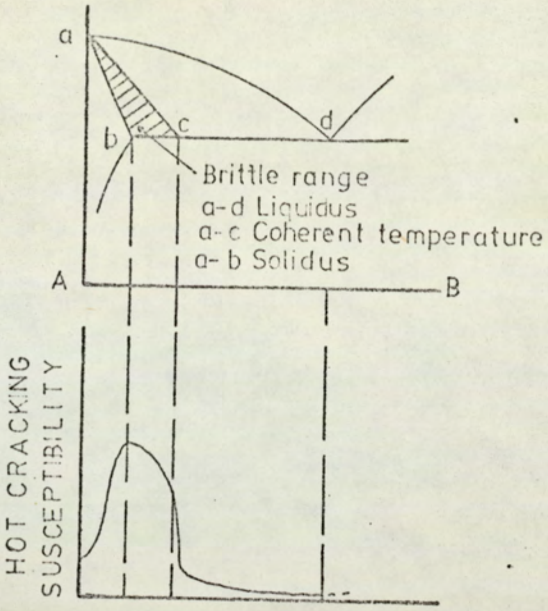
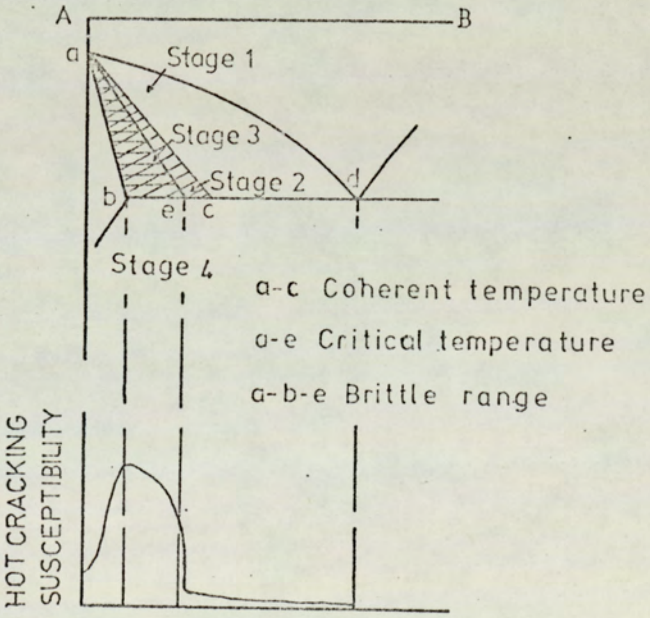


Fig.2. Solidification Cracking Susceptibility in relation to the Equilibrium Diagram (Shrinkage Brittleness Theory). (After Borland[51])



Stage 1. Dendrites dispersed freely in liquid - no cracking.
Stage 2. Interlocking of grains. Healing possible - "Accommodation" not important.
Stage 3. Critical solidification range - no healing possible if "accommodation" strain exceeded.
Stage 4. Solid - no cracking.

Fig. 3. Effect of Constitutional Features on Solidification Cracking in Binary Systems (Generalised Theory) (After Borland[51]).

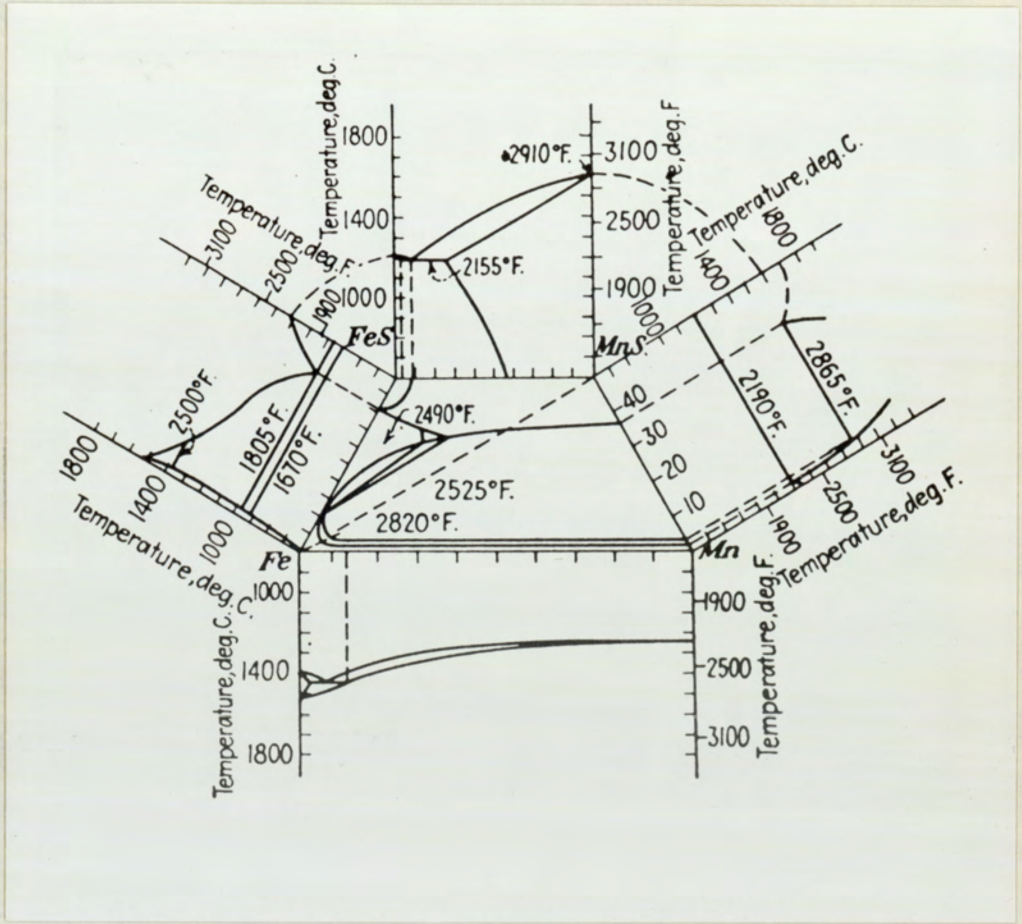


Fig. 4. The Iron-Manganese-Sulphur Ternary System, with the associated Binary Diagrams. (After Wentrup[62])

Note the liquid miscibility gap in the manganese-sulphur system, compared with the continuous liquid solubility between sulphur and iron.

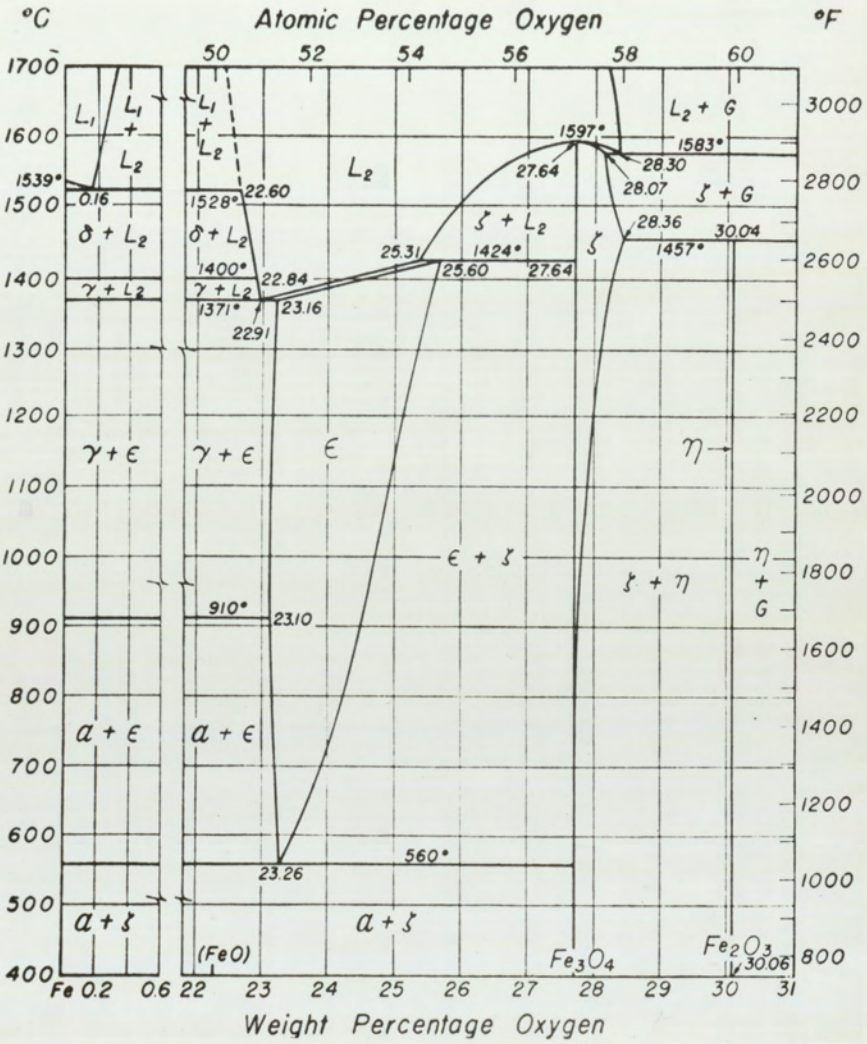


Fig. 5. The Iron-Oxygen Binary System (After Crafts and Hilty[58])

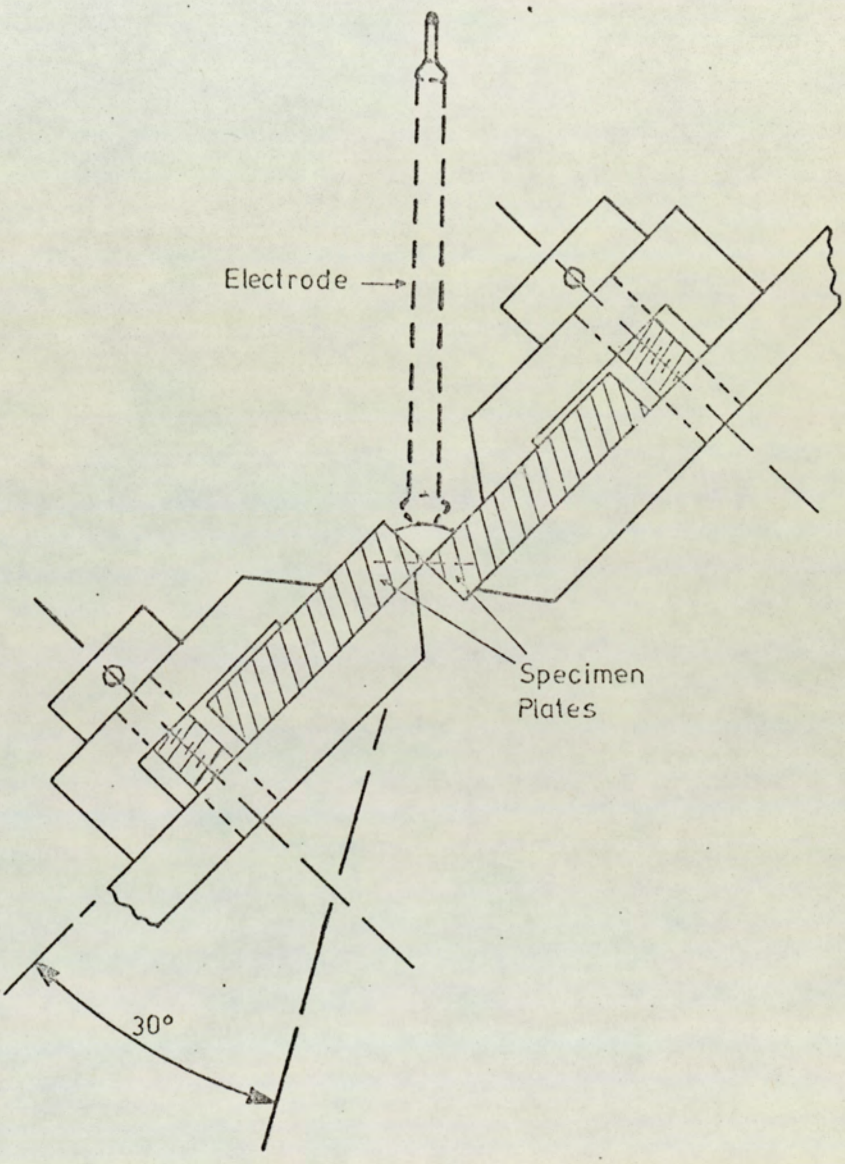


Fig. 6. Principle of the Murex Hot Crack Machine. (After Rollason and Roberts[98])

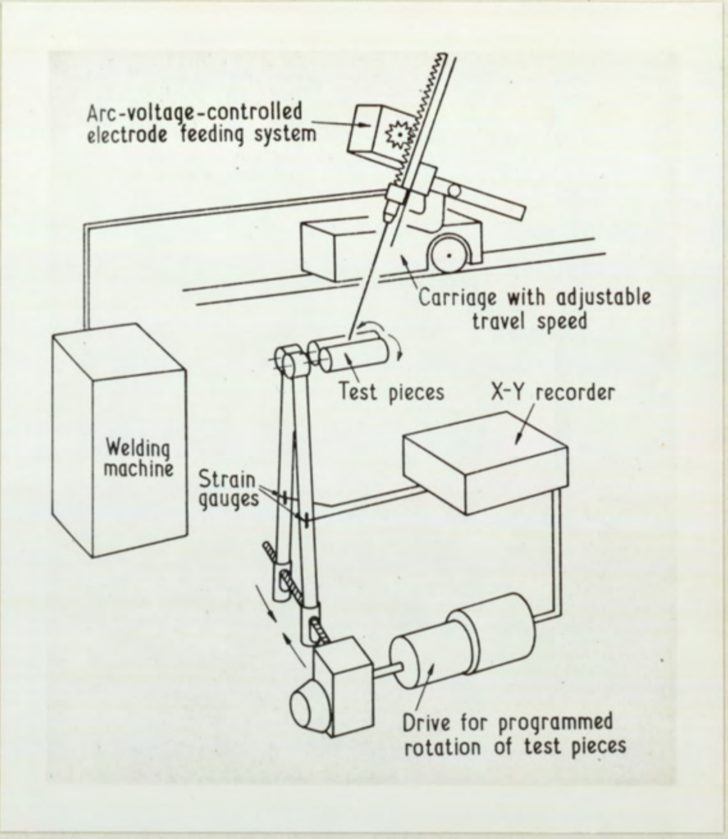


Fig. 7. Schematic Set-up of the K.S.L.A. Test. (After Fredriks and van der Toorn[73])

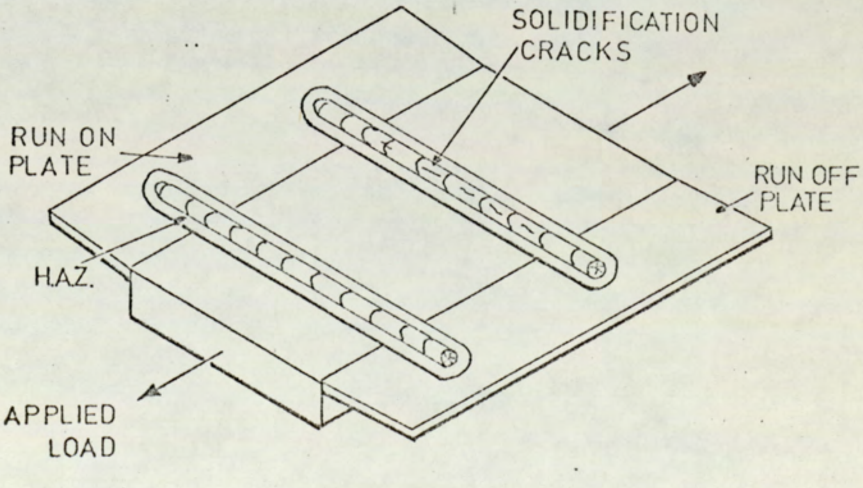


Fig. 8. Hemsworth's Crack Susceptibility Apparatus[76].

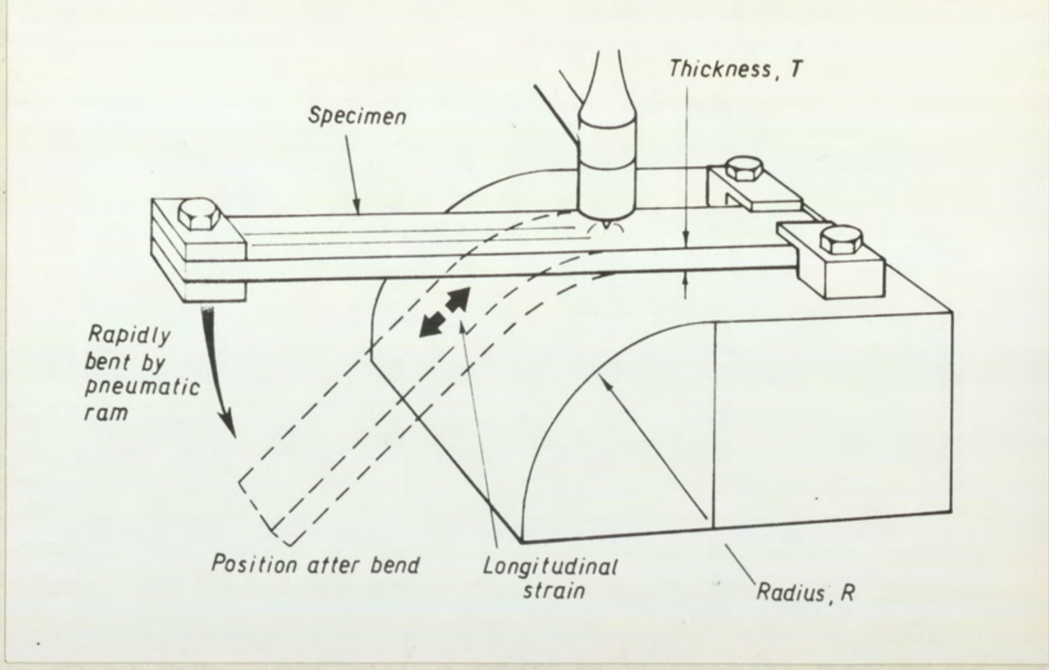


Fig. 9. The Principle of the Varestraint Test. (After Baker et. al.^{78})

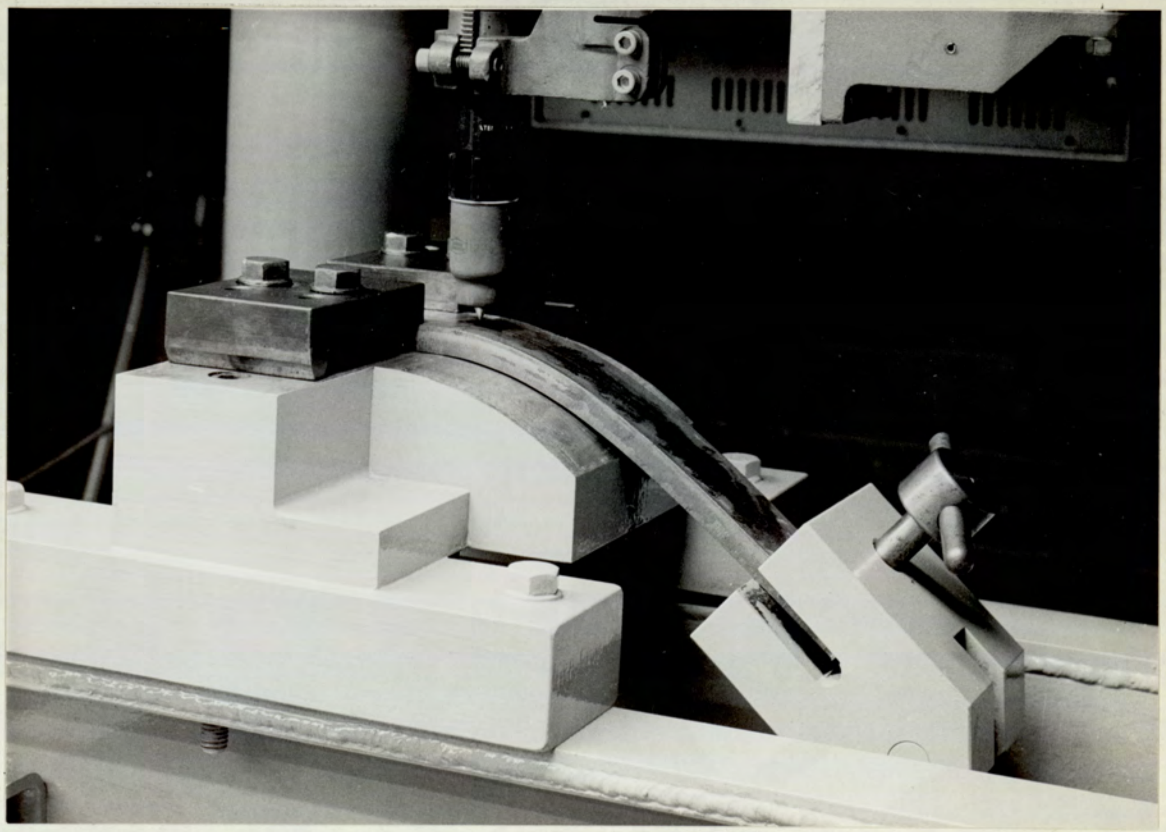


Fig. 10. The Varestraint Test Apparatus used in the present investigations, shown at the completion of welding.

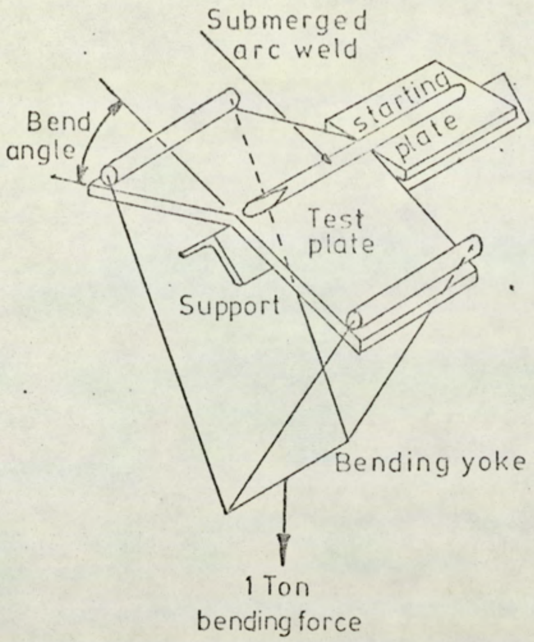


Fig. 11. Principle of the Brockhurst and Muir Test[77].

