Electron Beam Welding of C/Mn Steels—Toughness and Fatigue Properties

EB welding speed and weld width variations can substantially alter weld metal microstructures and toughness

BY S. ELLIOTT

Introduction

Electron beam welding is a deep penetration welding process. It is capable of producing single pass, complete joint penetration welds in steels up to 200 mm (7⅞ in.) thick. This type of weld is characterized by a narrow parallel sided fusion zone, with a narrow HAZ. As such it produces a geometry quite different from that obtained by conventional arc welding processes (*e.g.*, SMA or submerged arc). This type of weld is generally associated with low distortion because of the relatively low heat input compared with arc welds. When fabricating a component by electron beam welding, welding conditions are usually chosen to produce a particular weld bead appearance and to avoid discontinuities such as porosity and cracking. It is therefore important to know how the choice of welding conditions, *e.g.*, welding speed and weld width, may affect the mechanical properties of welds made.

The work described in this paper was carried out to investigate the microstructural variations produced in carbon manganese steels as a result of changes in the process conditions used, and the resultant effects on toughness characteristics of the welds. The effect of changes in the weld bead geometry on the fatigue properties of these welds was also included in this investigation; the reason for this is that the top bead geometry of electron beam welds is quite different from that commonly encountered with arc welds. In particular, the top weld bead has a less severe toe geometry than that produced by arc welding. For this reason, it might be expected to exhibit a relatively high fatigue strength.

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Background Literature

The amount of published material on EB welding is quite vast. Nonetheless, the toughness properties of EB welds are not well documented. Few systematic surveys covering a wide range of steels have been reported. This is because most of the published literature is confined to achieving acceptable levels of toughness for particular welded components. However, one survey of mechanical properties of EB welds in a wide range of different structural steels has been published by Arata *et al* (Refs. 1-4). In addition, both Yada and Nisida (Ref. 5) and Steffens and Sepold (Ref. 6) have published reports on the strength aspects of EB welds in C-Mn, low alloy, and mild steels, respectively.

Other similar investigations are reported by Russell *et al* (Ref. 7) and Matting and Sepold (Ref. 8). These reports show that, in general, acceptable toughness values can be obtained in EB welds, although there is often a need for preheat or postweld heat treatment (PWHT) to achieve these values. The level of heat input and cooling rate is an important factor in determining toughness and is discussed by several of these authors (Ref. 1-6, 8).

A major problem in applying the Charpy V-notch test to the narrow fusion zones, typical of EB welds, is highlighted in the literature by Goldak and Nguyen (Ref. 9) and Arata *et al* (Ref. 4). These authors note that some variation in fracture mode occurs in that there is a tendency for fracture to propagate into the base metal rather than weld metal. These failures can be classified into three types:

1. Where fracture occurs totally through weld metal.
2. Where fracture deviates from the weld to the HAZ.
3. Where fracture deviates from the weld to the base metal.

The tendency for fracture to propagate into the base metal rather than the weld metal can give misleading results. This occurs because the resulting absorbed energy, often a high value, is not representative of the energy absorbed by the weld metal.

Goldak and Nguyen (Ref. 9) suggest that this problem arises because the yield strength of the weld metal is higher than that of the base metal. Thus in narrow welds, when the stress field around the crack tip of a Charpy specimen is large enough to extend into the base metal, the base metal yields first, easing the strain around the crack tip. This facilitates ductile failure in the base metal by a "plastic hinge" mechanism, *i.e.*, the base metal fracture strain will be exceeded before the weld metal reaches its fracture strain. As a result, the fracture will propagate through the base metal, even if the weld metal has lower toughness.

Goldak and Nguyen (Ref. 9) dismiss the argument that if a fracture did not propagate through the weld metal above the Charpy V notch transition temperature, then such a weld would not be likely to suffer brittle failure in service. This is argued on the grounds that the increased size of a welded structure, compared to a Charpy specimen, would cause the plastic hinge failure mechanism to be less favorable. The fracture path would then be more likely to occur in the brittle weld metal. This latter argument would also imply that toughness testing using larger specimens, *e.g.*, fracture initiation tests such as the crack tip opening displacement (CTOD) tests, would be more appropriate to narrow welds. On the basis of these arguments, Goldak and

Nguyen suggested that the use of fatigue precracked Charpy specimens may be more favorable than V-notched specimens. The reason was that the stress field around the crack tip should be smaller and therefore less likely to extend into the base metal.