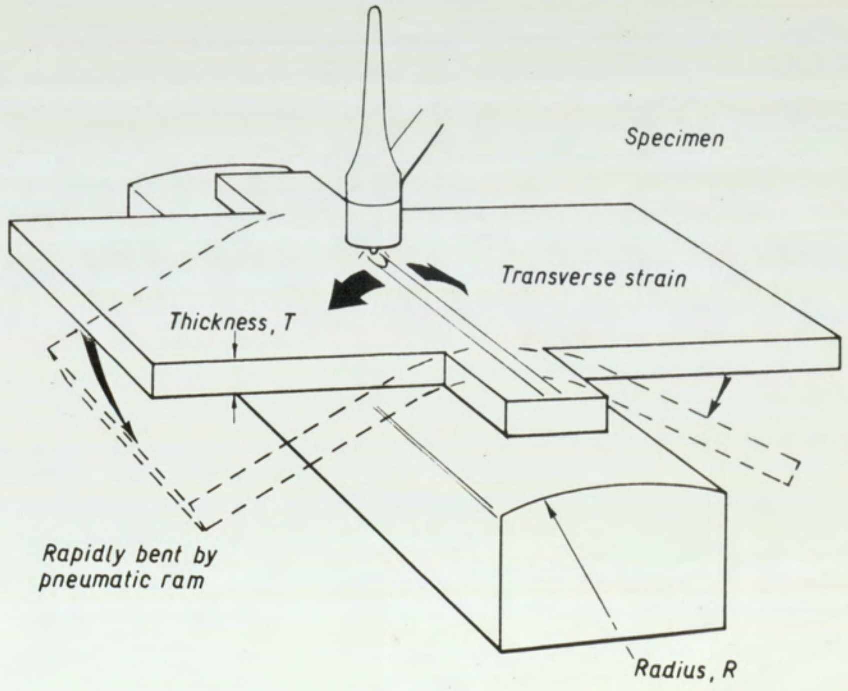


Fig. 12. Principle of the Transvarestraint Test.
(After Baker et. al.[78])

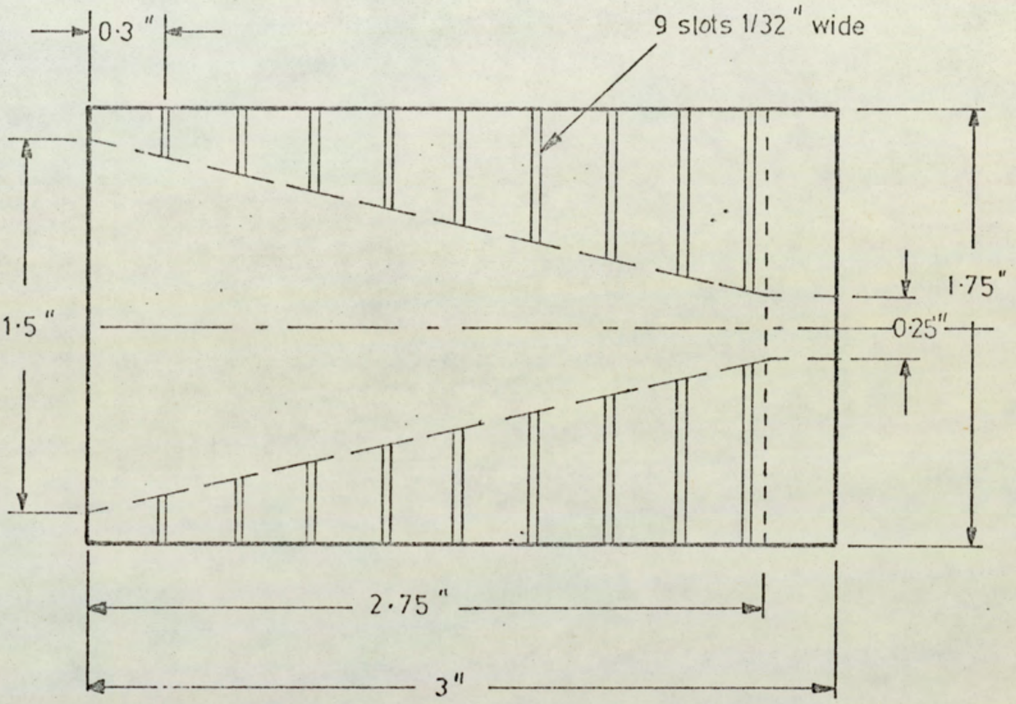


Fig. 13. Specimen used for the Houldcroft Fishbone Test. (After Kammer et. al.[82])

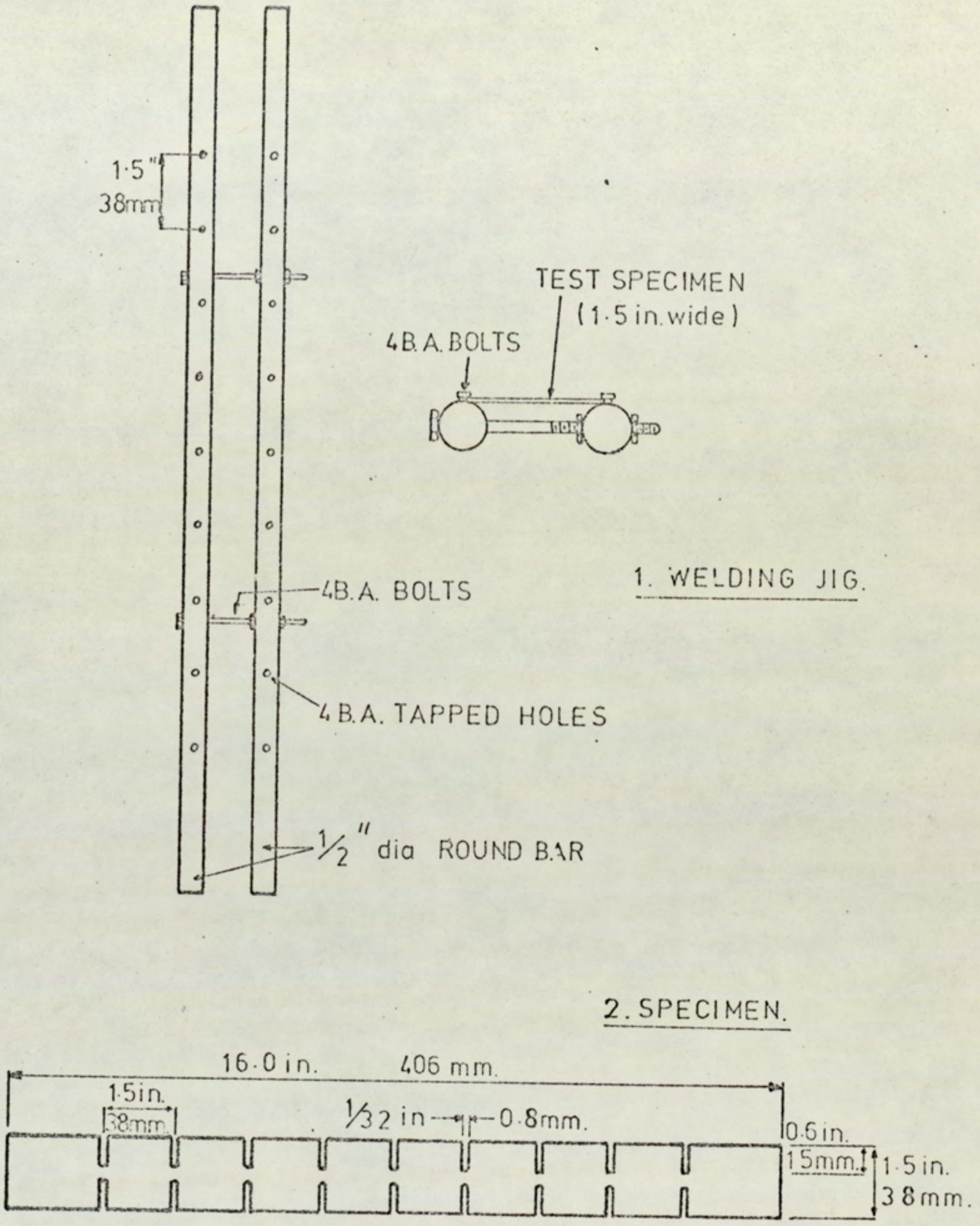


Fig. 14. Diagram of the Welding Jig and Specimen used for the Huxley Test[80].

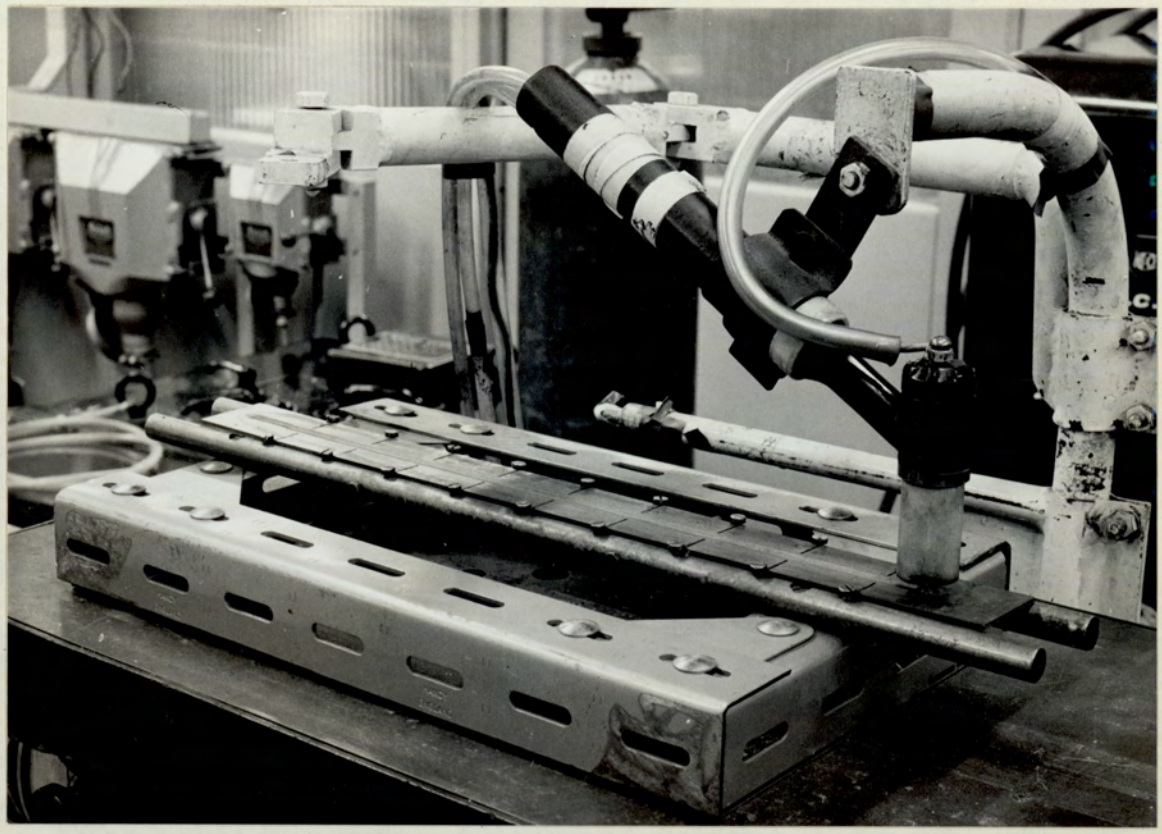


Fig. 15. The Huxley Test Apparatus used in the present investigations, shown at the completion of welding.

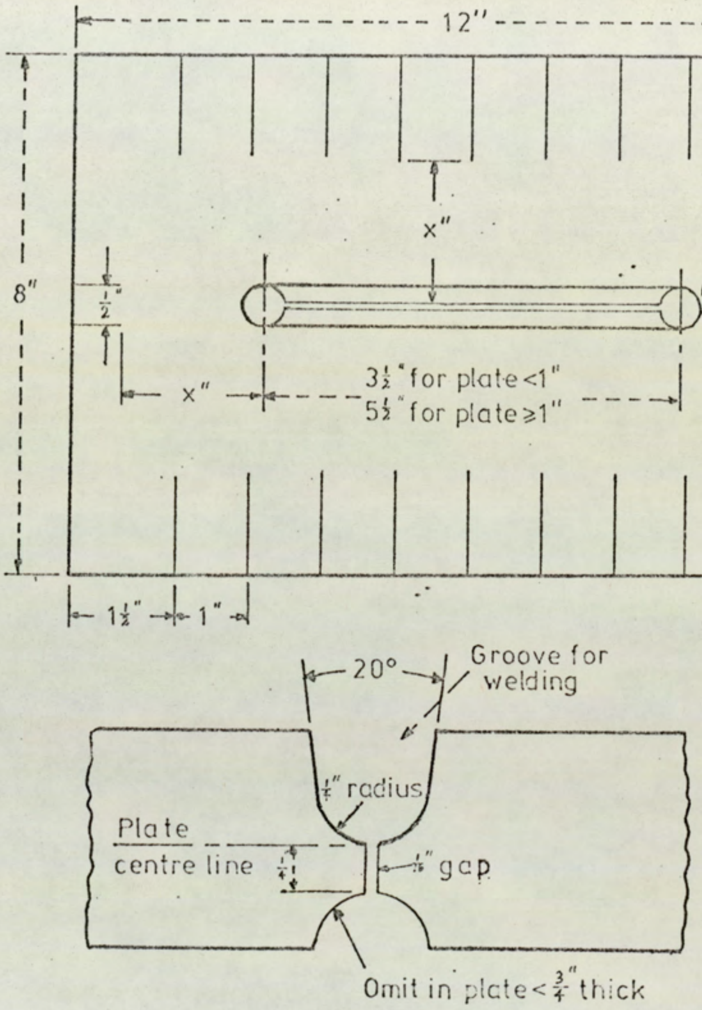


Fig. 16. The Lehigh Restraint Test Specimen (After Kammer et. al.[82])

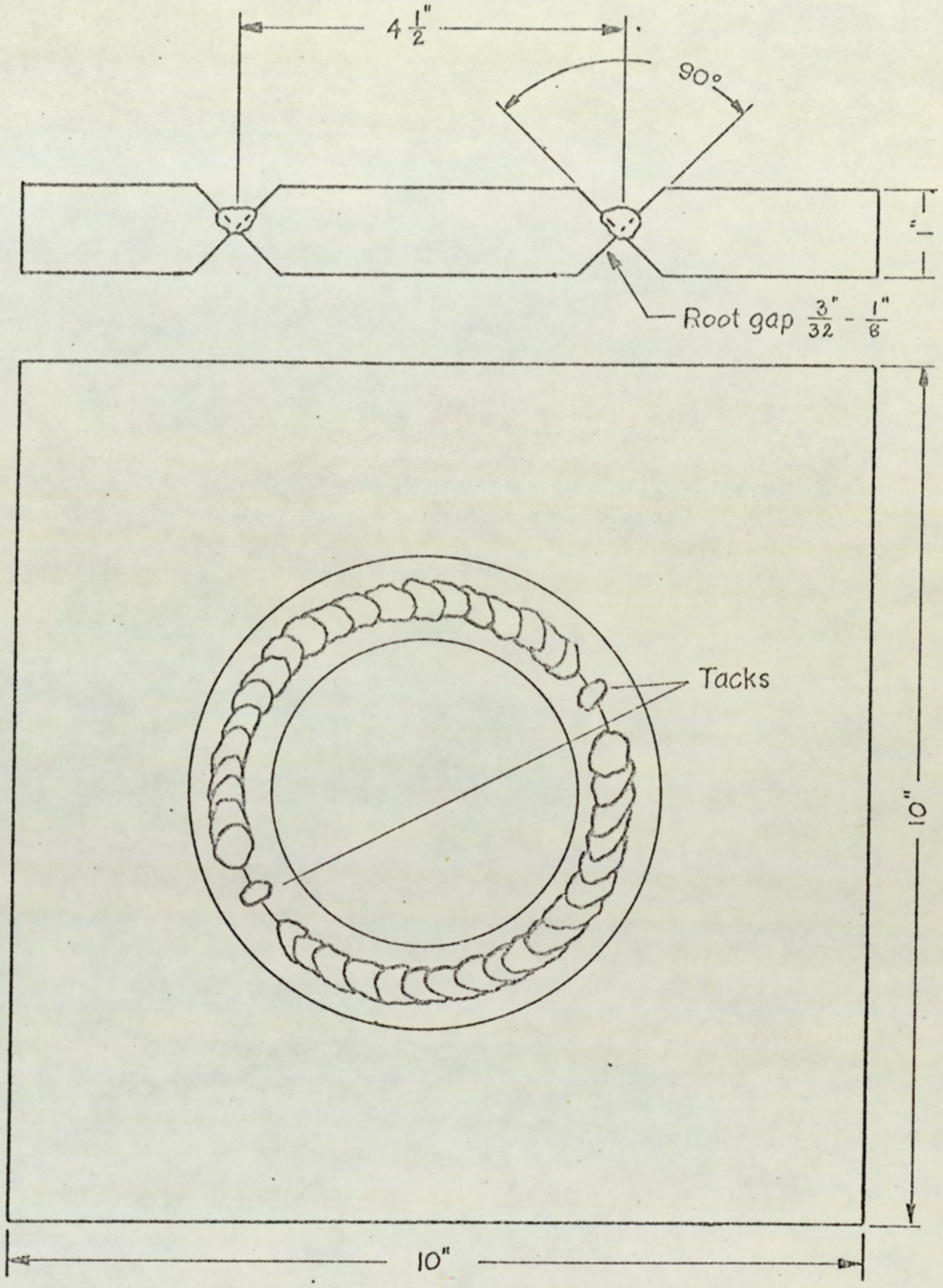


Fig. 17. Specimen for the Circular Patch Test
(After Williams et. al.[19])

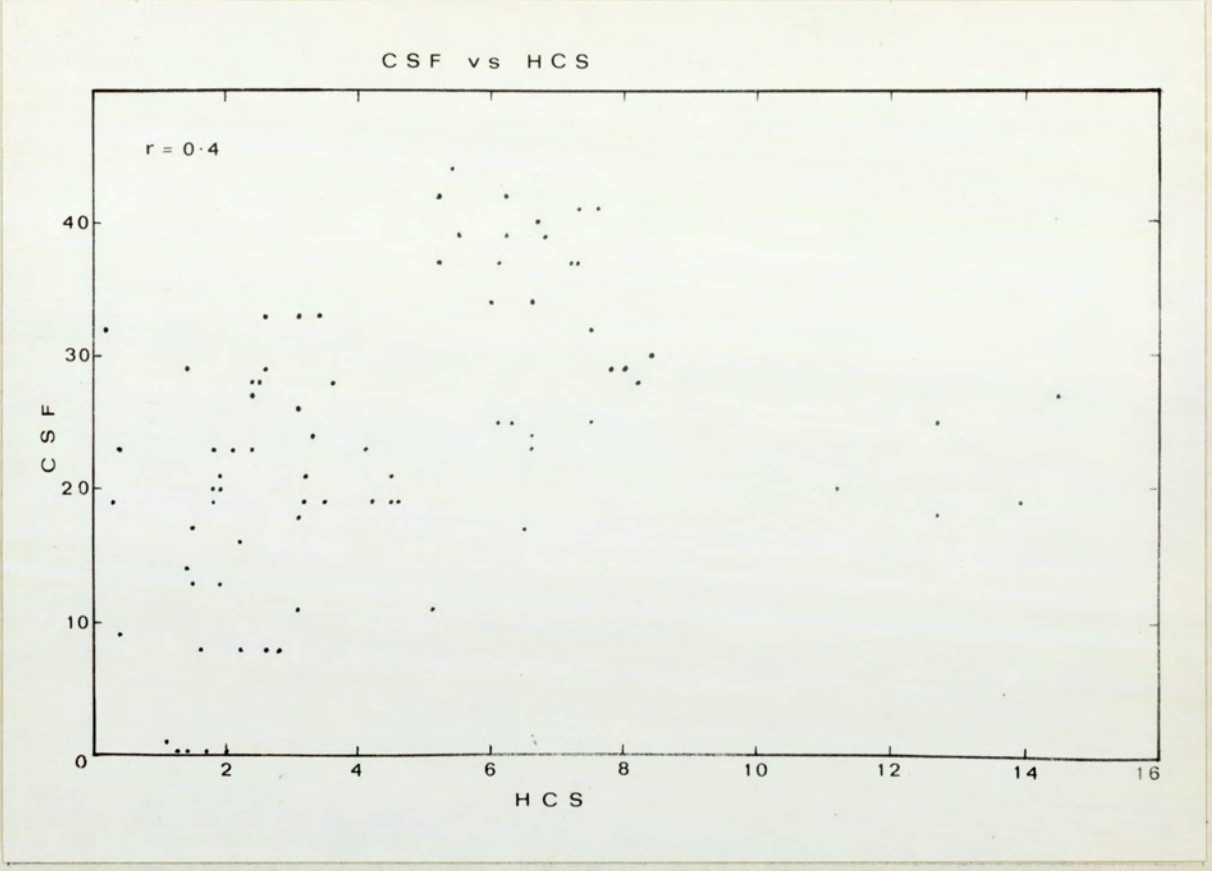


Fig. 18. The Relationship between Observed Crack Susceptibility (C.S.F.) and the H.C.S. Value calculated using Wilkinson's Formula[7].

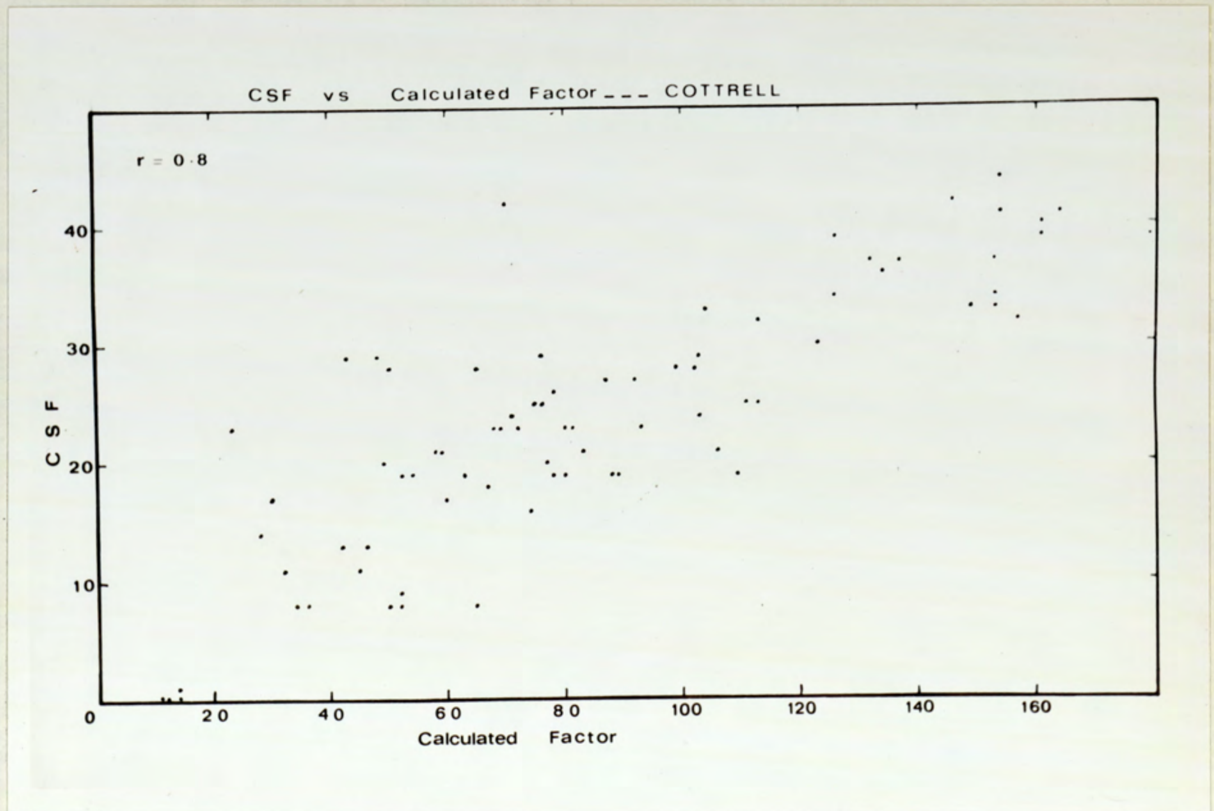


Fig. 19. The Relationship between Observed Crack Susceptibility (C.S.F) and the Factor calculated using Cottrell's Formula[42].

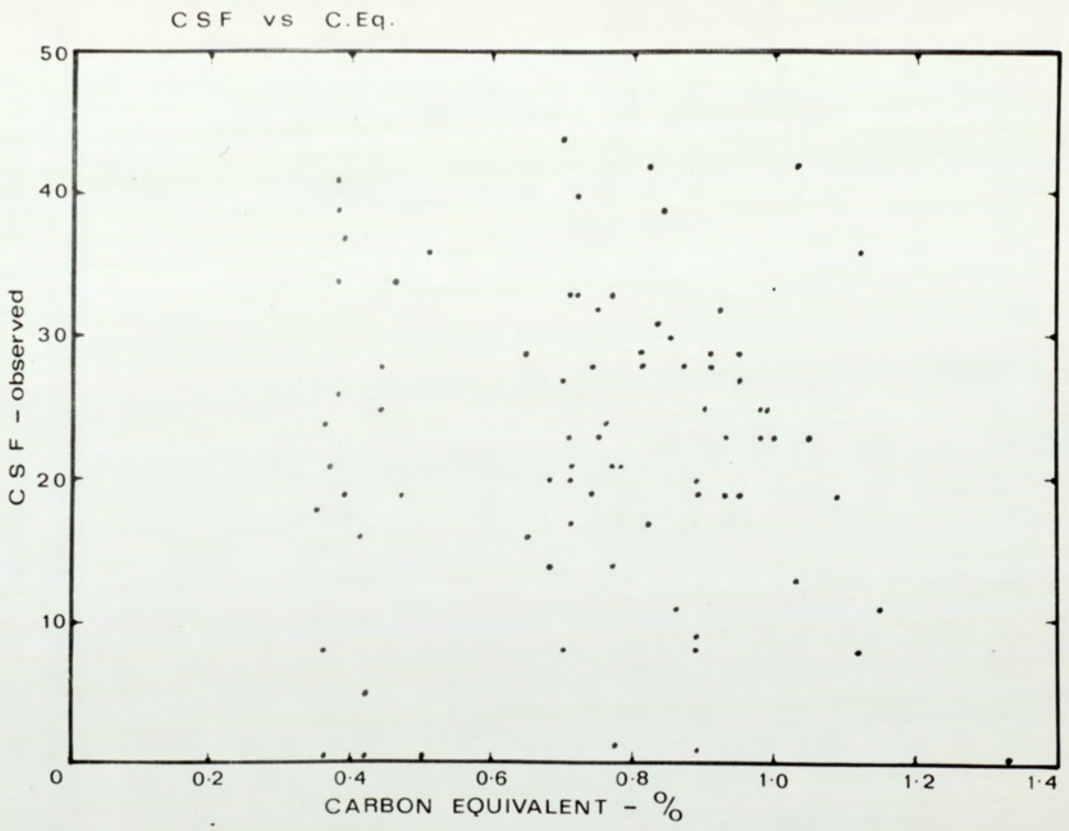


Fig. 20. The Relationship between C.S.F. and the Carbon Equivalent calculated using Ostrovskaya's Formula[11]

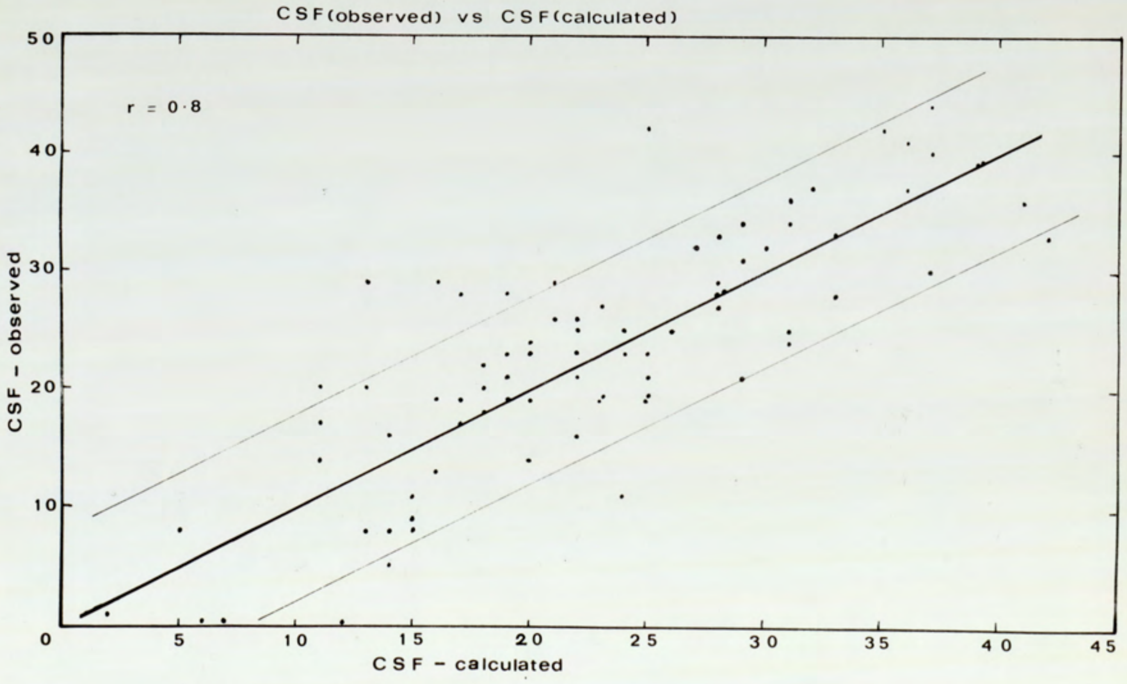


Fig. 21. The Relationship between the Observed and Calculated C.S.F. values using the Expression:
$C.S.F = 36C+12Mn+8\ Si + 540S+812P+5\ Ni+3.5\ Co-20V-13$
The 80% Confidence Limits are shown. (Present Work)

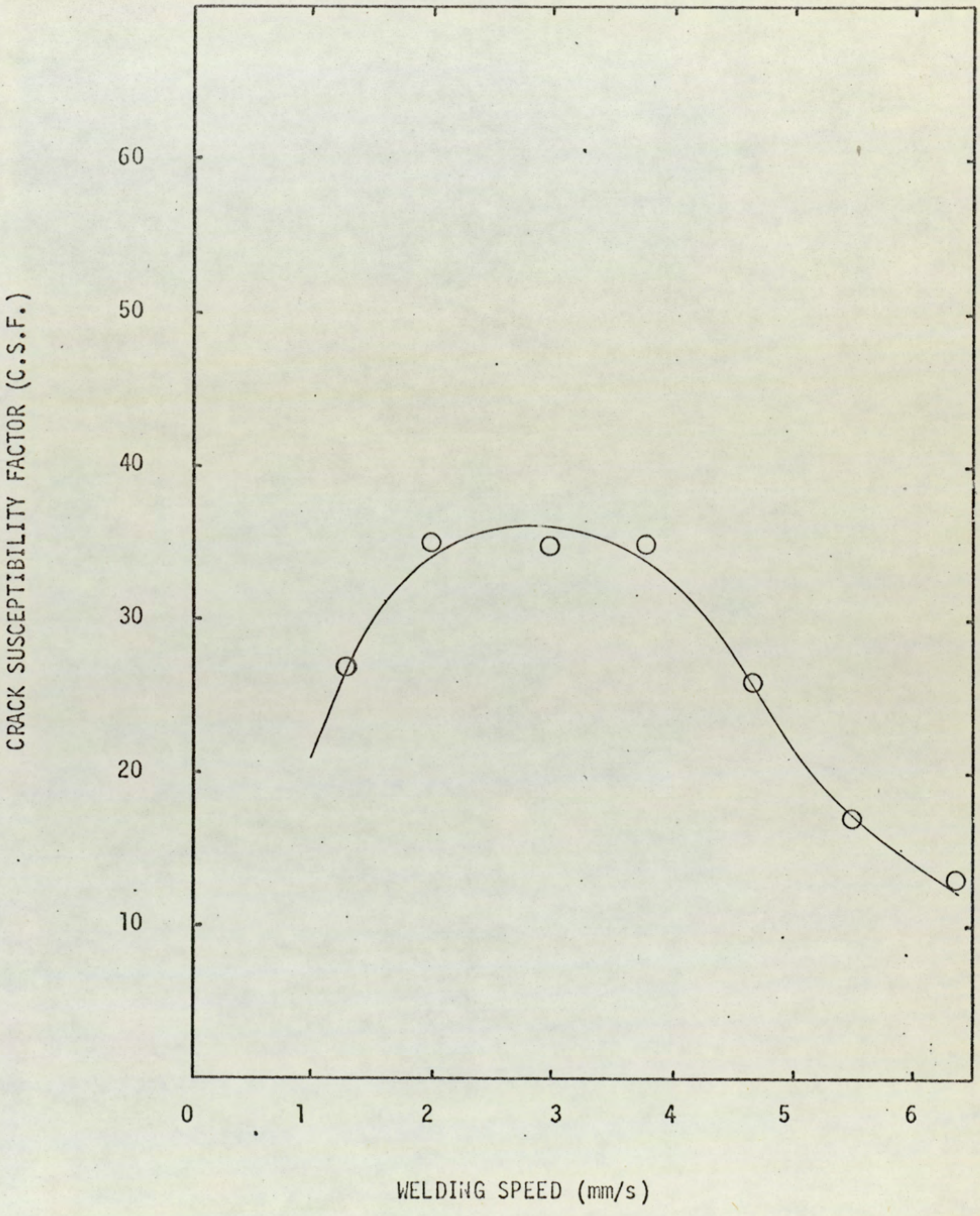


Fig. 22. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

Aircraft Steel RS120

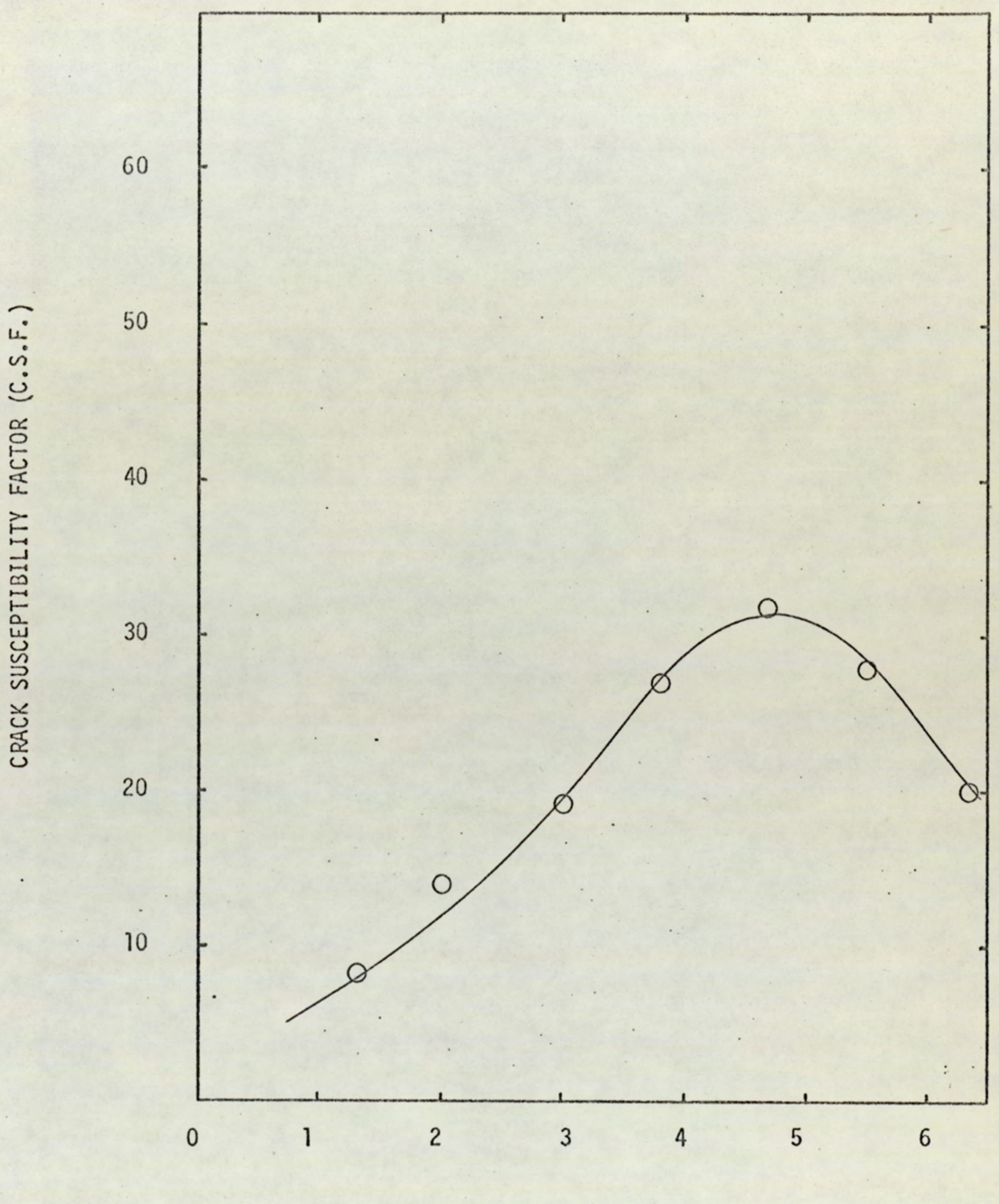


Fig. 23. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

Aircraft Steel RS130

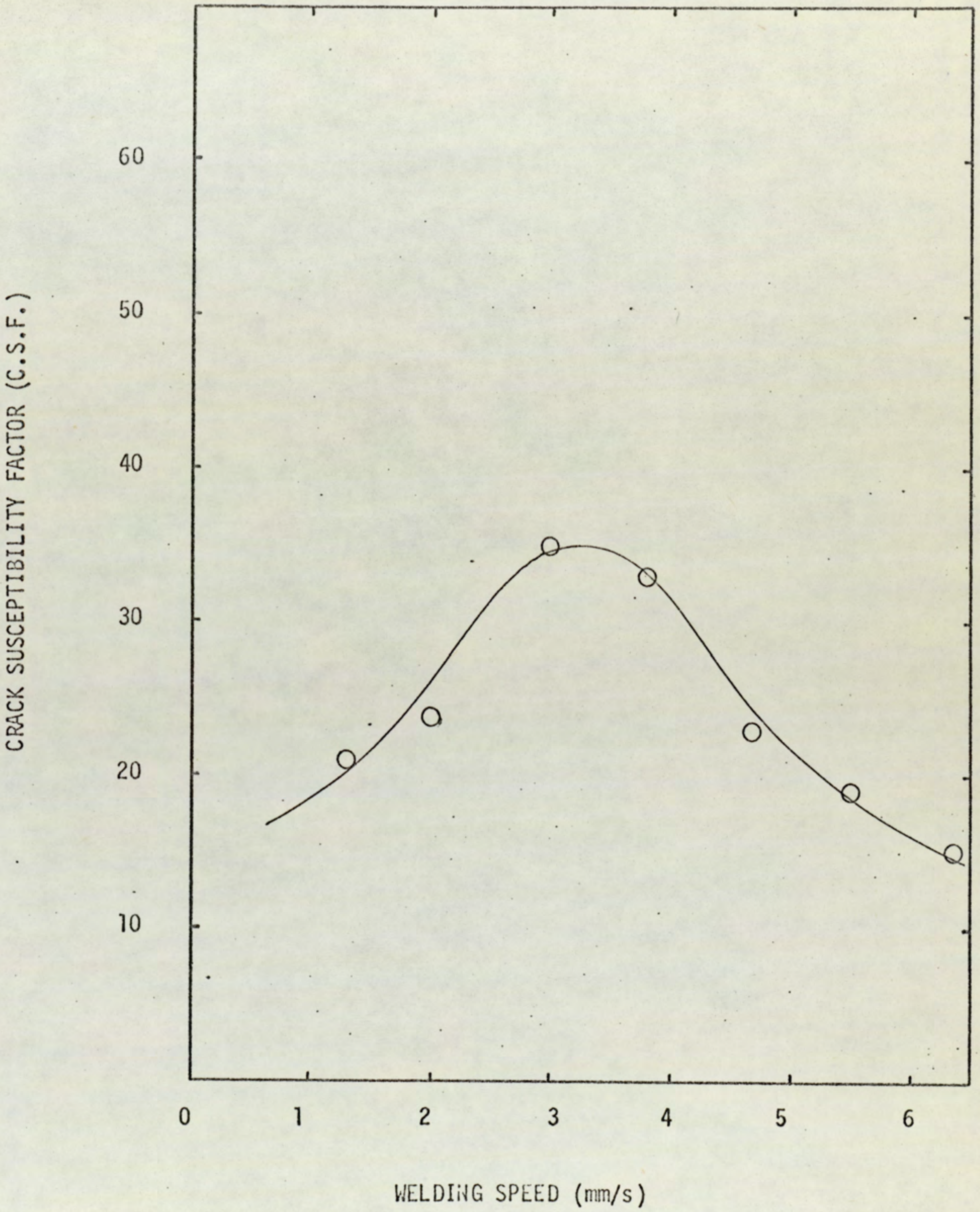


Fig. 24. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

Aircraft Steel RS140

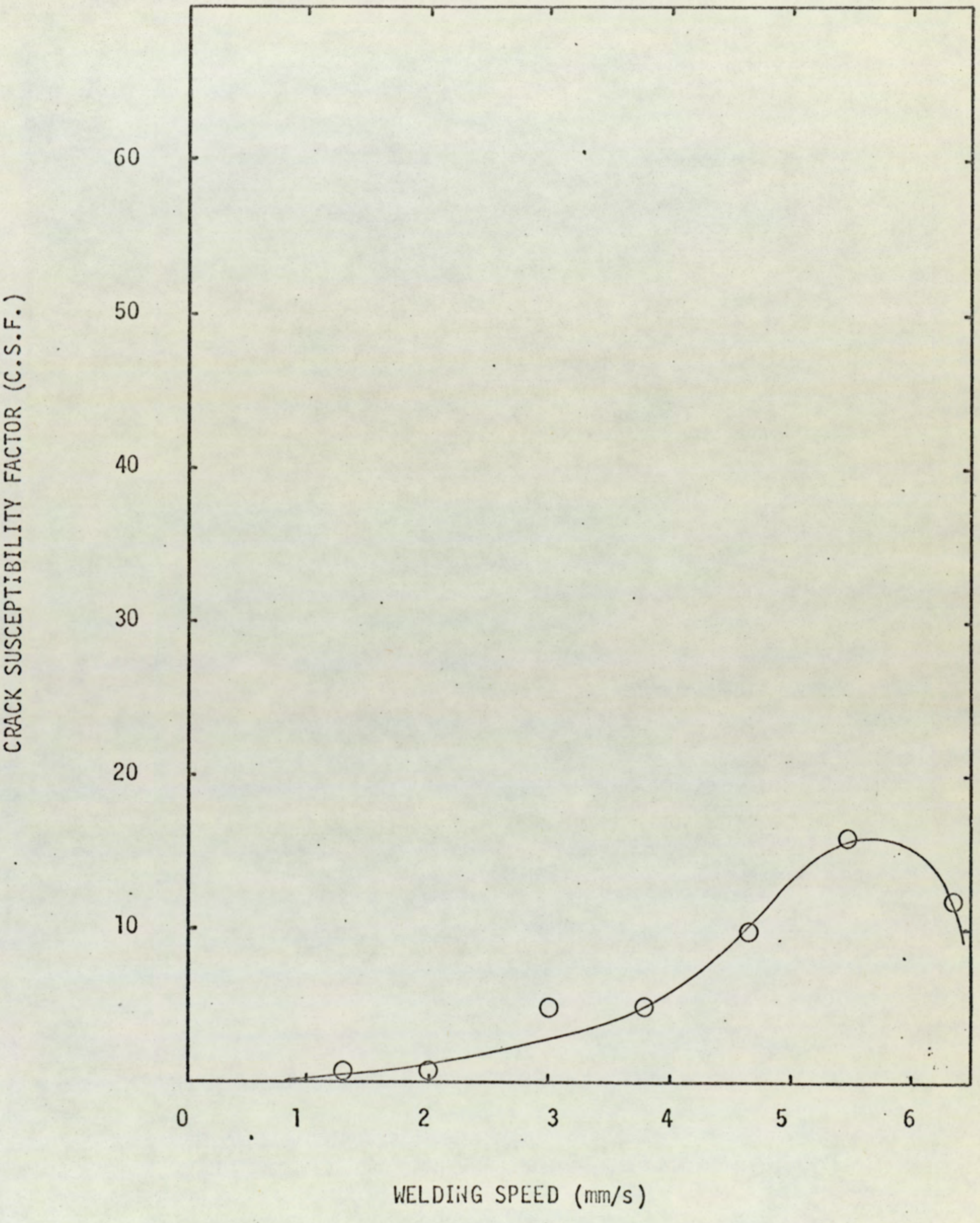


Fig. 25. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

ASTM A387B Steel AC

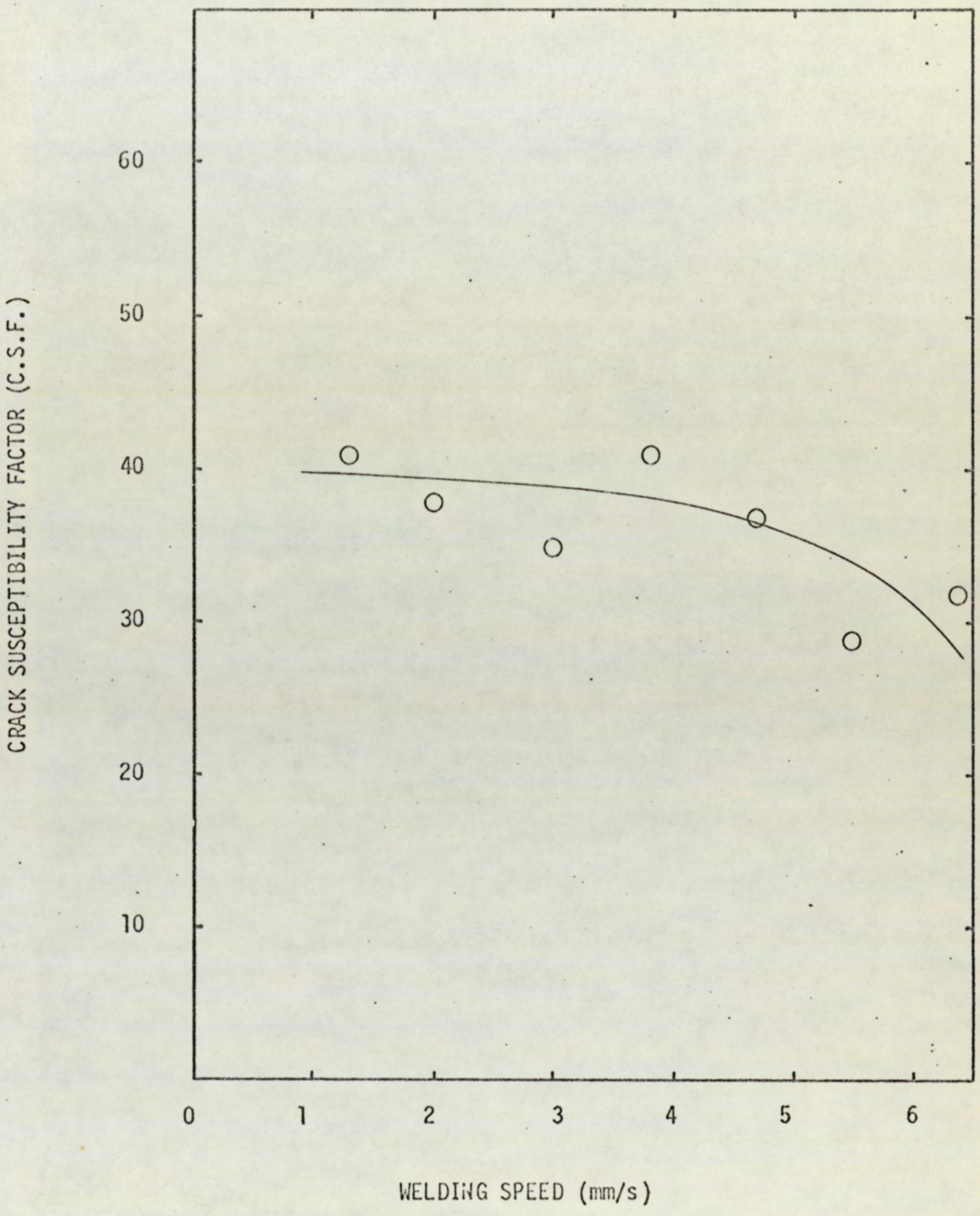


Fig. 26. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

ASTM A387B Steel AD

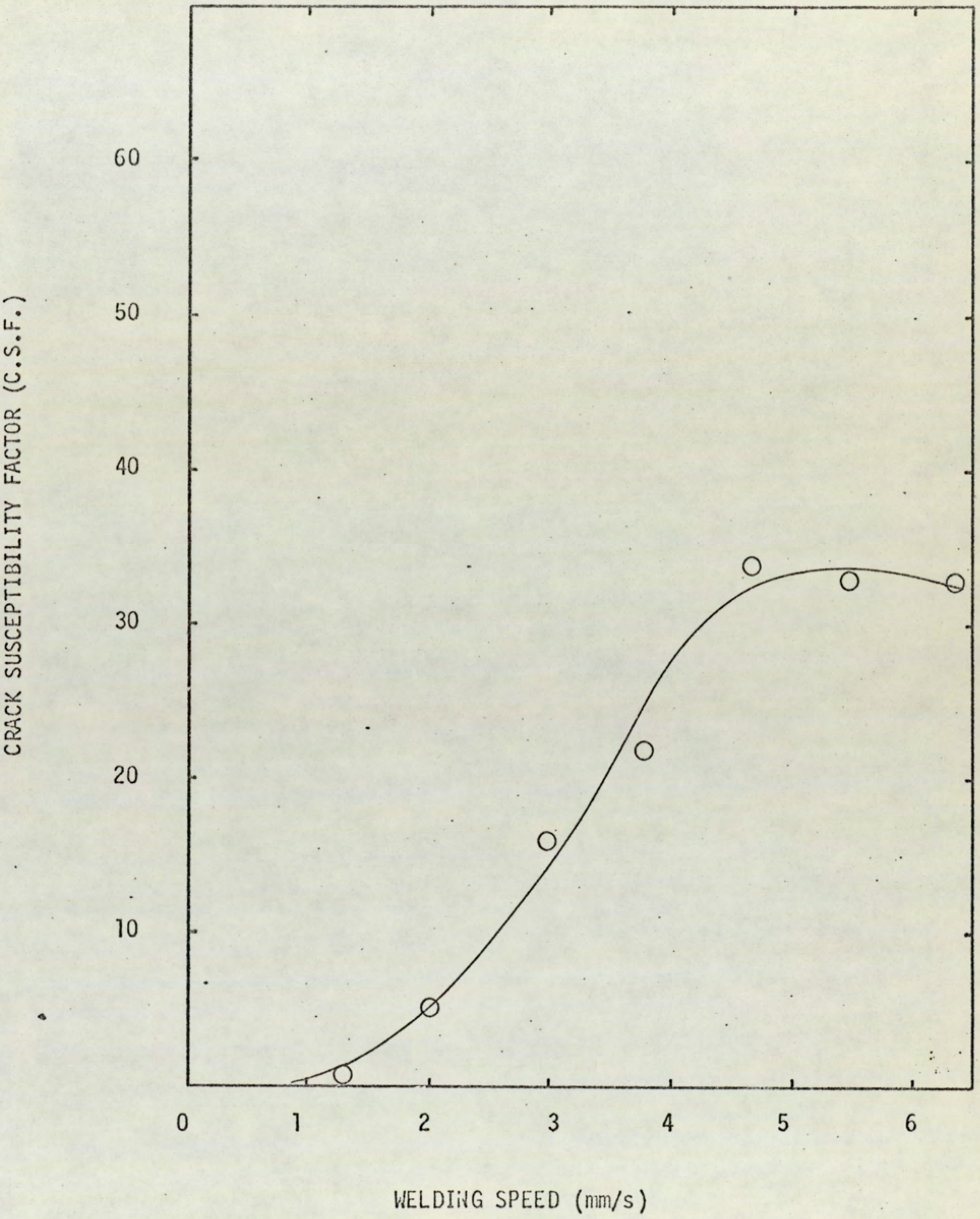


Fig. 27. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

ASTM A387B Steel AA4

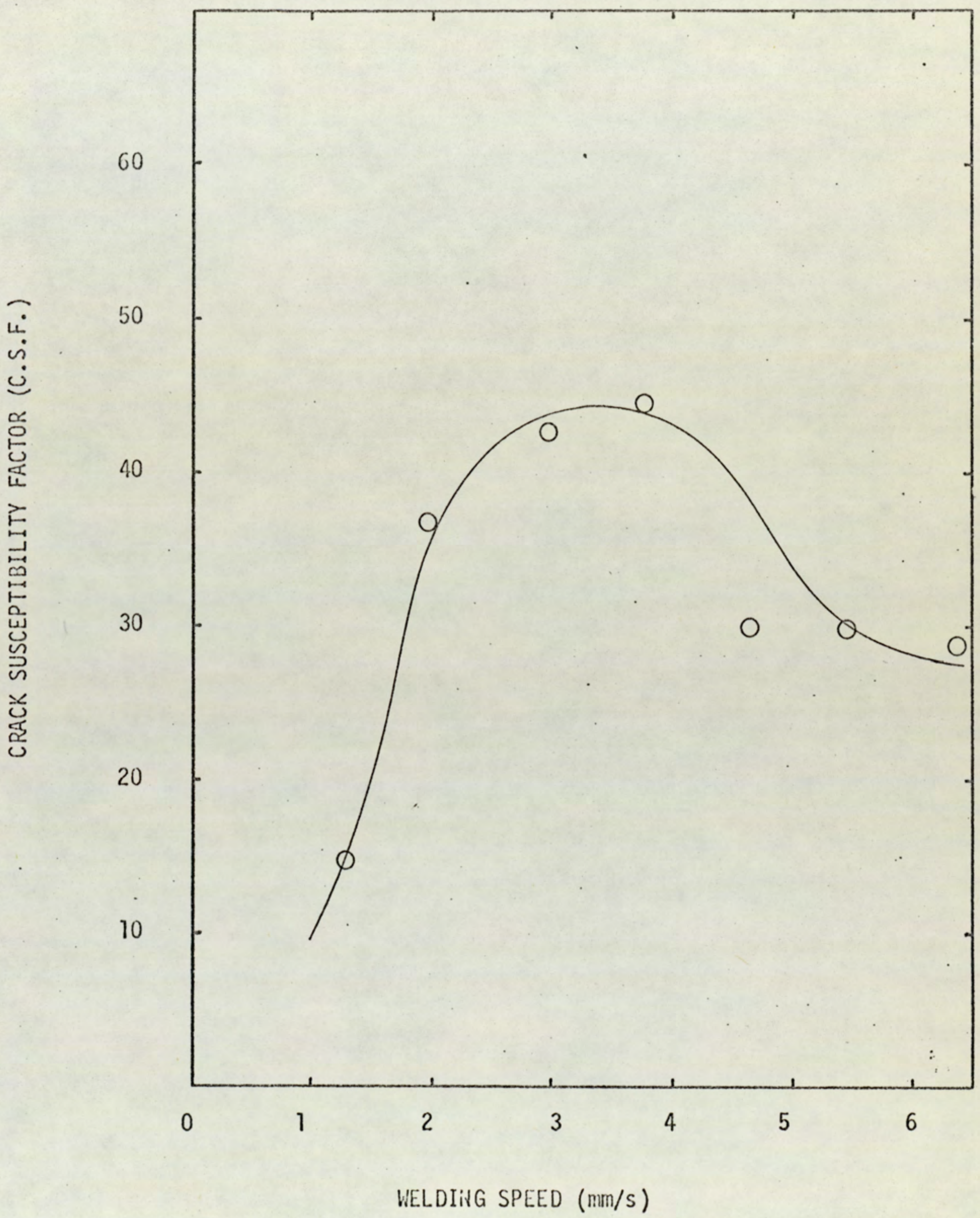


Fig. 28. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

ASTM A387B Steel AA5

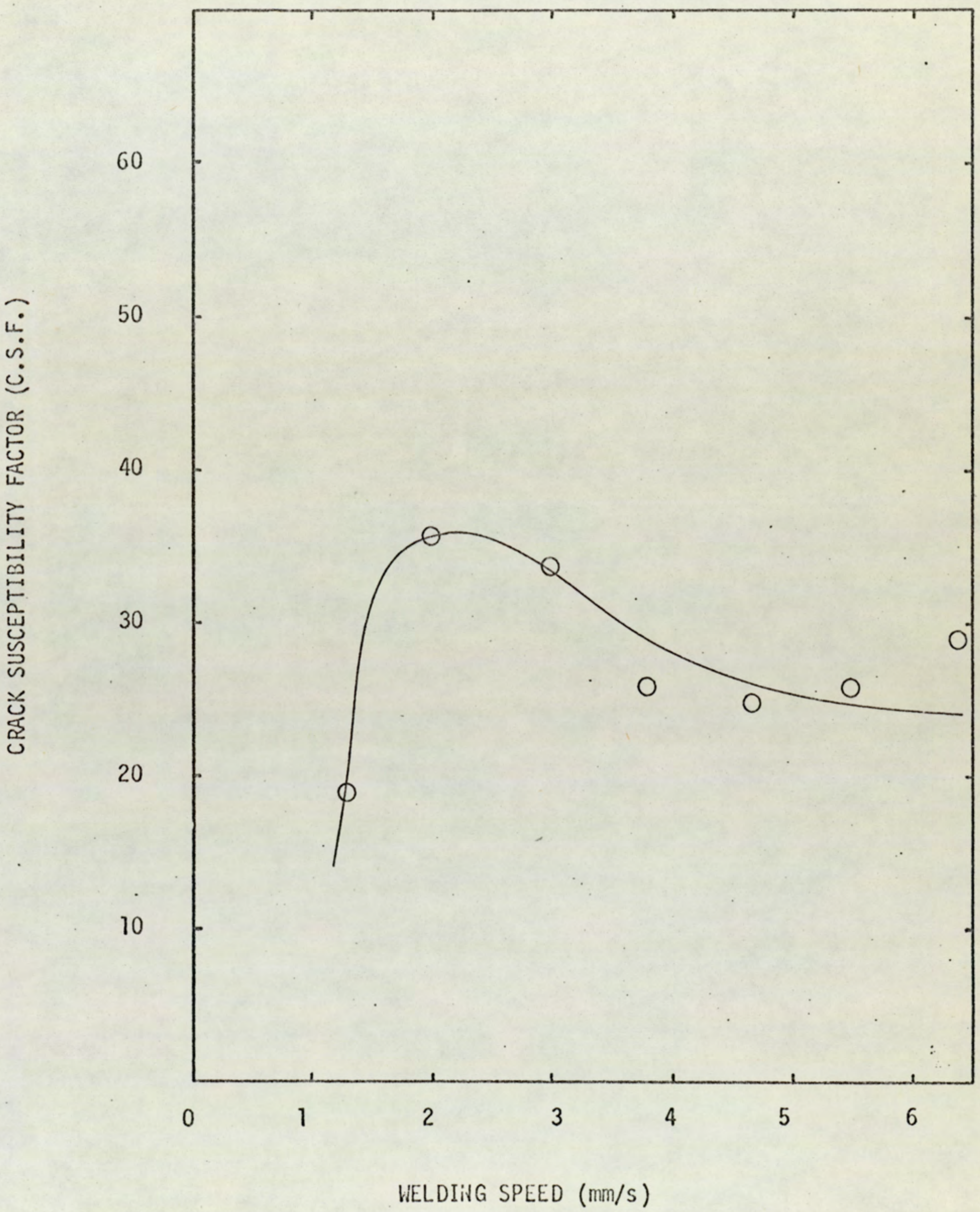


Fig. 29. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

En24 Steel EB

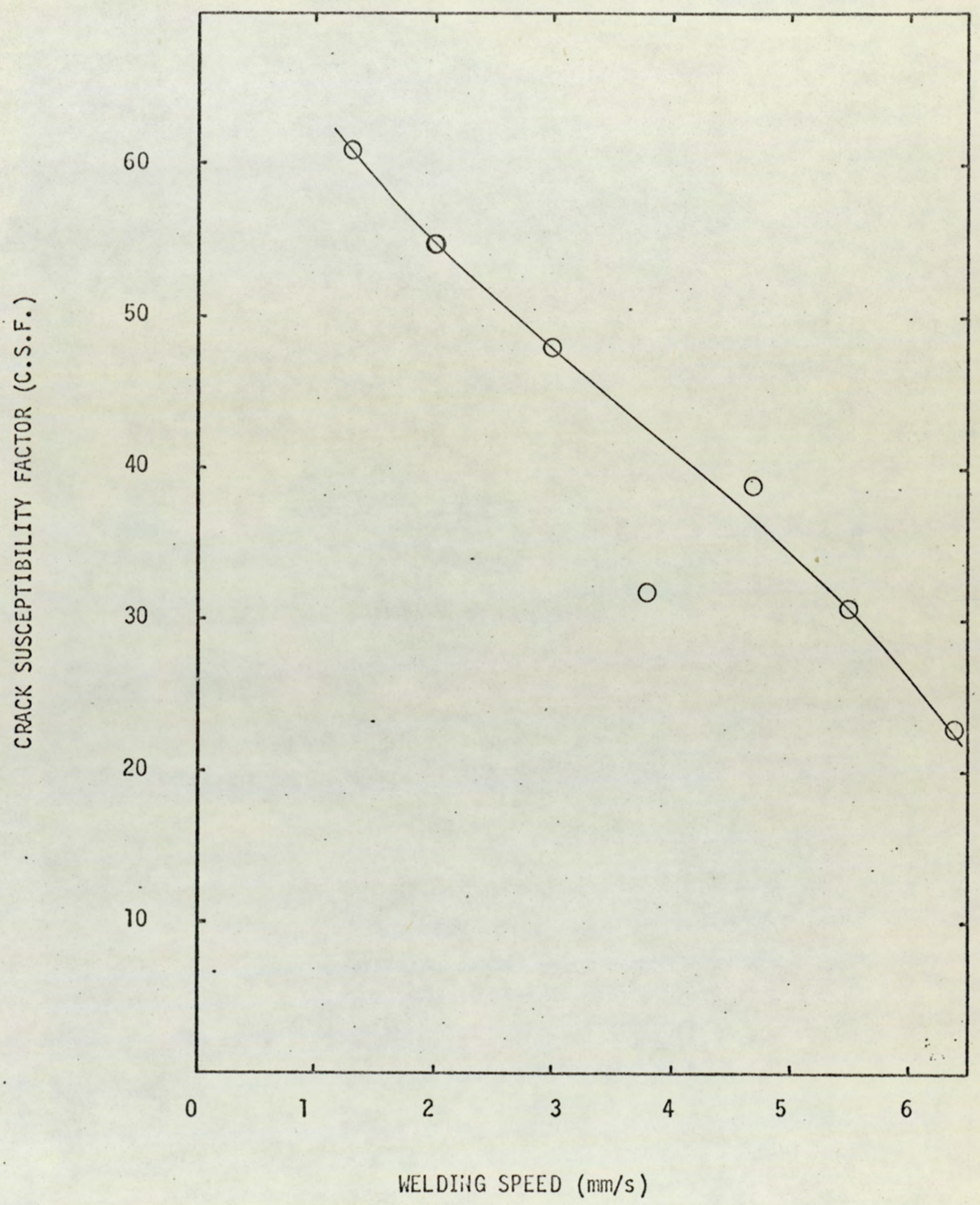


Fig. 30. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

En24 Steel EC

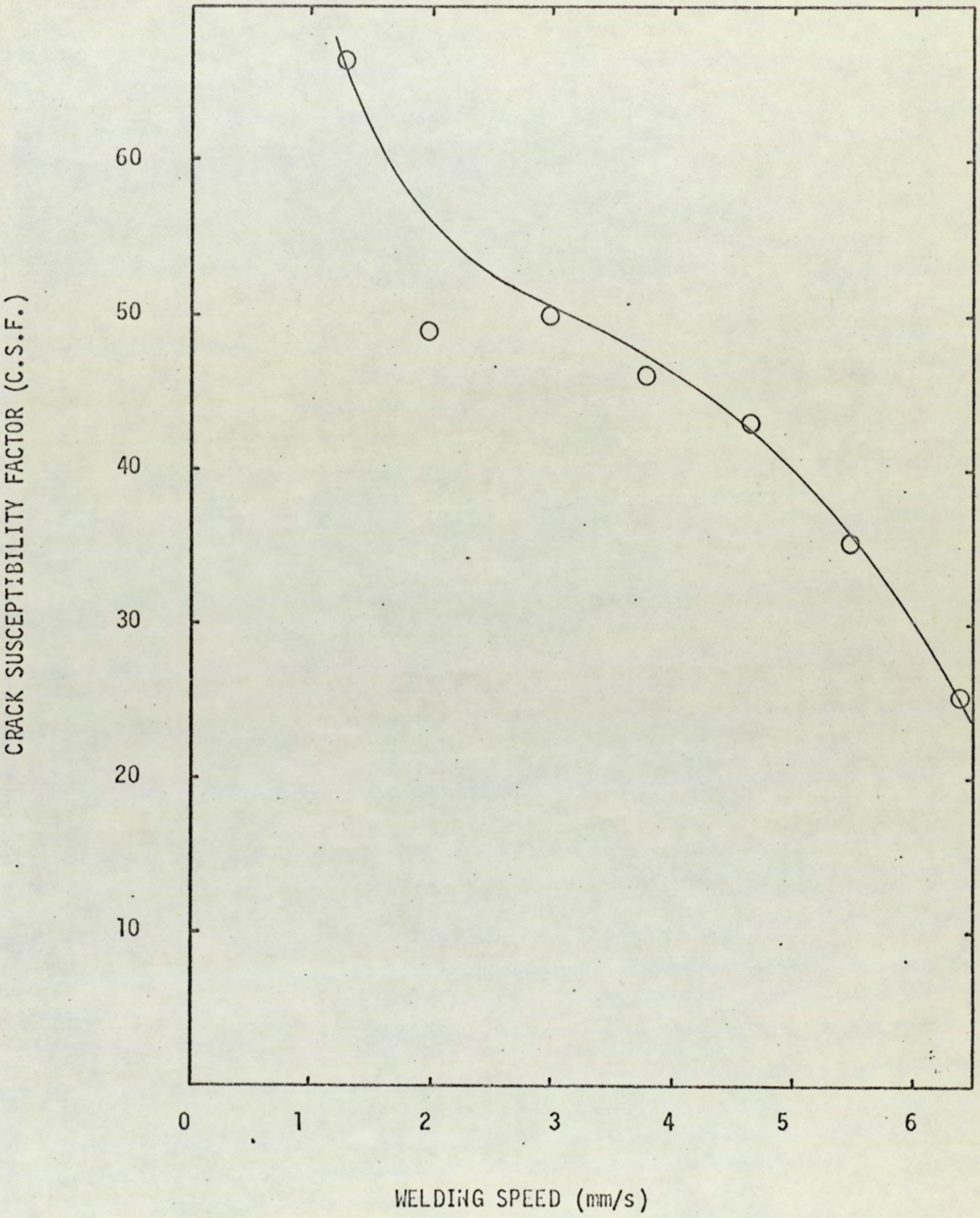


Fig. 31. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

En24 Steel EA4

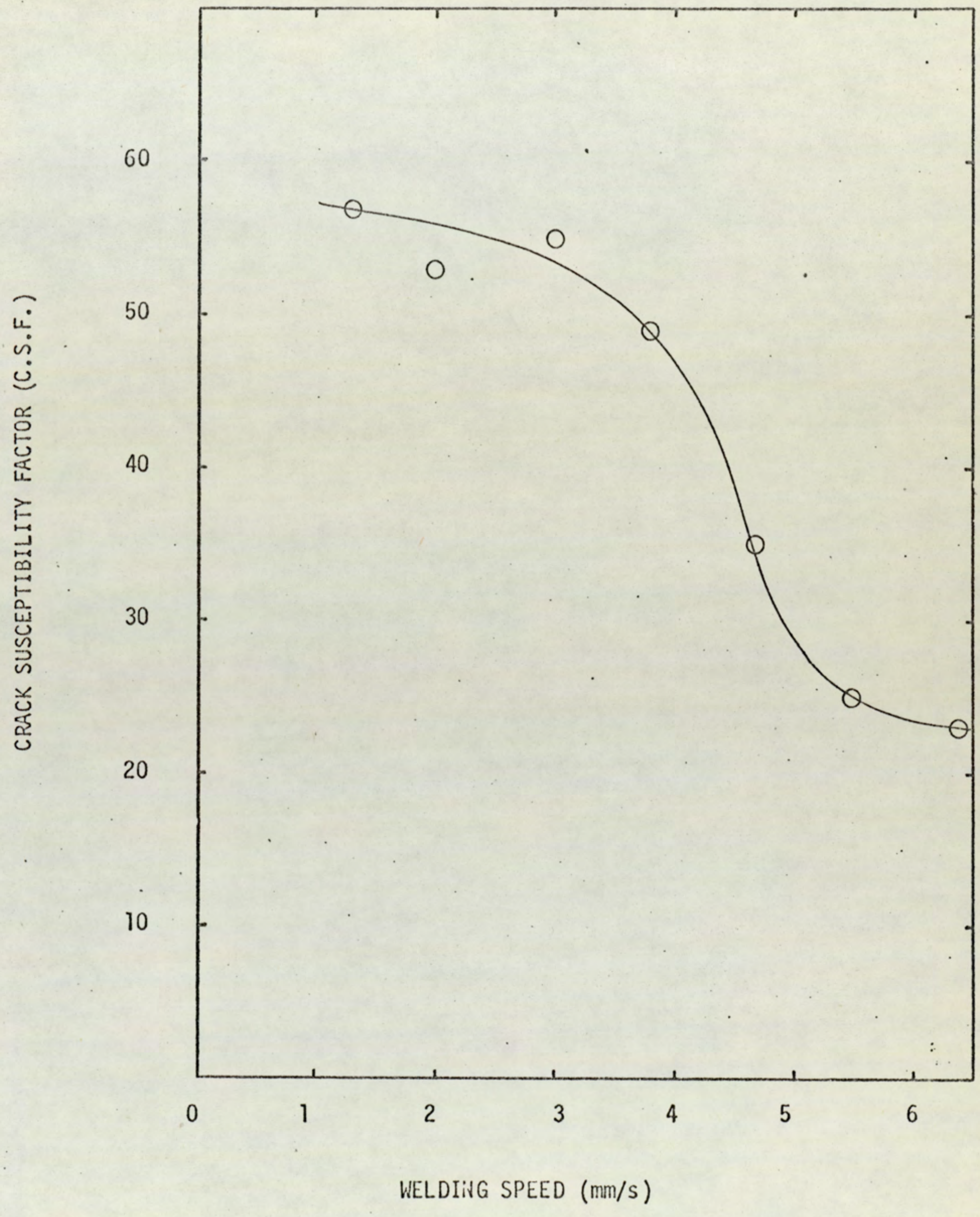


Fig. 32. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

En24 Steel EA5

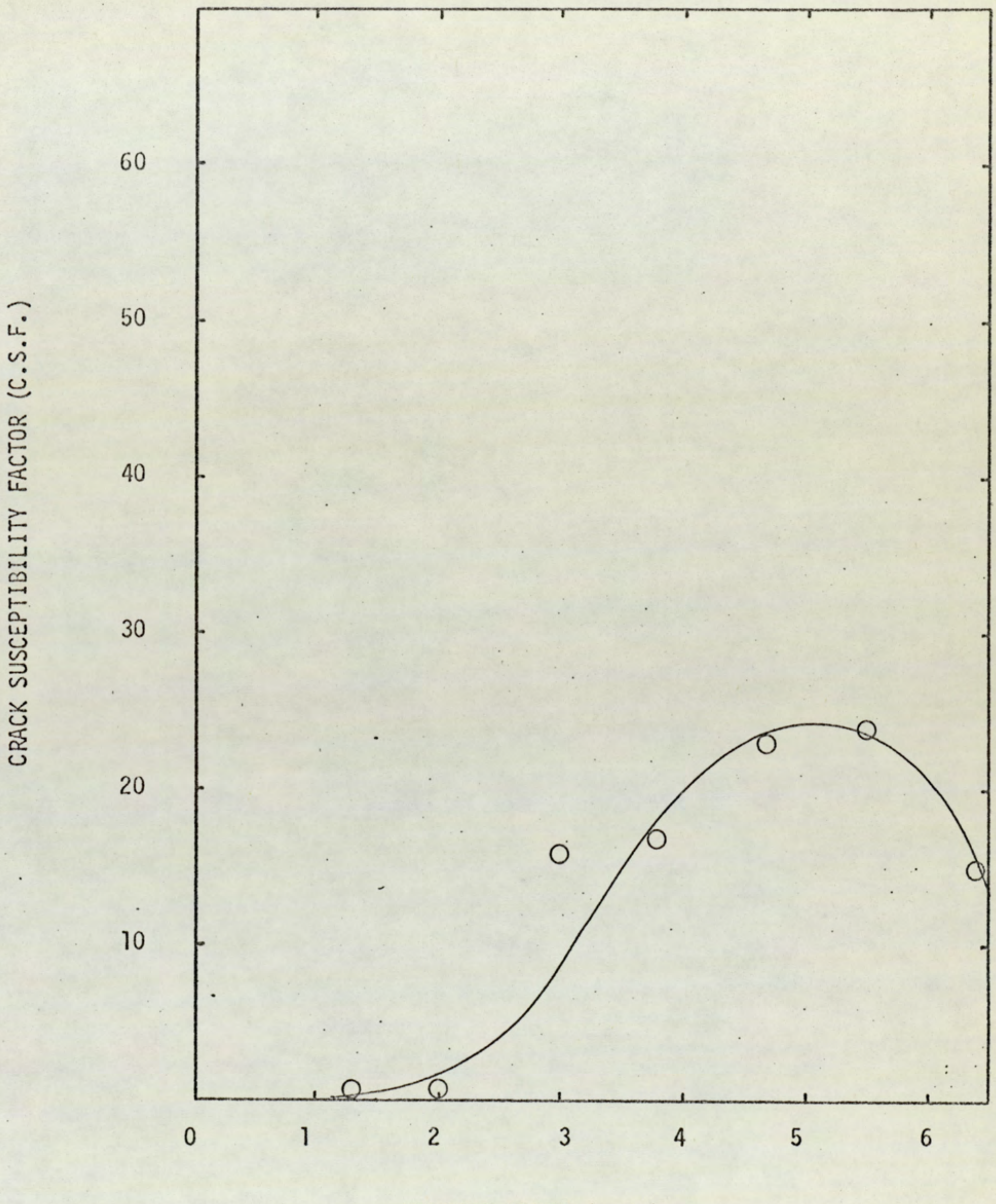


Fig. 33. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

SAE4130 Steel SB

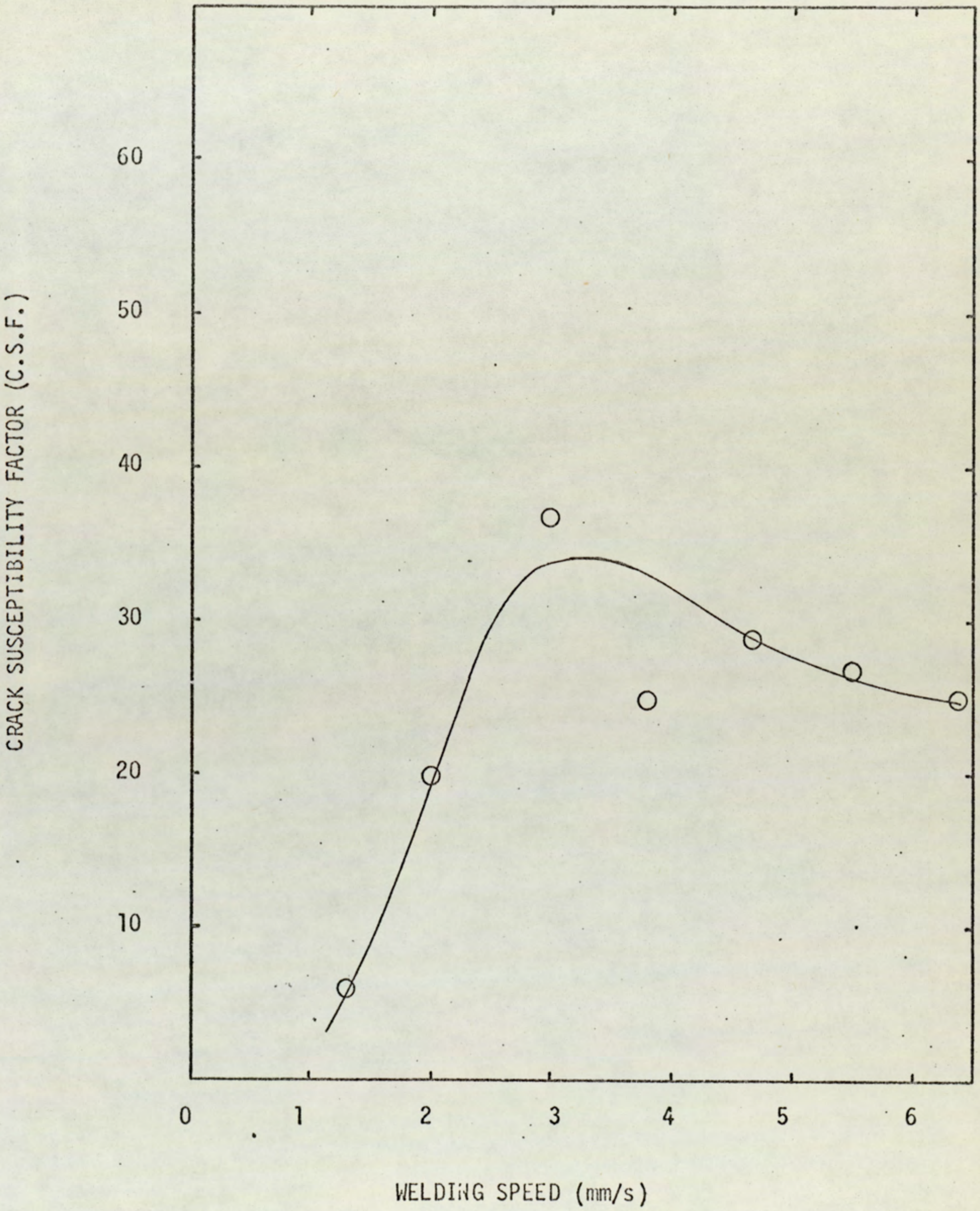


Fig. 34. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

SAE 4130 Steel SC

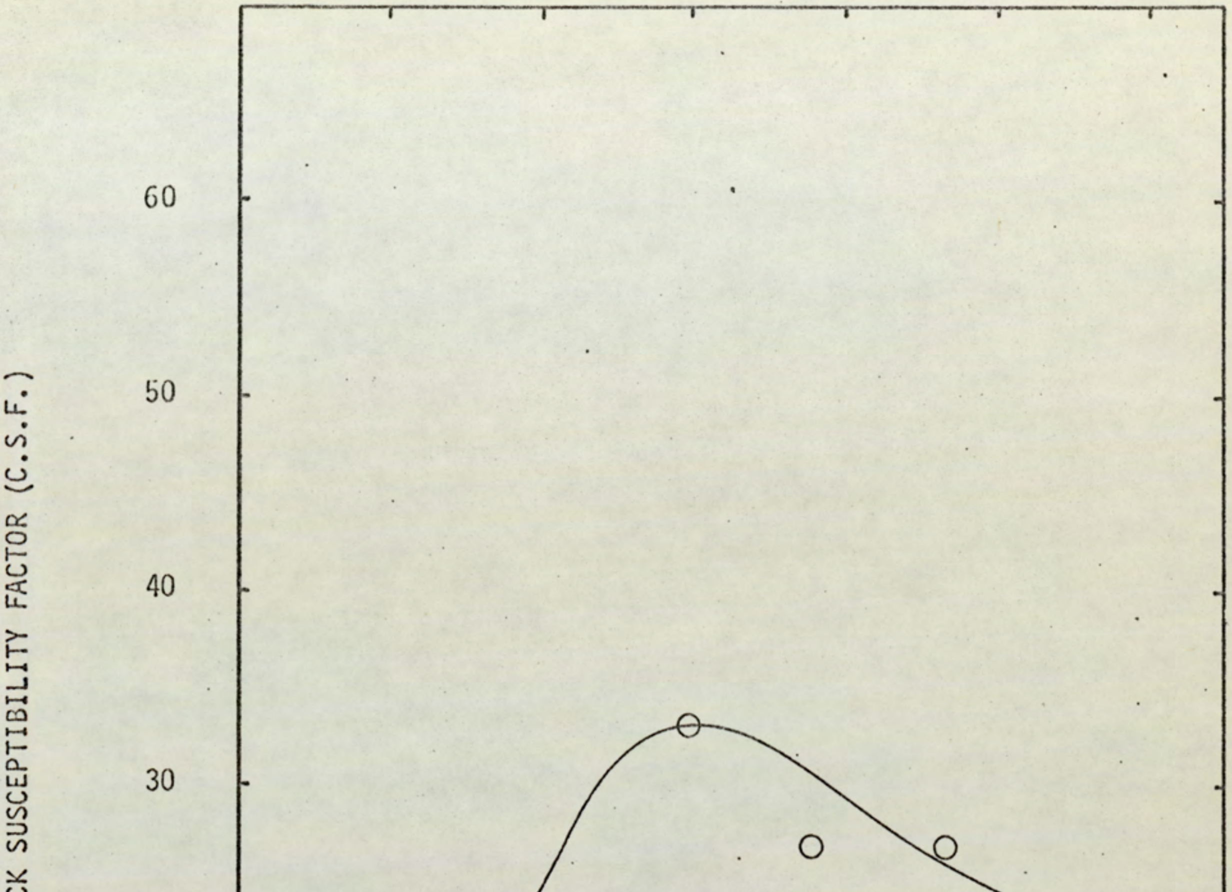


Fig. 35. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

SAE 4130 Steel SD

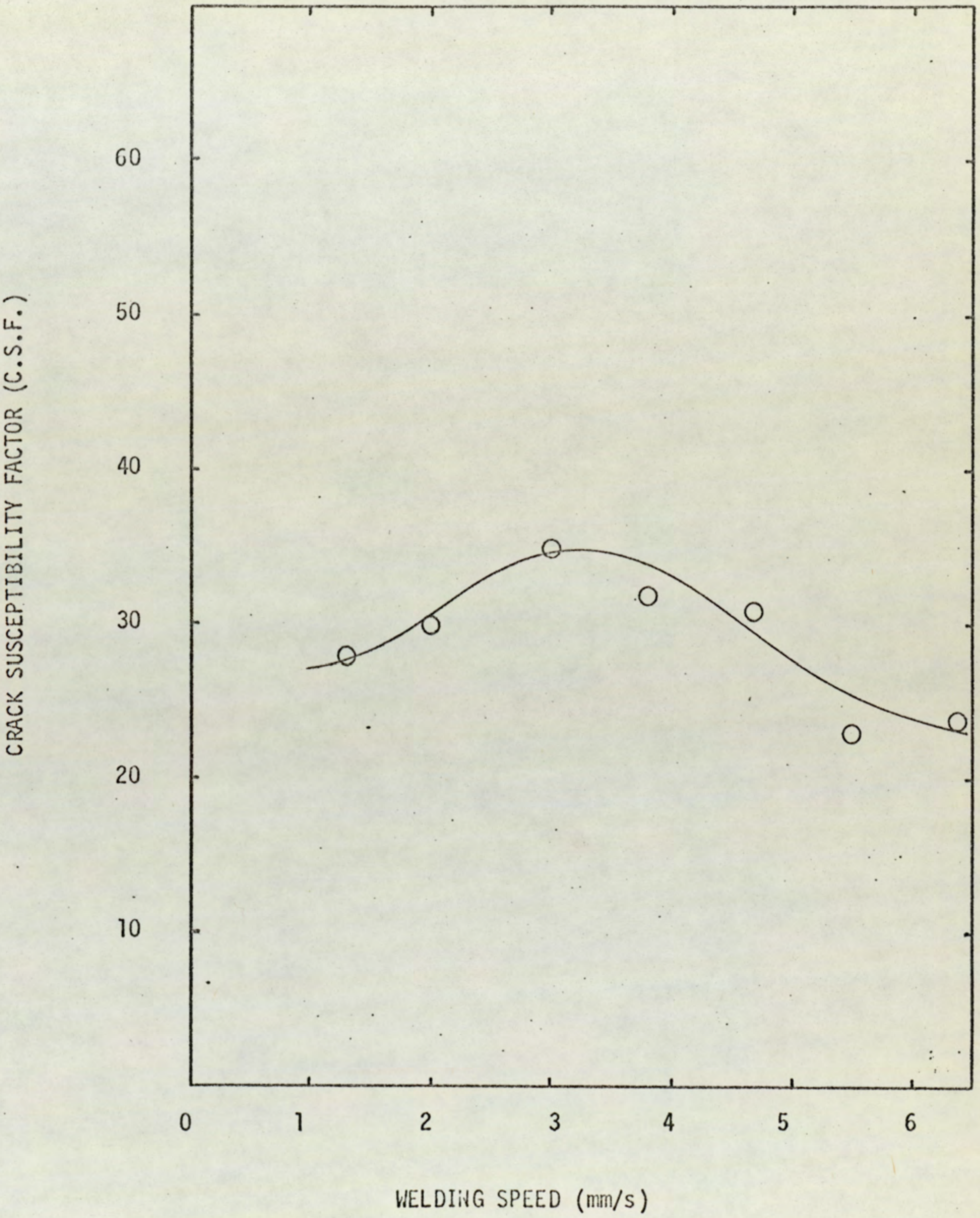


Fig. 36. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test

SAE4130 Steel SE

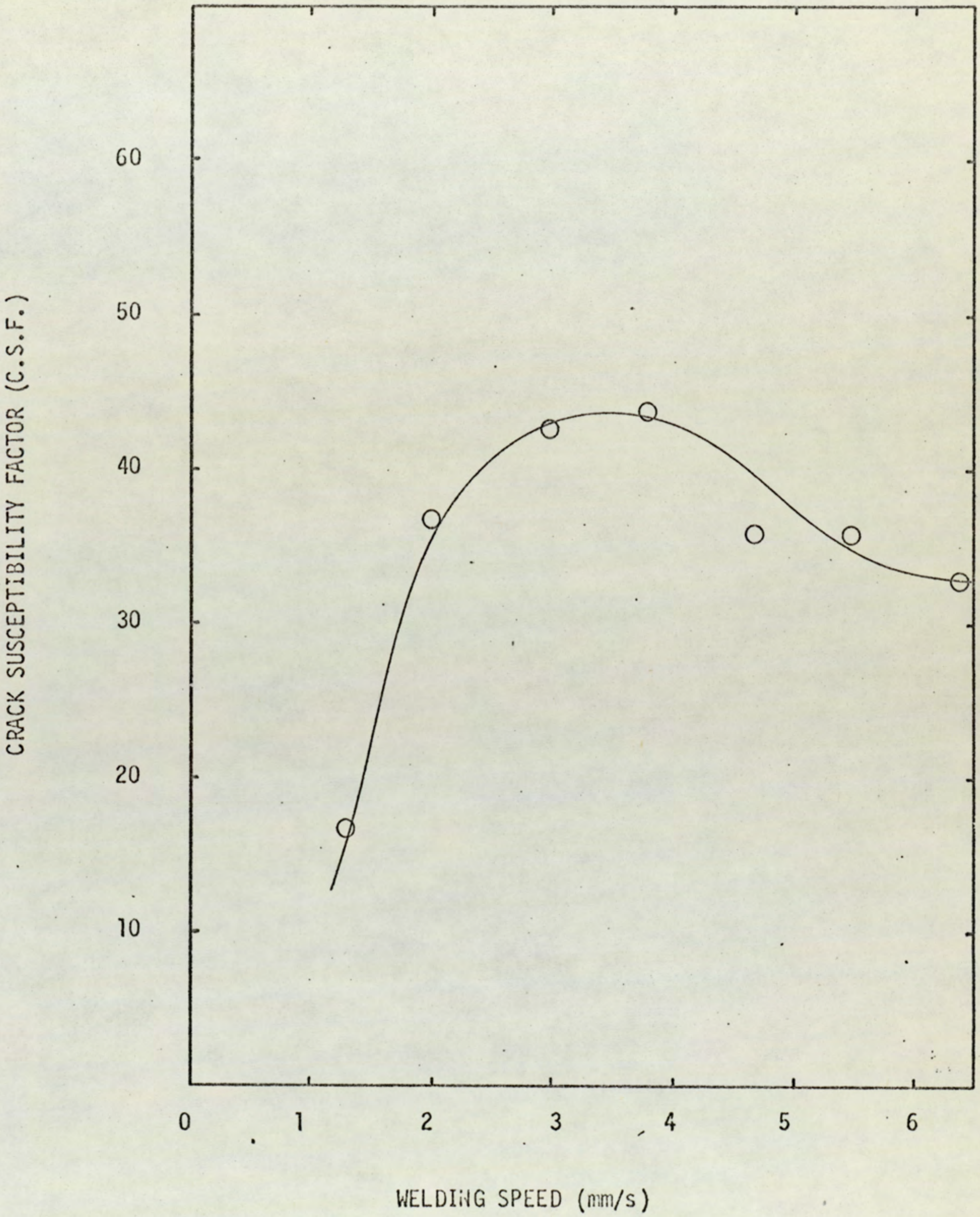


Fig. 37. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

SAE4130 Steel SA4

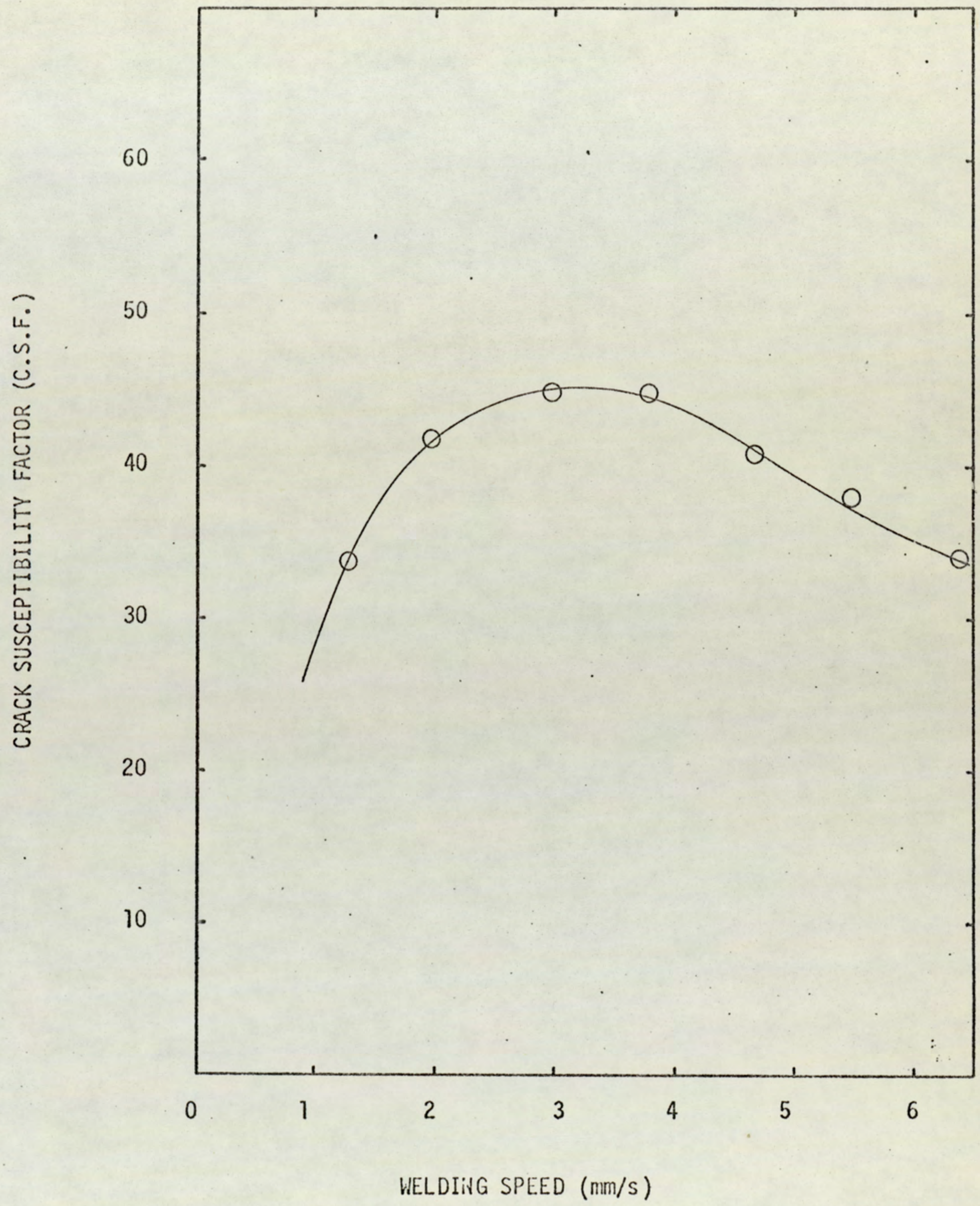


Fig. 38. The Relationship between Welding Speed and Crack Susceptibility as assessed by the Huxley Test.

SAE4130 Steel SA5

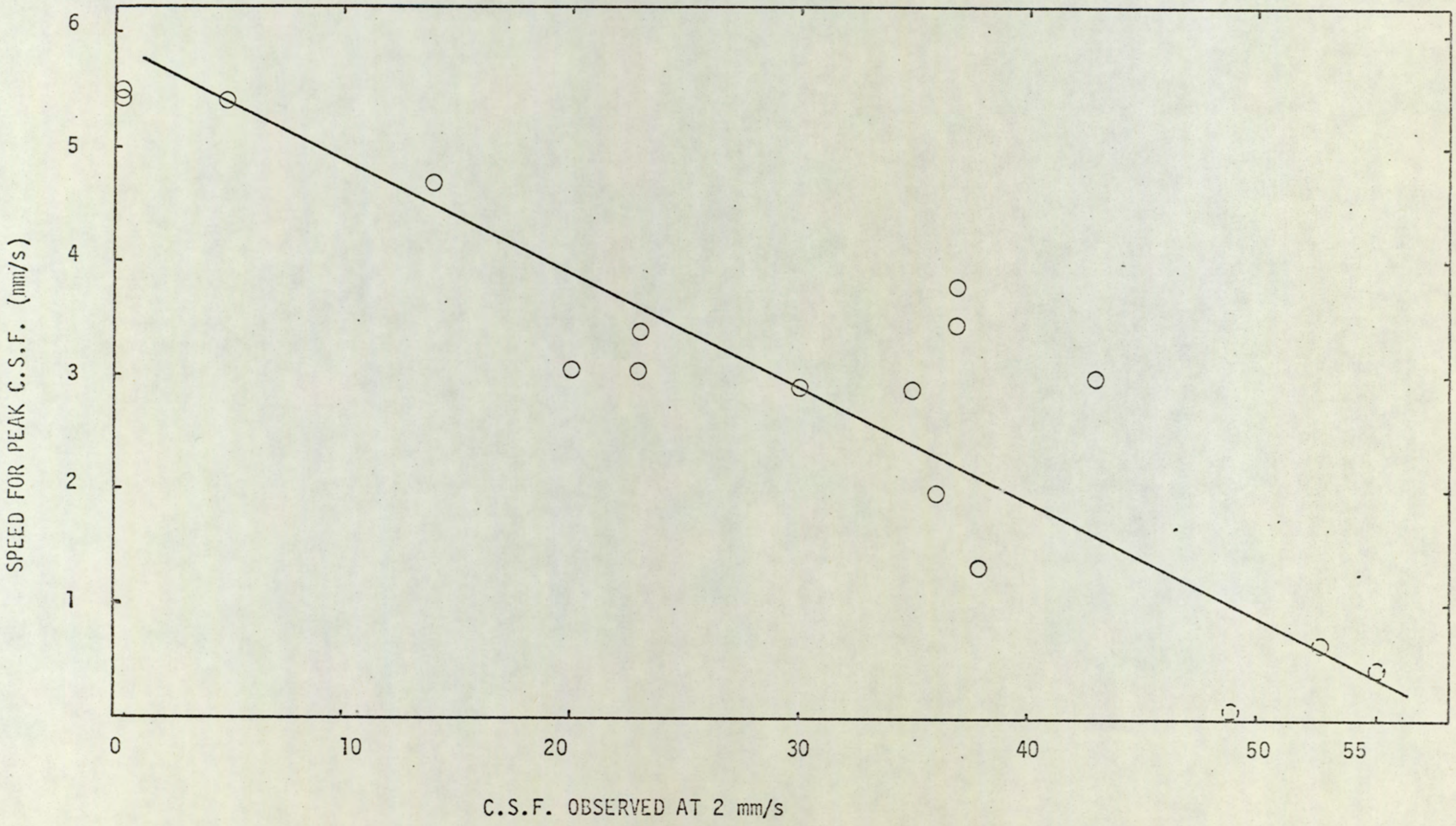


Fig. 39. Speed for peak C.S.F. versus C.S.F. observed at 2 mm/s in experimental steels.

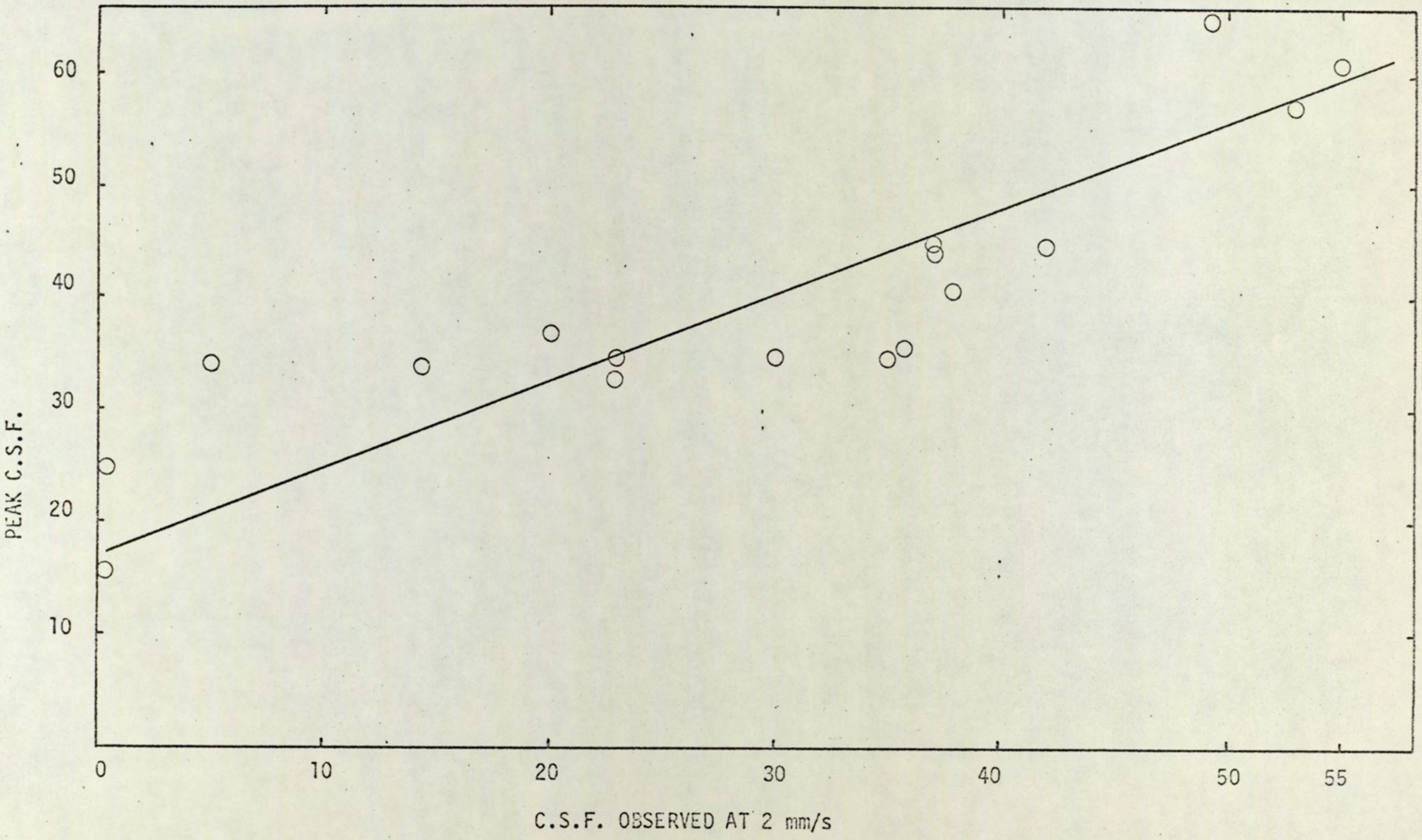


Fig. 40. Peak C.S.F. versus C.S.F. observed at 2 mm/s in experimental steels.

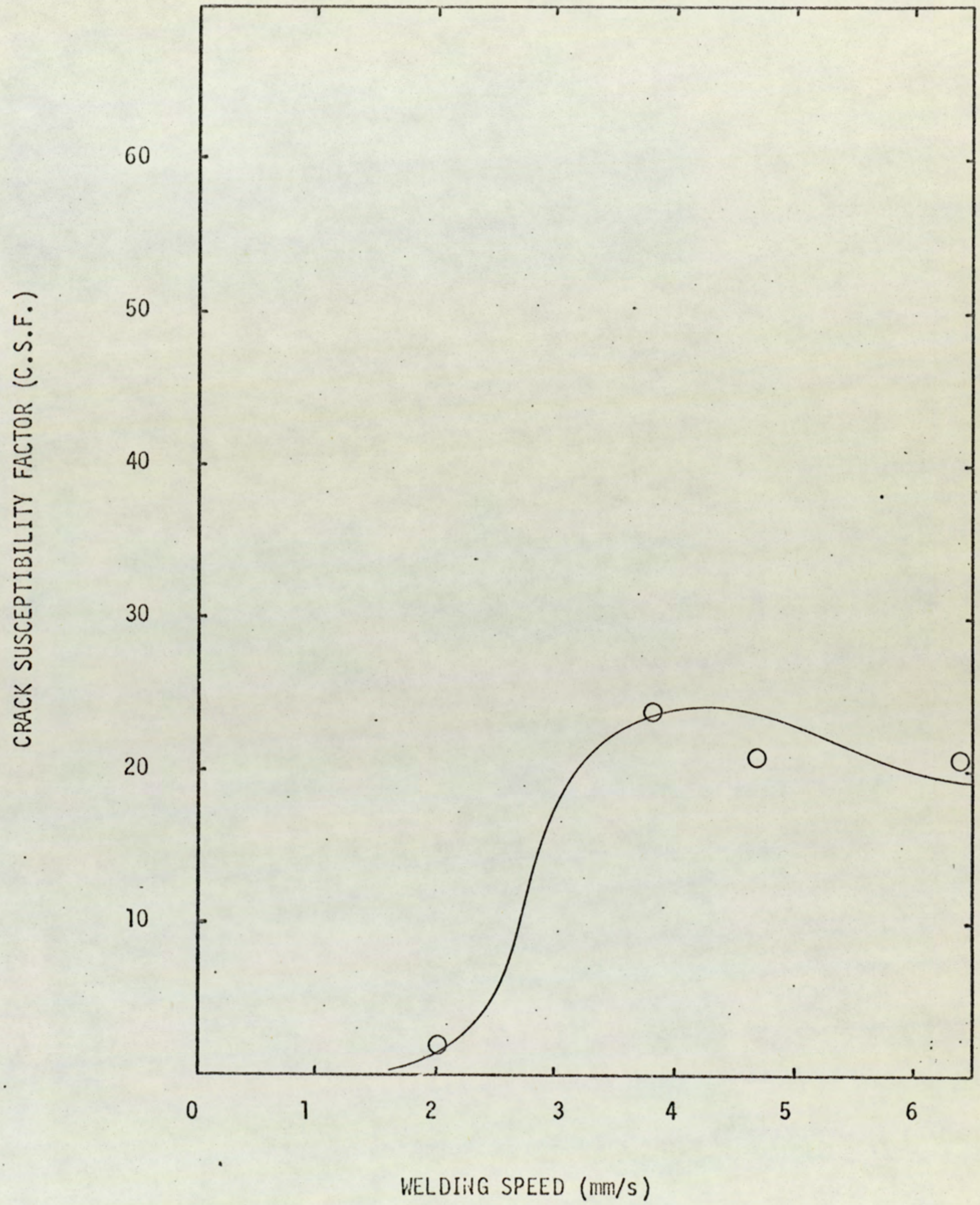


Fig. 41. The Relationship between Welding Speed and measured C.S.F. for Steel AA5, using Argon + 5% Oxygen Shielding Gas.

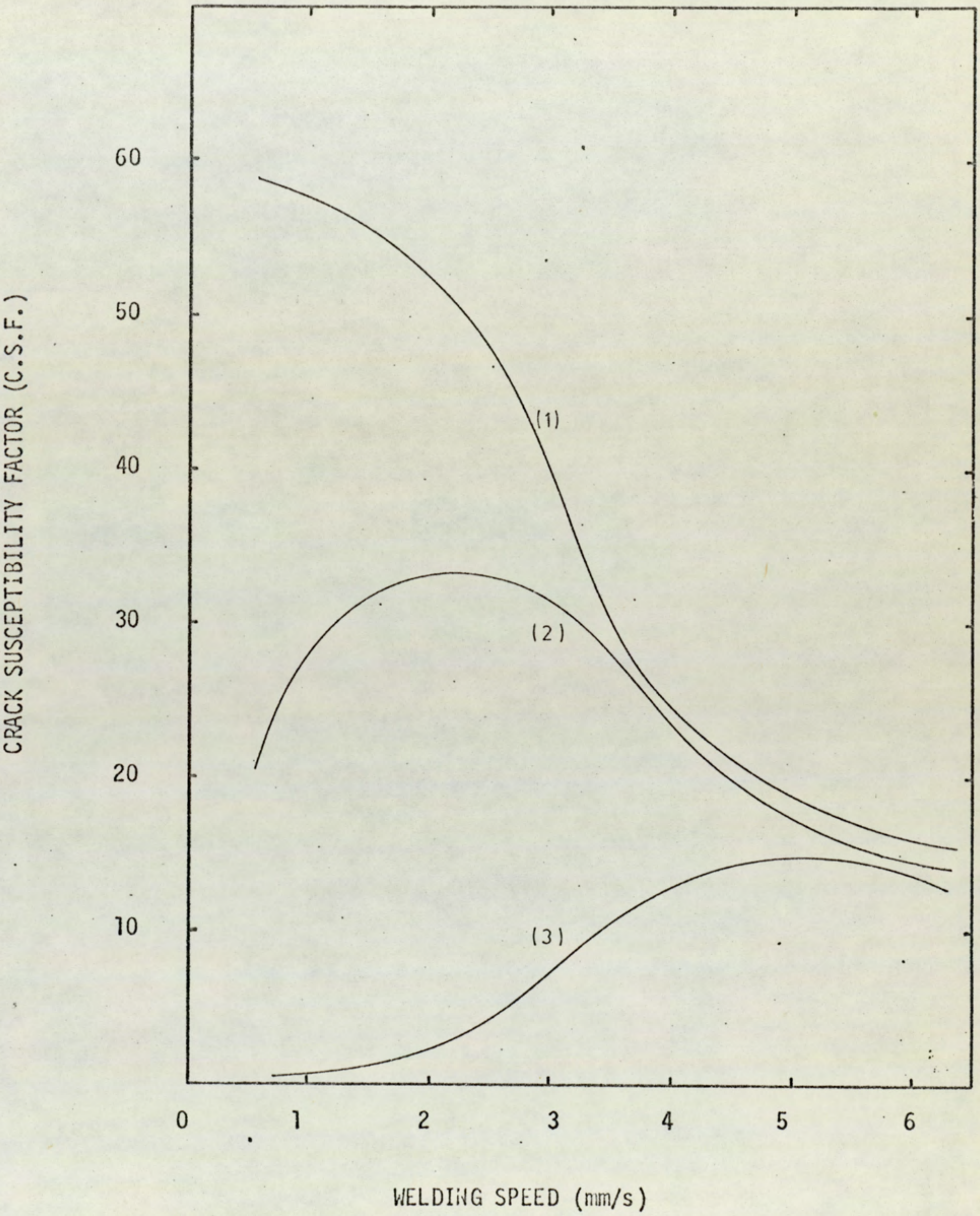


Fig.42. Typical Forms of C.S.F. - Speed Curve

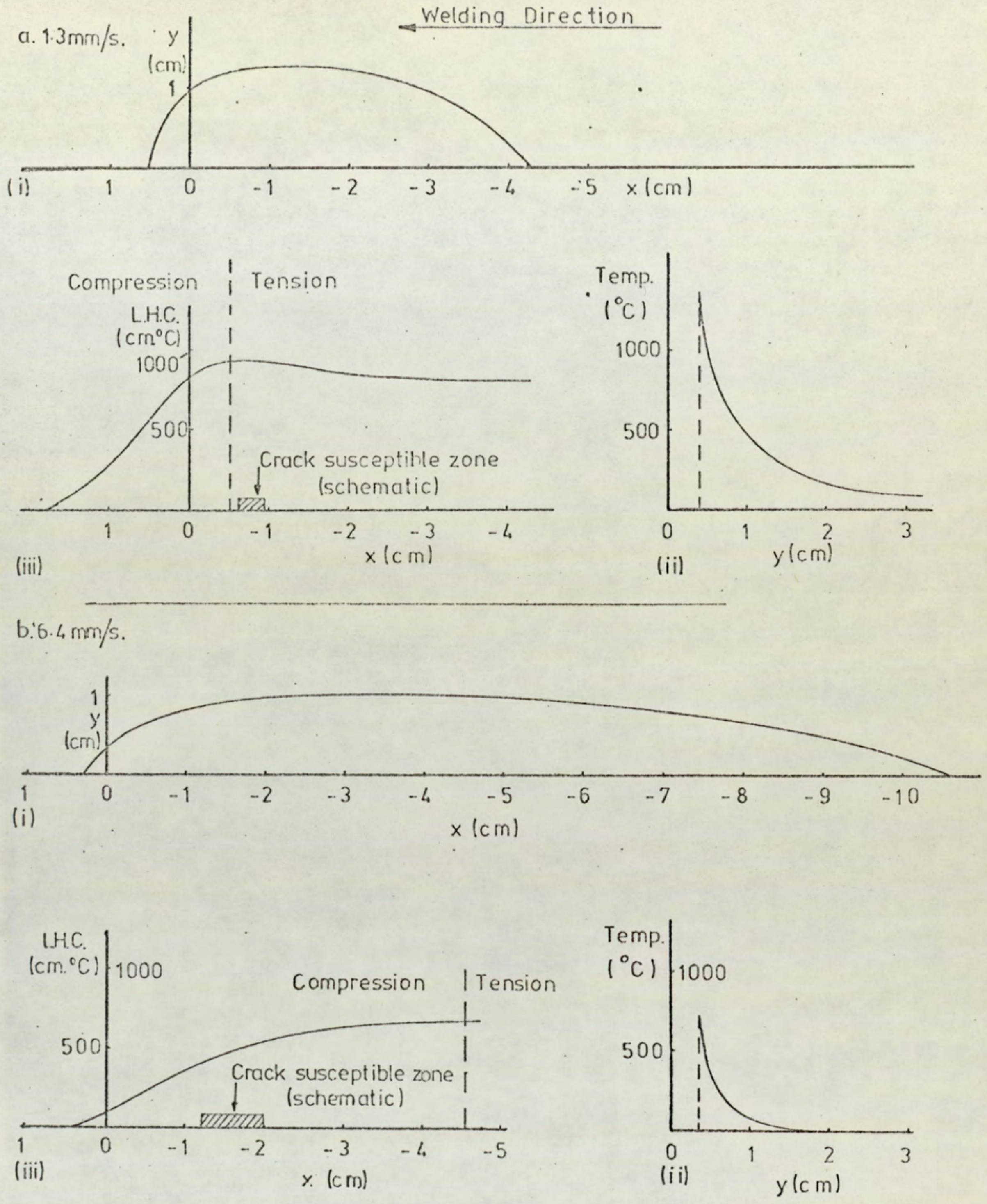


Fig. 43. The Thermal Distribution around Welds made at a) 3 ipm, 1.3 mm /s, and b) 15 ipm, 6.4 mm/s.

(i) The Positions of the 400°C Isotherms in both Welds.
x = Longitudinal Distance from Heat Source
y = Lateral Distance from Heat Source.

(ii) Lateral Temperature Distribution about the Heat Source at x = 0.

(iii) Longitudinal Distribution of Linear Heat Content (L.H.C) and the forces acting on the Centreline and Crack-Susceptible Zone.

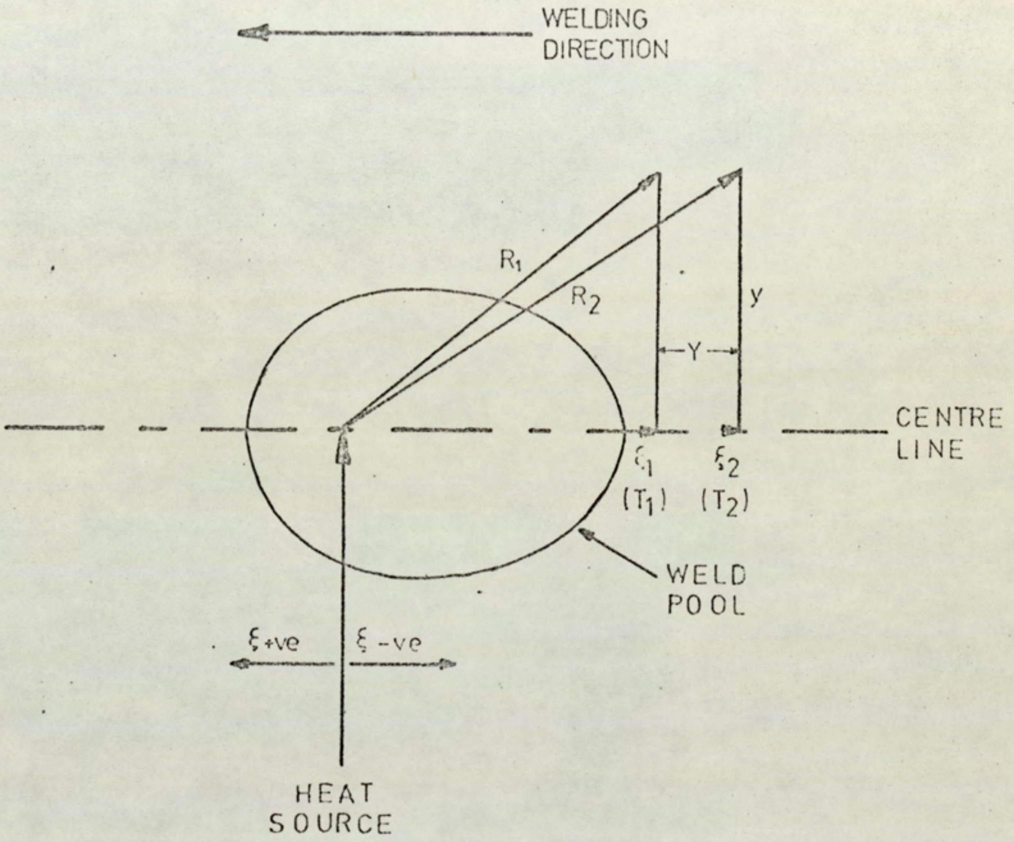


Fig. 44. The H.A.Z. Coordinates used to calculate the temperature distribution around the weld pool. (After Coleman[90], modified)

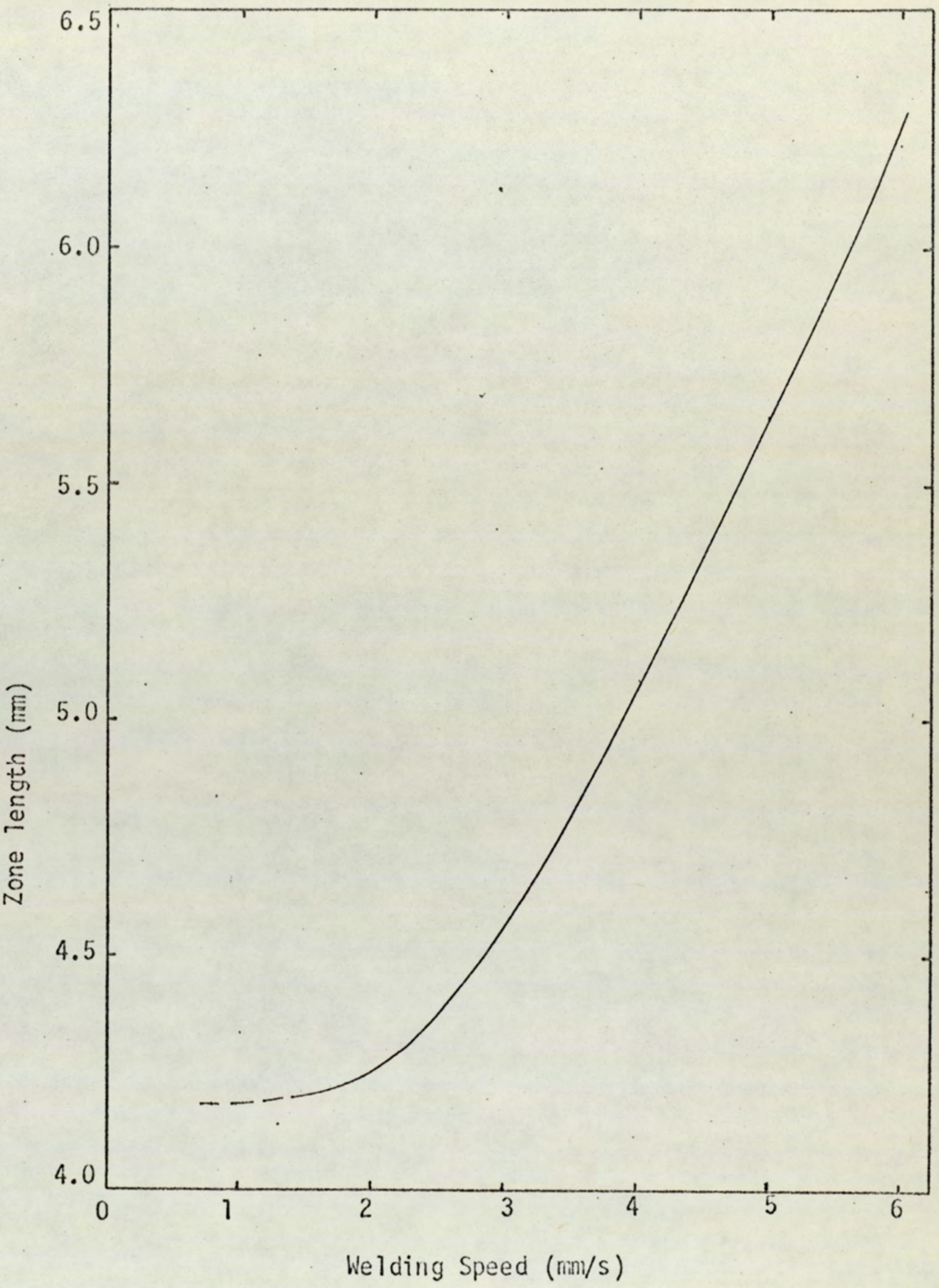


Fig. 45. Computed Solidification Zone Length as a Function of Welding Speed.

$T_1 = 1500^{\circ}C$. $T_2 = 1000^{\circ}C$ (See Fig. 44)

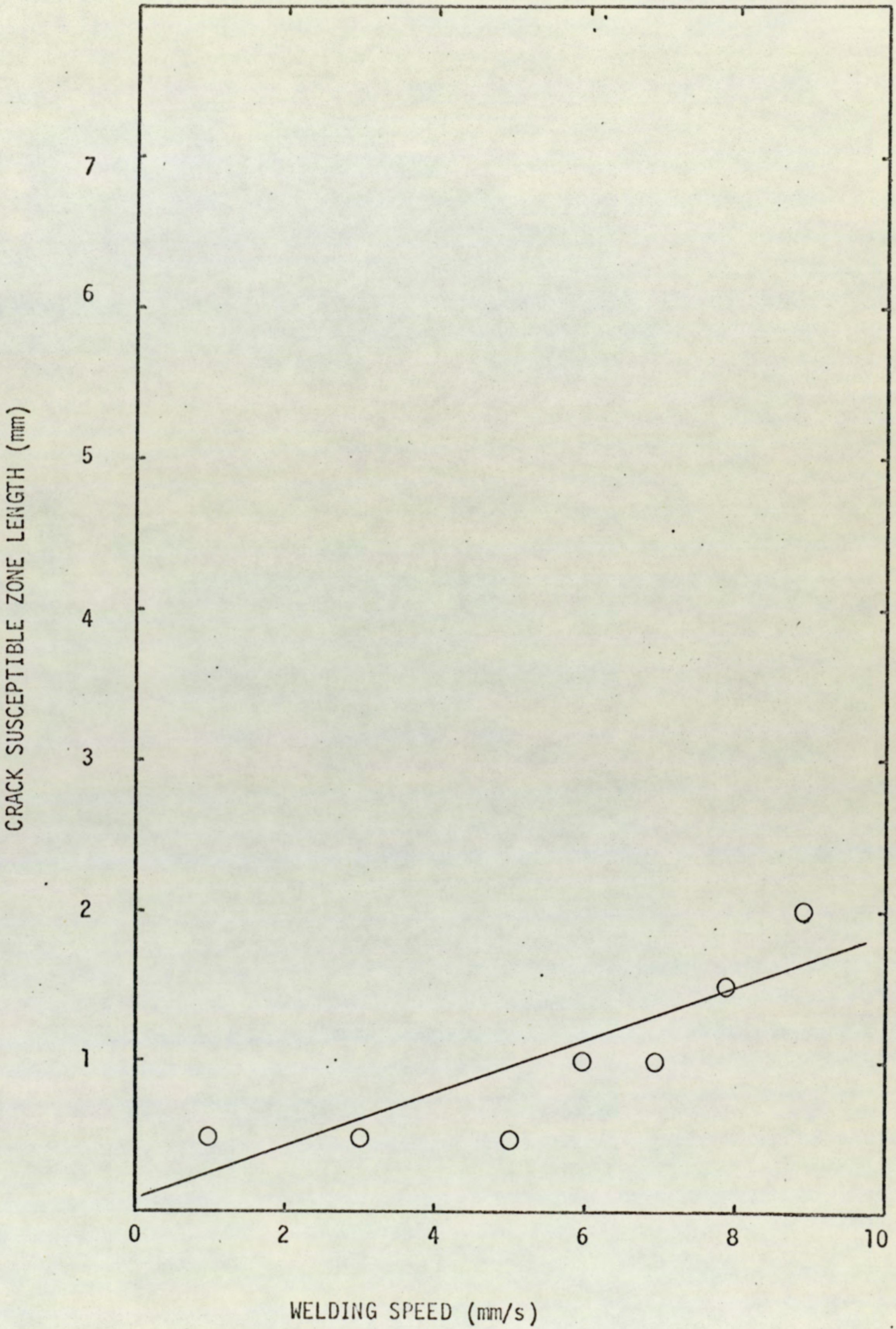


Fig. 46. The Relationship between Welding Speed and Crack Susceptible Zone Length as measured by the Varestraint Test.

En24 Steel EB

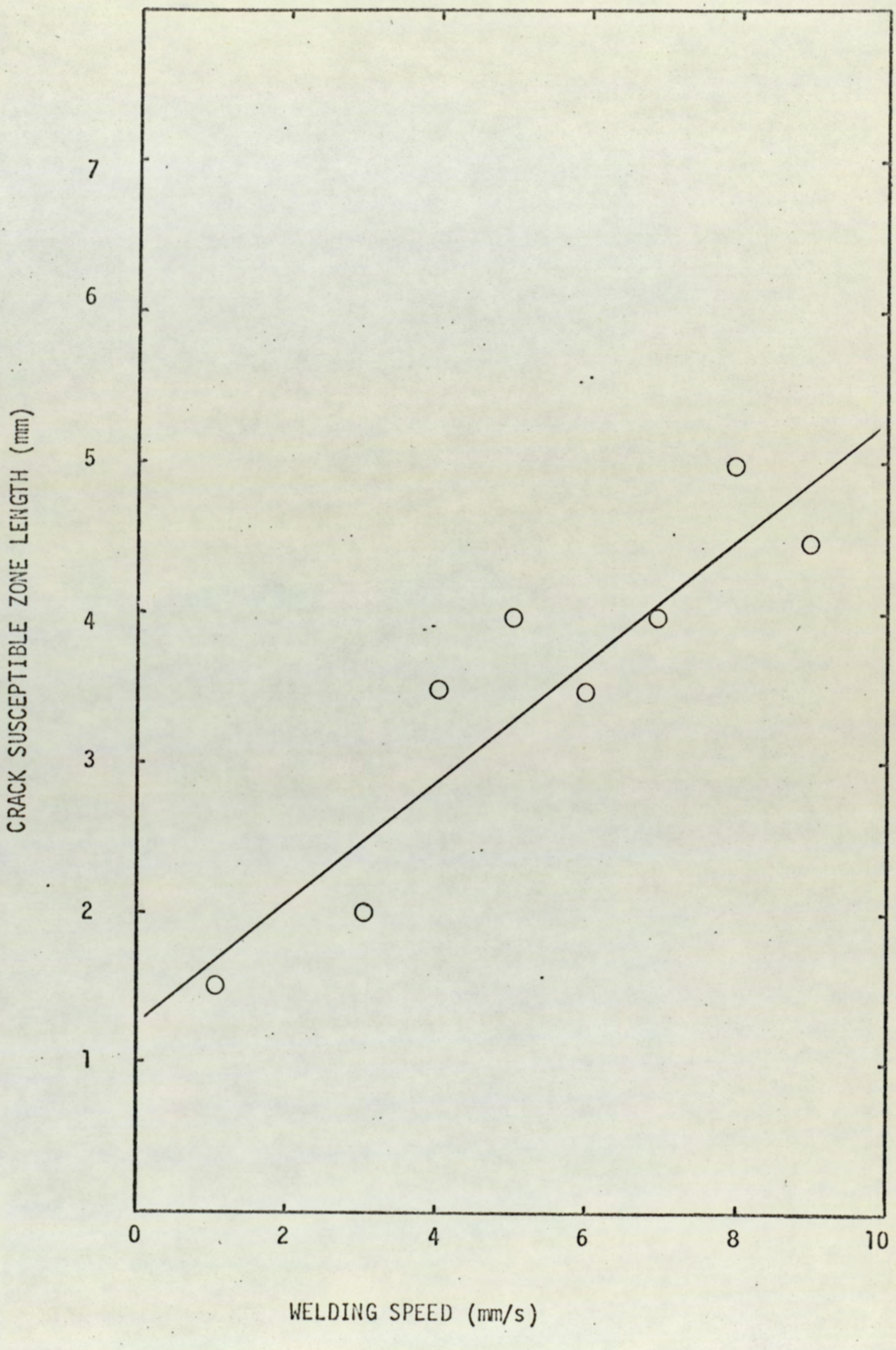


Fig. 47. The Relationship between Welding Speed and Crack Susceptible Zone Length as measured by the Varestraint Test.

En24 Steel EC

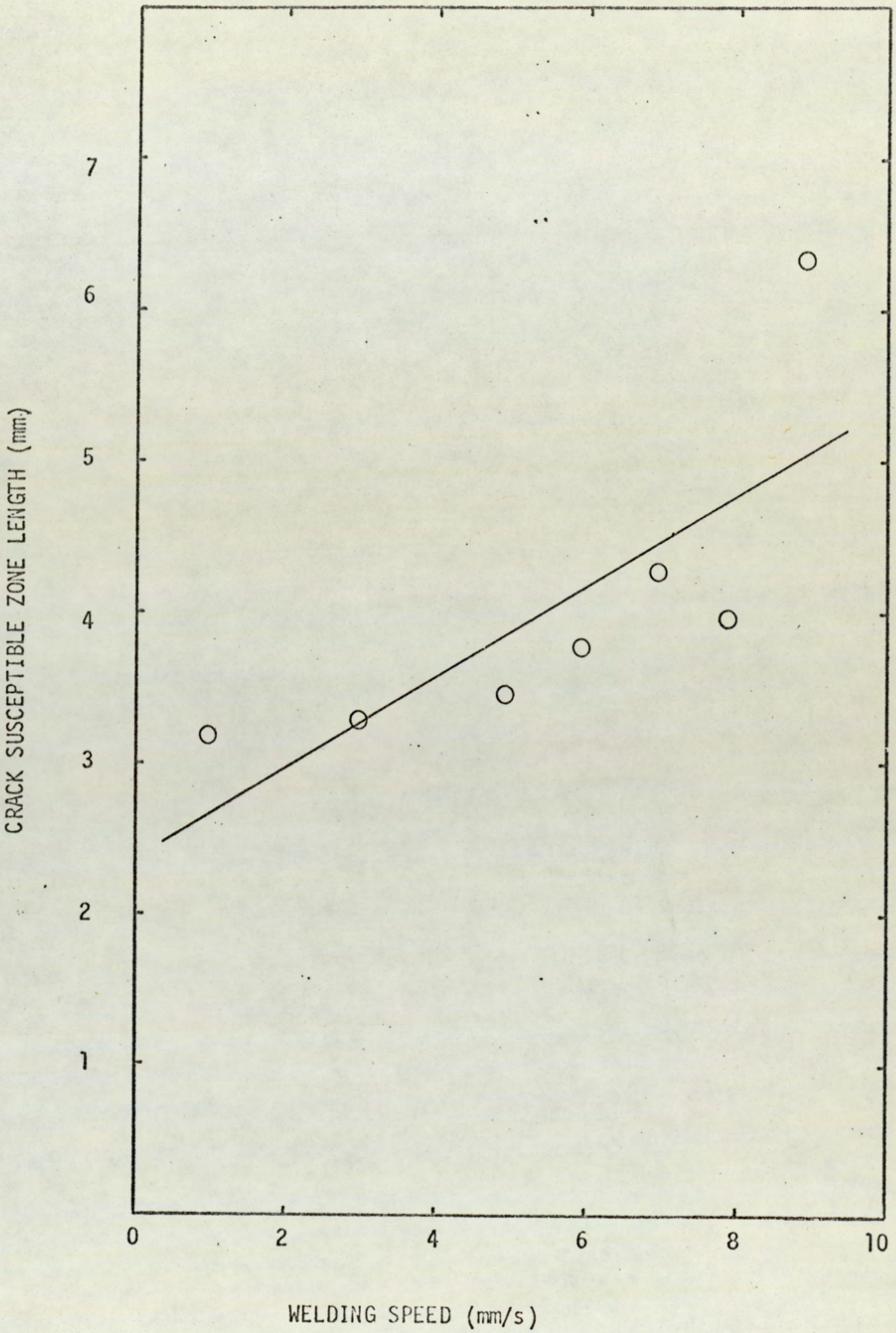


Fig. 48. The Relationship between Welding Speed and Crack Susceptible Zone Length as measured by the Varestraint Test.

EN24 Steel EA5

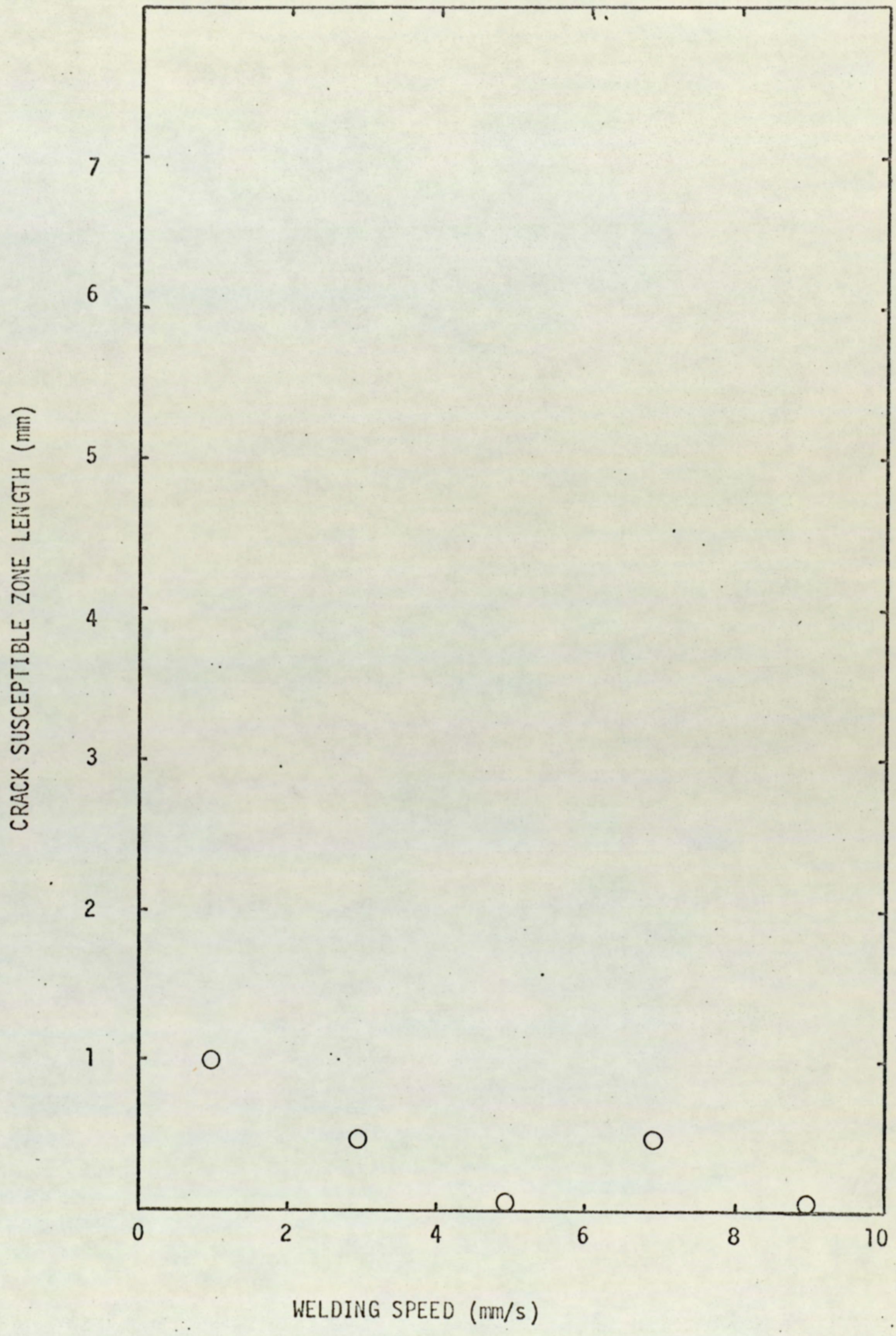


Fig. 49. The Relationship between Welding Speed and Crack Susceptible Zone Length as measured by the Varestraint Test.

SAE4130 Steel SB (Relationship not detectable)

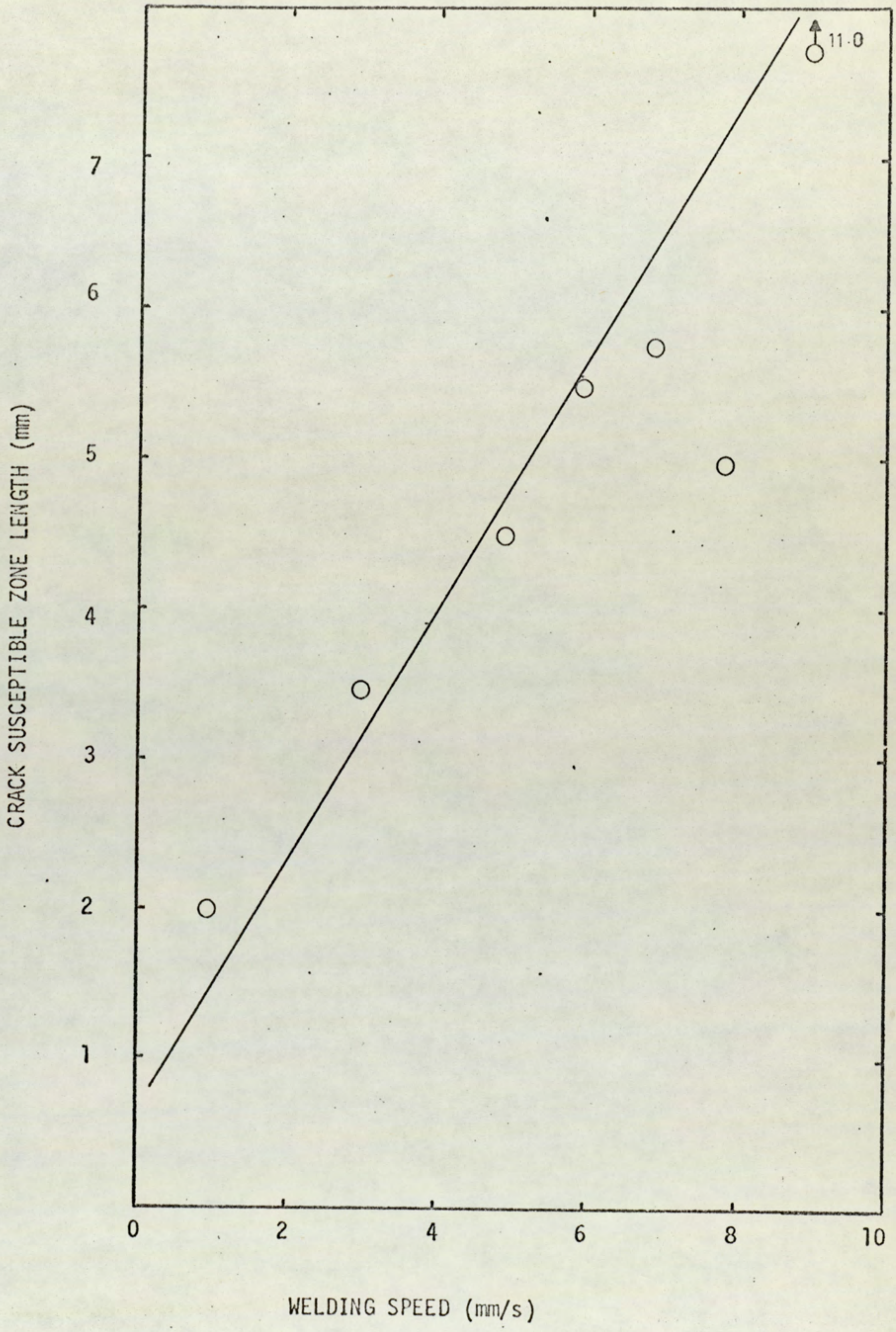


Fig. 50. The Relationship between Welding Speed and Crack Susceptible Zone Length as measured by the Varestraint Test.

SAE4130 Steel SC

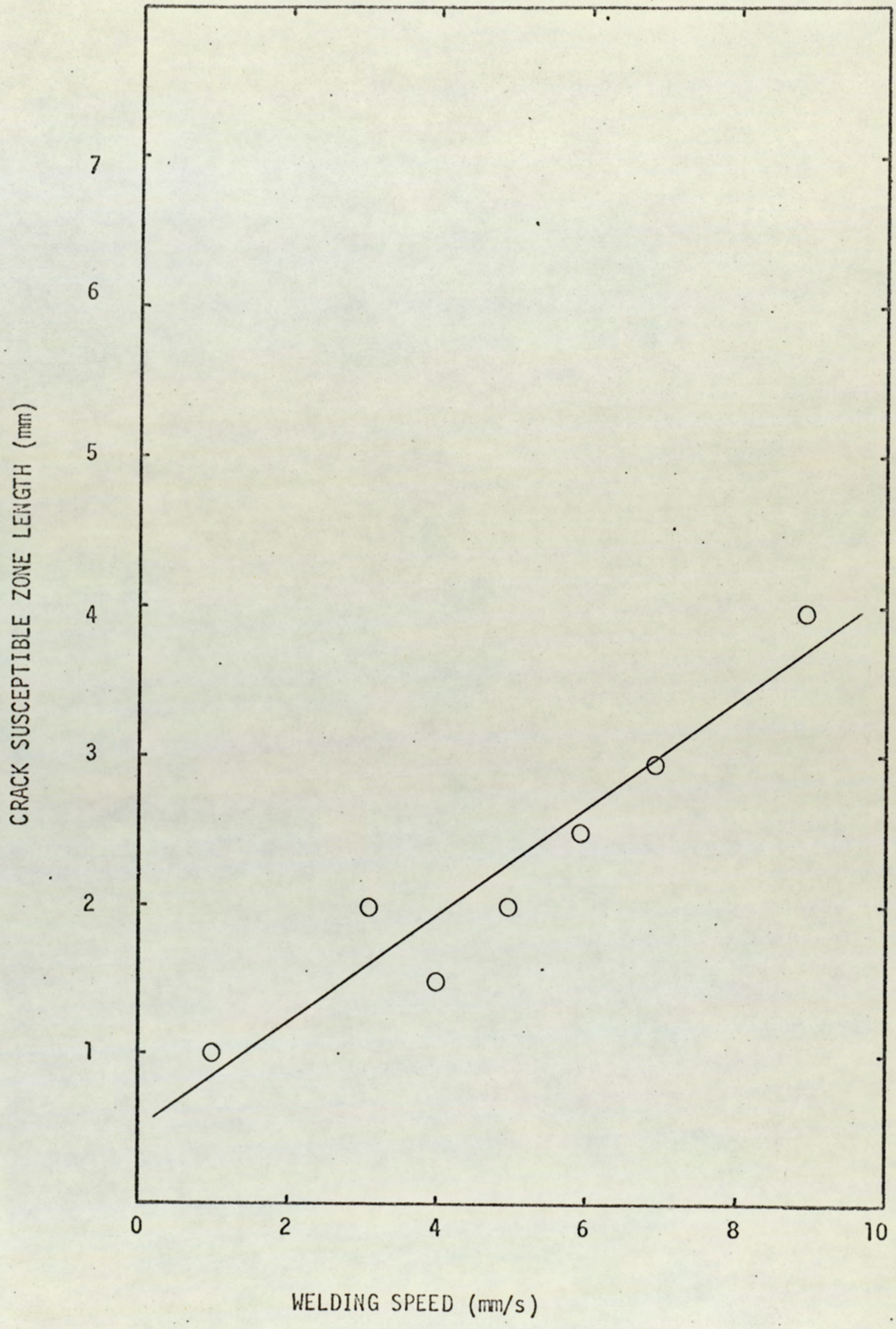


Fig. 51. The Relationship between Welding Speed and Crack Susceptible Zone Length as measured by the Varestraint Test.

SAE4130 Steel SD

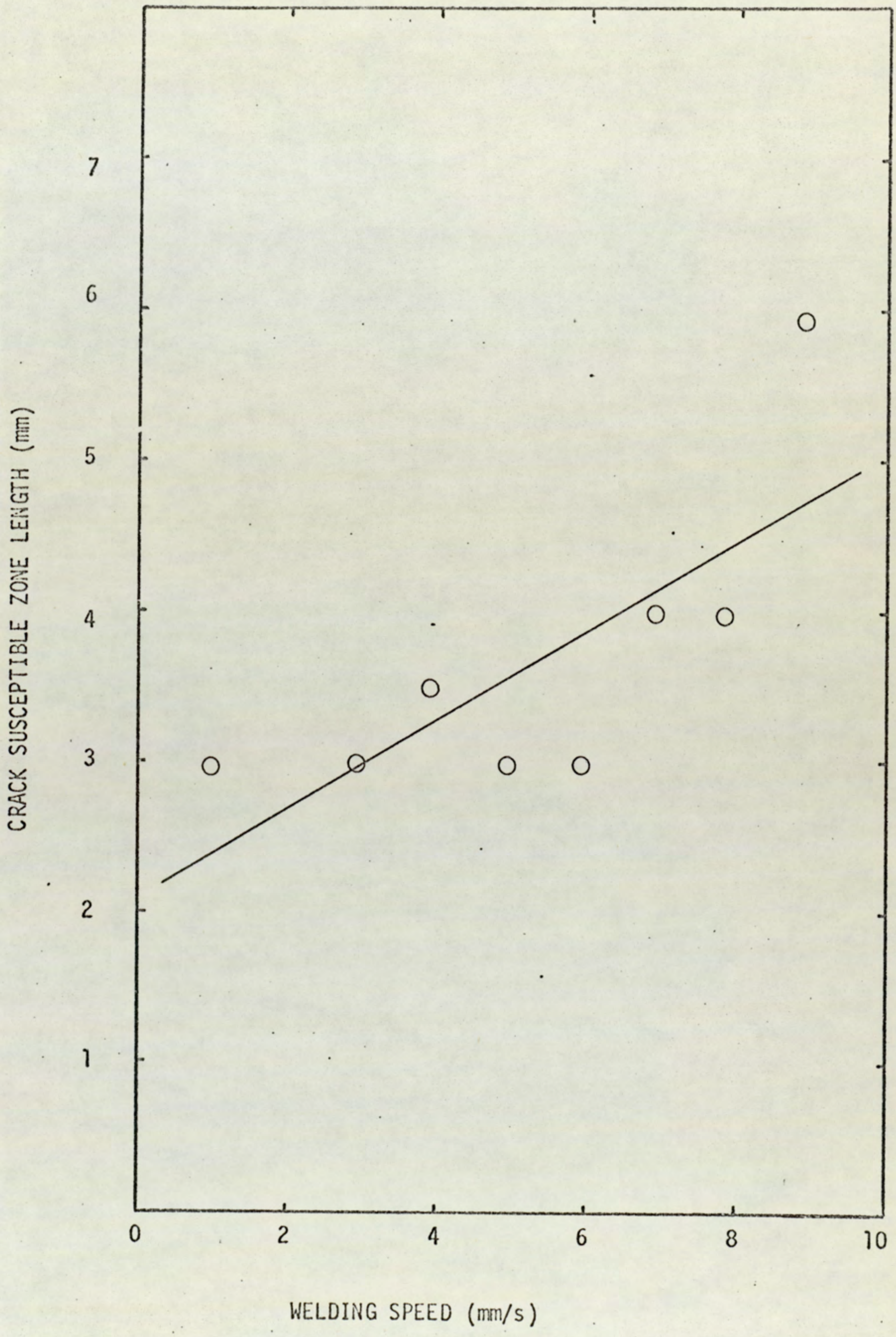


Fig. 52: The Relationship between Welding Speed and Crack Susceptible Zone Length as measured by the Varestraint Test.

SAE4130 Steel SE

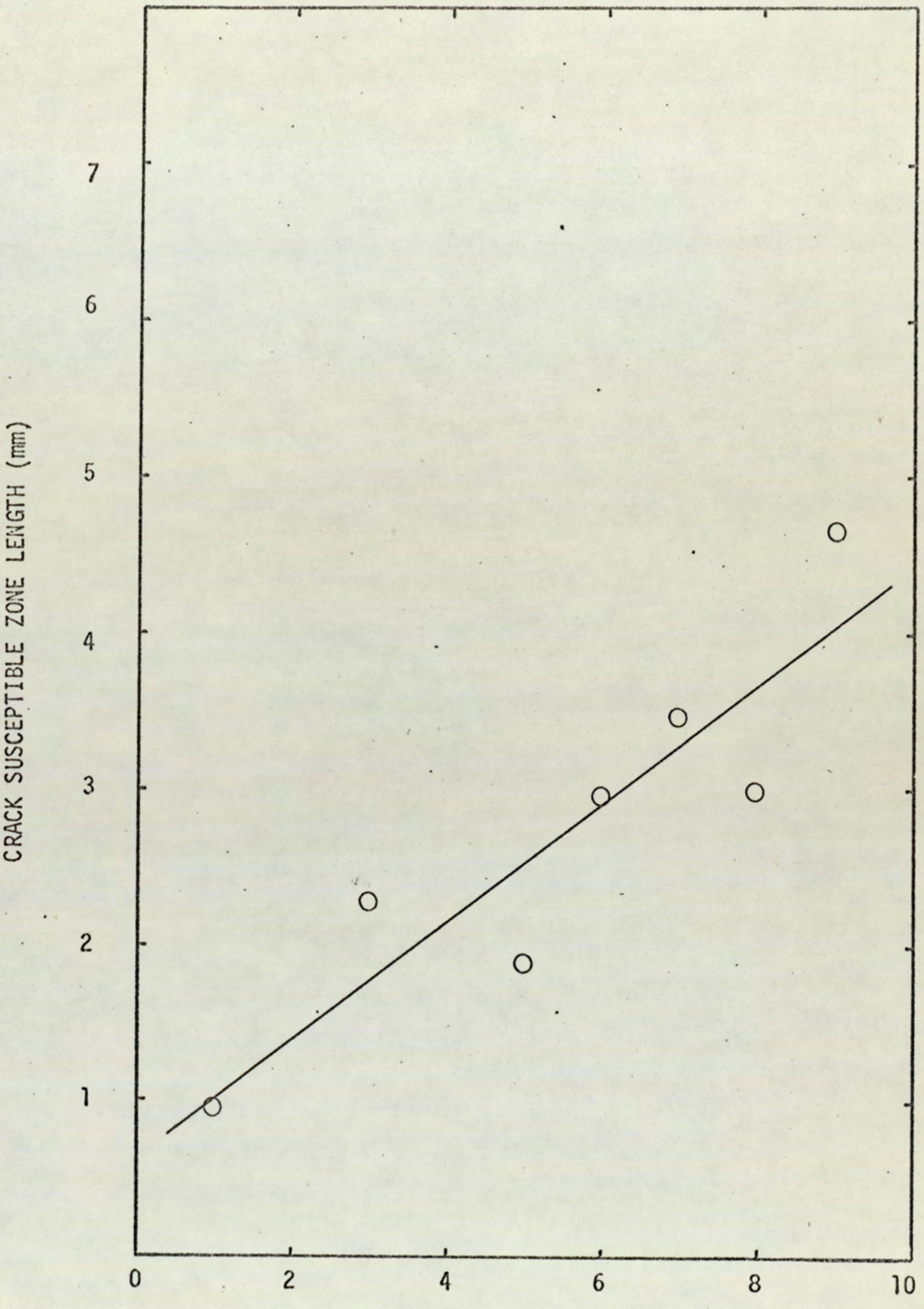


Fig. 53. The Relationship between Welding Speed and Crack Susceptible Zone Length as measured by the Varestraint Test.

SAE4130 Steel SA5

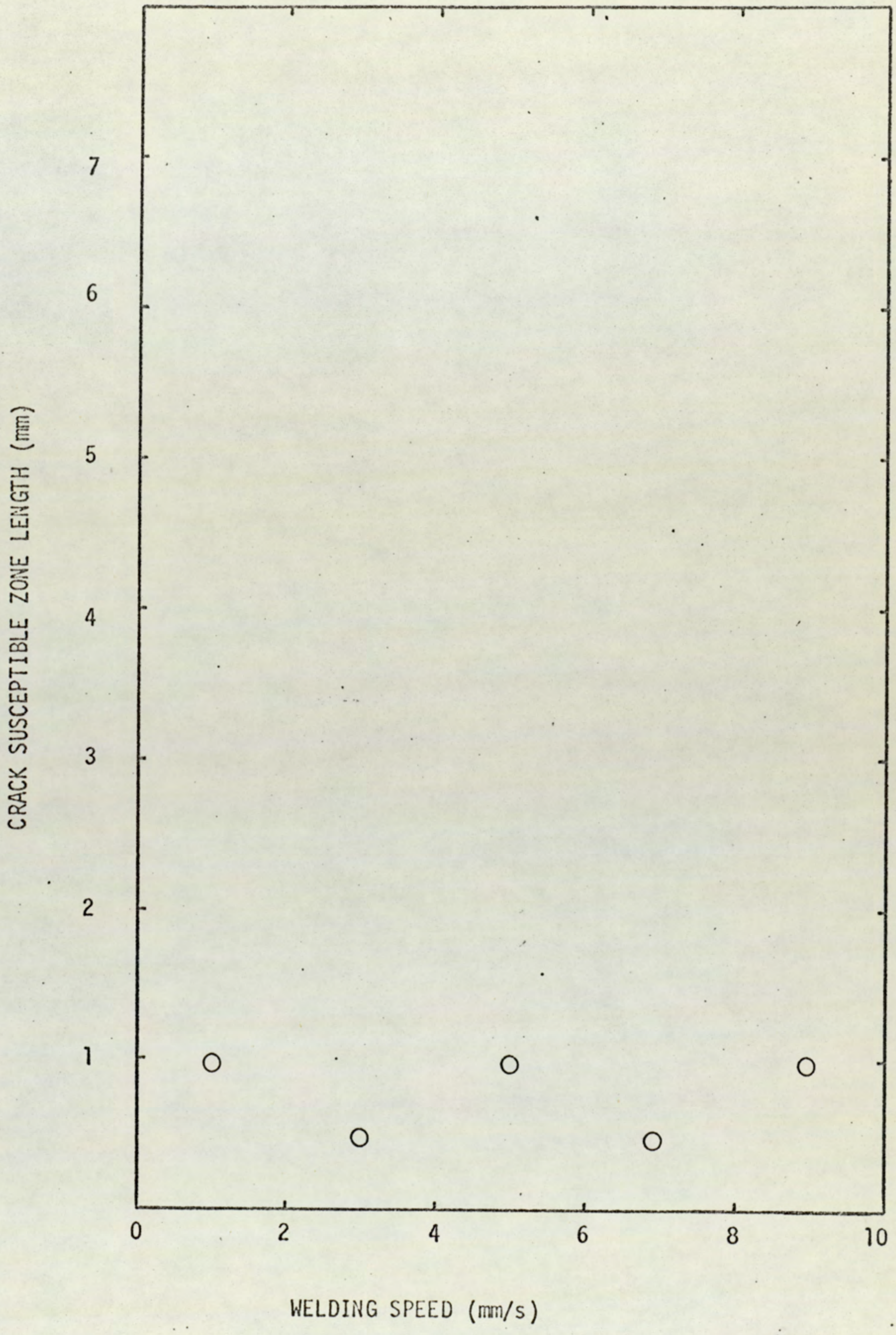


Fig. 54. The Relationship between Welding Speed and Crack Susceptible Zone Length as measured by the Varestraint Test.

ASTM A387B Steel AD (Relationship not detectable)

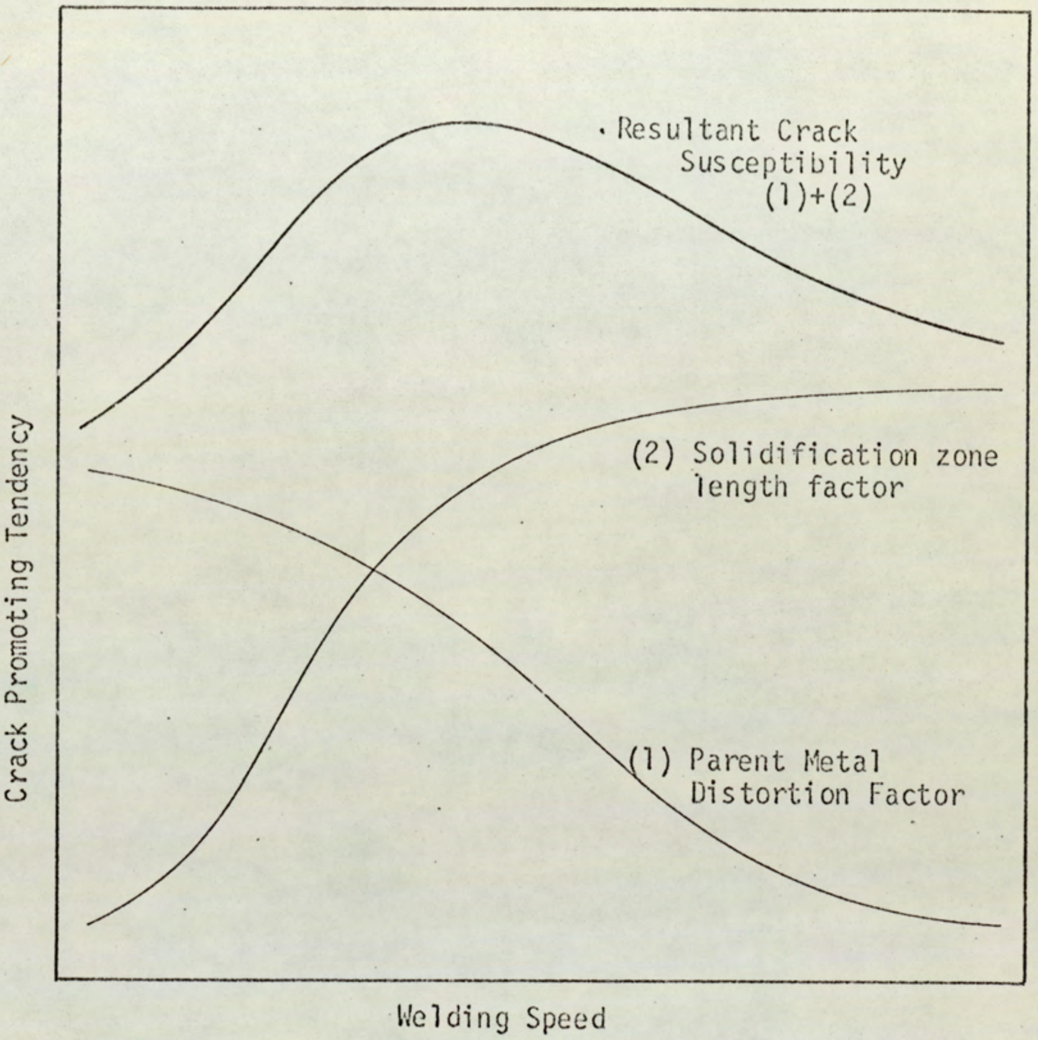


Fig. 55. Schematic Representation of the Separate Influences of Welding Speed on Crack Susceptibility.

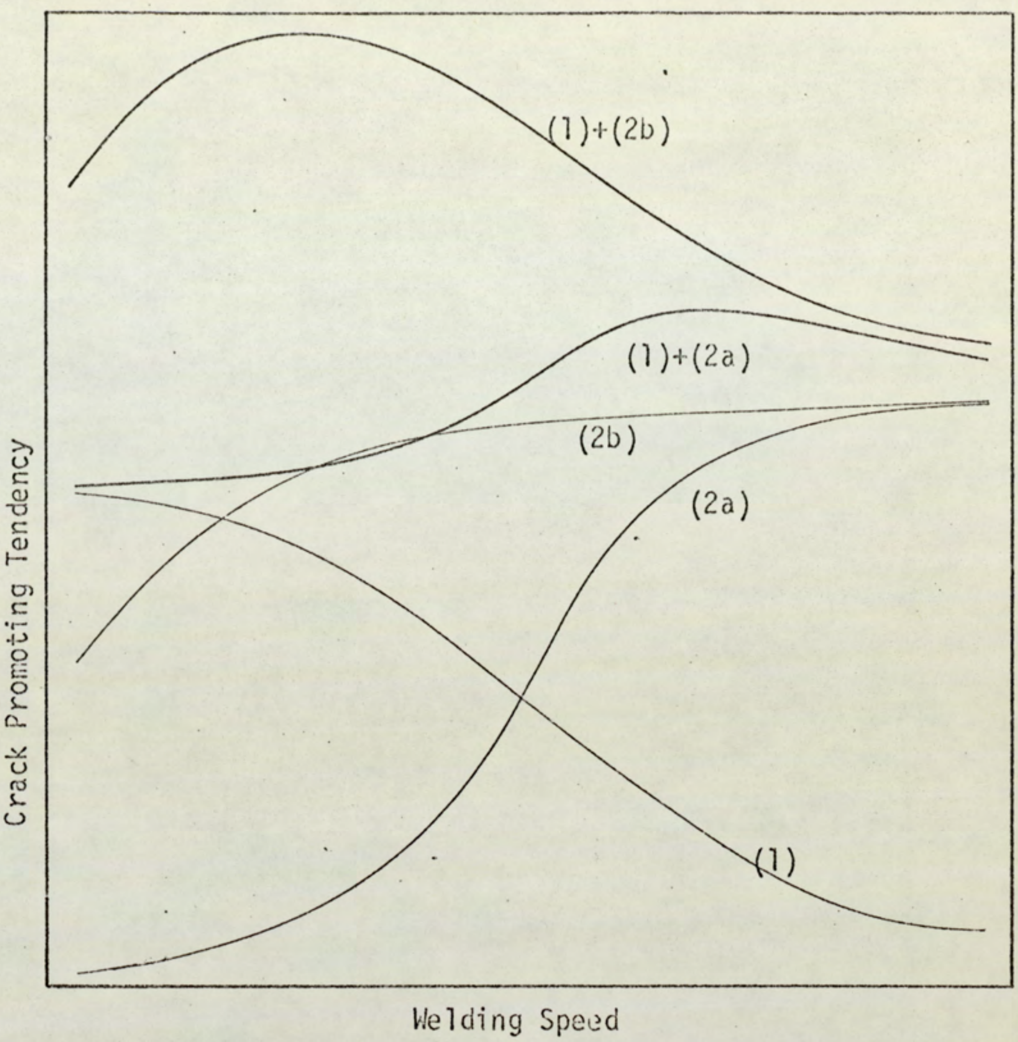


Fig. 56. Schematic Representation of the Influence of Composition on the Cracking - Speed Relationship.

(1)	Parent Metal Distortion Factor.
(2a)	Solidification Zone Length Factor (Short zone, Steel of "Low Crack Susceptibility")
(2b)	Solidification Zone Length Factor (Long zone, Steel of "High Crack Susceptibility")
(1)+(2a)	Resultant Crack Susceptibility (Low Susceptibility Steel)
(1)+(2b)	Resultant Crack Susceptibility (High Susceptibility Steel)

Fig. 57. Weld metal from steel SD, showing grain boundary enrichment in phosphorus (Black).
Etch: 2% Nital and Picric Acid (Consecutive),
x 120

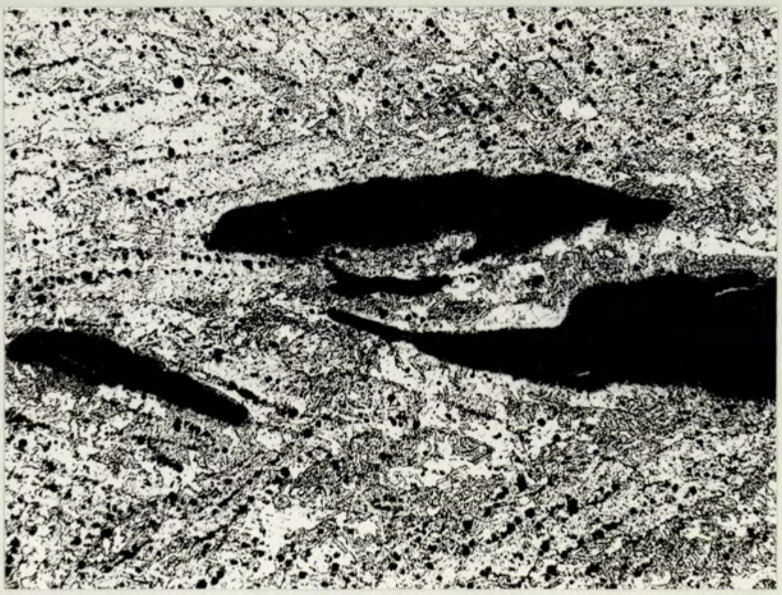


Fig. 58. Cracks following the solidification pattern, delineated by phosphorus segregation.
Etch: 2% Nital and Picric Acid (Consecutive),
x 20

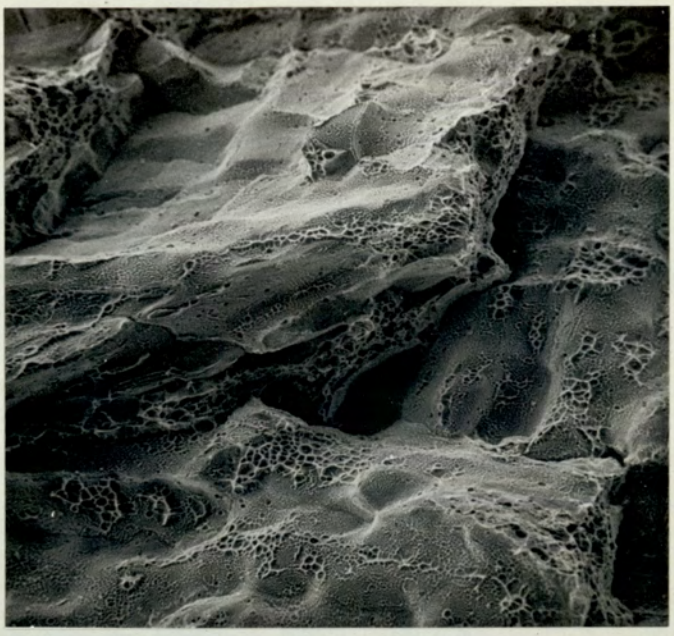


Fig. 59. Extension of solidification crack in AD weld metal, showing fine dimples in smooth faceted film - like array. Scanning Electron Micrograph, x 340

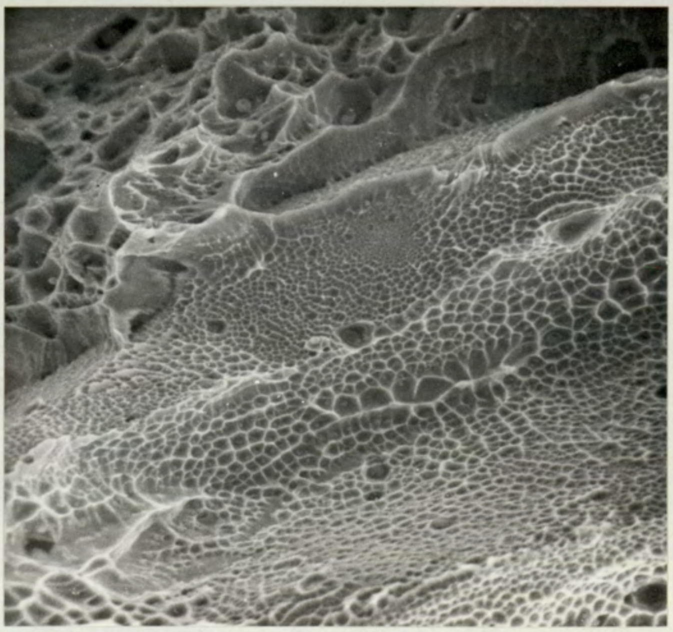


Fig. 60. Same material as Fig. 59, showing fine particles present in dimples. Scanning Electron Micrograph, x 1,700

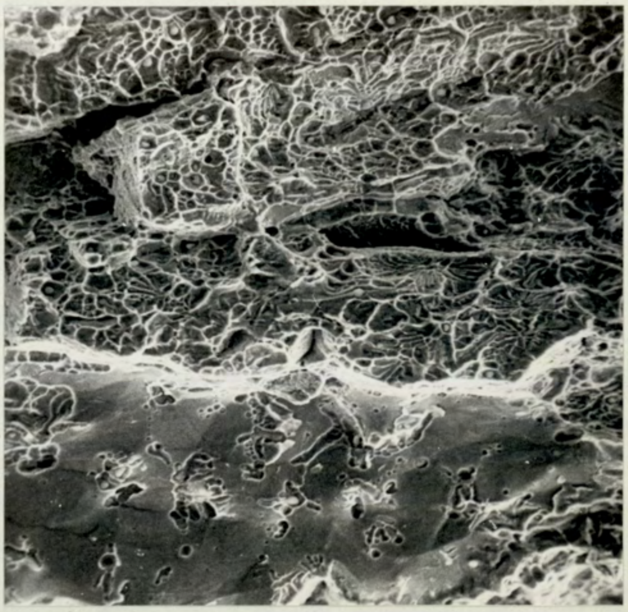


Fig. 61. Extension of solidification crack in EC weld metal, showing continuous film and palmate sulphide eutectic. Scanning Electron Micrograph x 800

Fig. 62. Extension of solidification crack in EB weld metal showing absence of film-like material. Scanning Electron Micrograph x 1,600

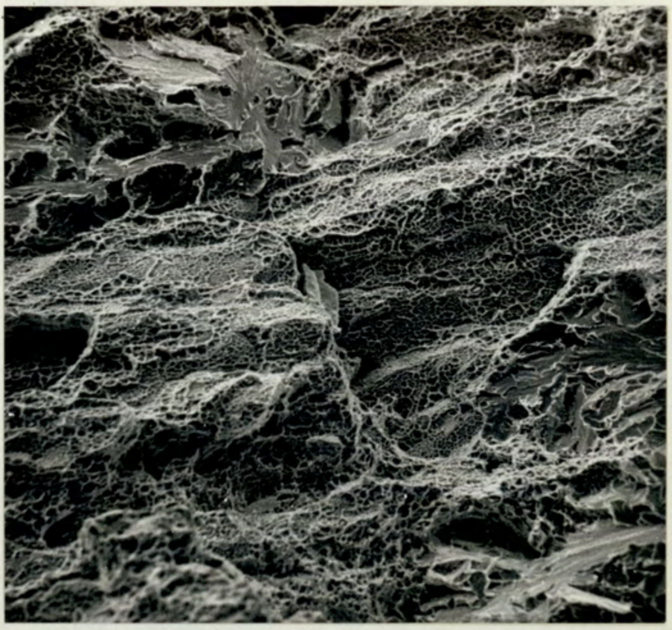


Fig. 63. Extension of solidification crack in SE weld metal showing coarser dimple array. (Compare with Fig. 59).
Scanning Electron Micrograph, x 300

Fig. 64. Same material as Fig. 63, showing lenticular and extended inclusions.
Scanning Electron Micrograph. x 3,000

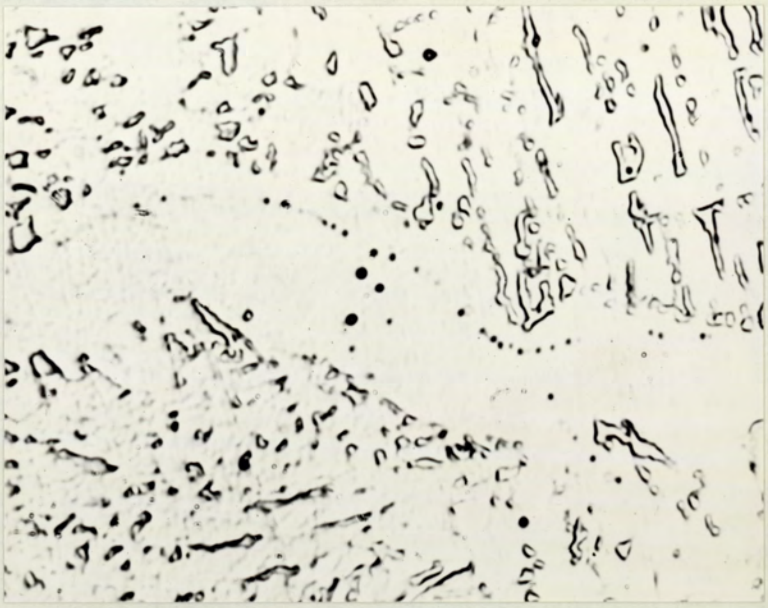


Fig. 65. Weld metal from steel AD, showing chain of fine inclusions.
Etch: 2% Nital. x 1,000

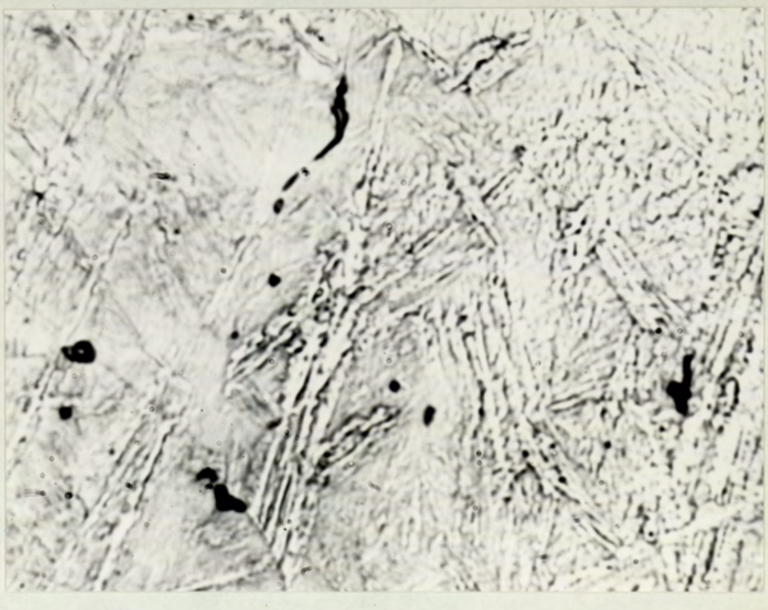


Fig. 66. Weld metal from steel EC, showing elongated inclusions.
Etch: 2% Nital. x 1,000

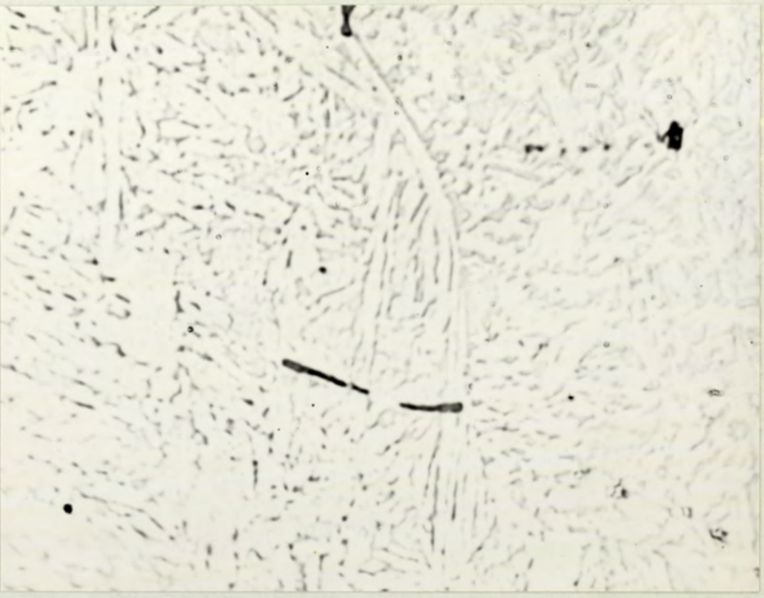


Fig. 67. Weld metal from steel SE, showing elongated inclusions.
Etch: 2% Nital. x 1,000

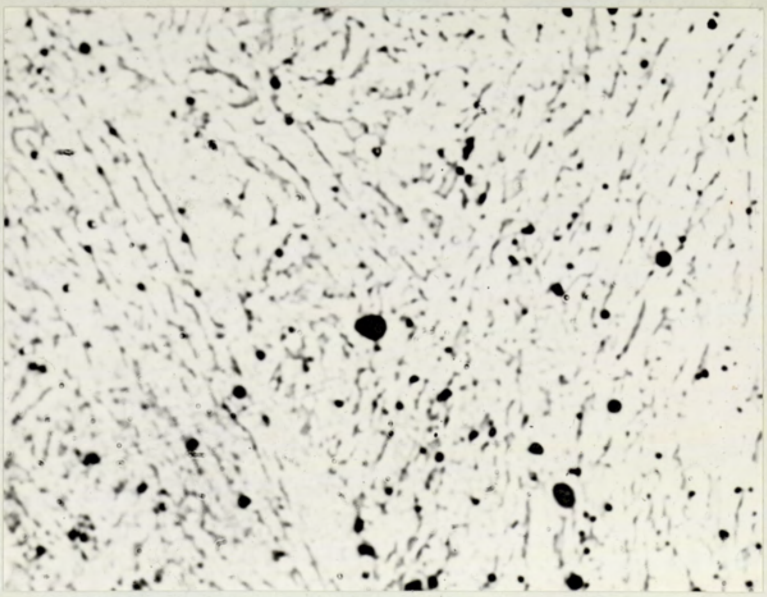


Fig. 68. Weld metal from steel SC, showing rounded inclusions.
Etch: 2% Nital. x 1,000