Fatigue pre-cracked Charpy specimens were used in some later work reported by Goldak and Bibby (Ref. 10). Over the series of trials conducted, the fatigue precracked specimens were found to produce a true transition temperature as measured by absorbed energy in only a few cases. These authors therefore concluded that the precracked Charpy test was not an effective solution to the problem of testing narrow welds.

Although this problem with Charpy testing narrow welds has been highlighted in the literature, earlier work at The Welding Institute such as that reported by Russell *et al* (Ref. 7) has shown that Charpy testing does not always produce a fracture away from the weld metal in EB welds. There must be a limiting weld width above which this phenomenon does not occur; there may also be an influence from the material used. It was, therefore, decided to use Charpy testing initially in this work so as to define the limits of its usefulness.

Materials

A series of killed carbon manganese steel plates within the thickness range 12.5-75 mm (approximately ½ to 3 in.) were used, the compositions of which are given in Table 1. The thinnest of these (A, 12.5 mm or 0.49 in.) is a niobium-containing steel made to BS4360 50D. The thickest steel (D, 75 mm or 2.95 in.) is made to the same specification but does not contain niobium. The 25 mm or 0.98 in. (B) and 50 mm or 1.97 in. (C) thick plates were made to BS1501 224, but are of similar composition to materials A and D.

Welding trials for Charpy impact testing were carried out on the full range of these steels. However, only the thinnest (A, 12.5 mm or 0.49 in.) was used for fatigue testing.

Welding Procedure

The electron beam (EB) welds were made in vacuum, using The Welding Institute's 75 kW EB facility. All welds were made as butt joints in the horizontal/vertical (HV) position, using an accelerating voltage of 150 kV. A variety of welding currents, welding speeds and beam deflection patterns were used to produce the required variations of weld width, weld speed and bead geometry (the latter for fatigue tests only). The welding conditions are not reproduced in detail in this paper, since they relate specifically to The Welding Institute equipment and thus are not transferable to other machines.

Welding conditions were chosen to give a range of welding speeds at a given weld width (as appropriate to the material thickness) and a range of weld widths at constant speed. The range of weld widths was produced by adjusting the beam focus position with respect to the surface; in each case, the beam power was adjusted to produce a complete penetrating, parallel sided weld joint after an initial low power run. It was found that the range of practical weld widths that can be produced in thinner materials (*e.g.*, 12 mm or 0.47 in.) is restricted; this is because excessive width leads to top bead sink and undercut. Similarly, the practical speed range for welding thick materials is limited in that fast welding speeds tend to be associated with the onset of solidification cracking. Variations of weld width and speed were, therefore, made within these confines.

Welds for fatigue testing were made at one welding speed (500 mm/min equivalent to 19.7 ipm) with a fusion zone width of 3 mm (0.012 in.). Variations in the weld bead profile, *e.g.*, undercut top bead, were produced by slight alterations in position of beam focus.

Weld Assessment

Each weld was sectioned and Vickers hardness measurements made. Toughness testing was carried out using standard Charpy V-notch specimens machined in the through-thickness direction with the notch located in the center of weld metal (Fig. 1A). The majority of welds were tested both as-welded and with a standard postweld heat treatment (2-3 h, depending on thickness, at 600°C, *i.e.*, 1112°F) over a range of test temperatures and transition curves produced.

Fatigue testing was carried out under axial loading at an applied minimum/maximum stress ratio (R) of zero using waisted transverse butt weld specimens (Fig 1B). The specimens were tested at frequencies within the range 5–20 Hz on hydraulic fatigue testing machines.

Results and Discussion

Effect of Welding Speed

12.5 mm (0.49 in.) Material. Maintaining a constant weld width of approximately 2 mm (0.079 in.), welds were made using speeds ranging from 150 to 750 mm/min (5.9 to 29.5 ipm). Changes in welding speed over this range produced a significant variation in microstructure as shown in Fig. 2. Variations ranged from predominantly lower temperature transformation products generally classified as ferrite with aligned MAC (martensite/austenite/carbide) (Ref. 11) at 150 mm/min (5.9 ipm) welding speed (Fig. 2A) to almost 100% martensite at the high welding speed of 500 mm/min (29.5 ipm)—Fig. 2B.

The difference in microstructure might be expected to have a significant effect on the weld metal toughness; indeed this is so, as indicated in Fig. 3. The faster welding speeds of 500 and 750 mm/min (19.7 and 29.5 ipm) indicate an apparent

Table 1—Material Chemical Compositions, wt-%

	Material[a]			
	A	B	C	D
C	0.19	0.22	0.20	0.18
S	<0.005	0.022	0.014	0.005
P	0.025	0.020	0.016	0.008
Si	0.42	0.25	0.28	0.24
Mn	1.34	1.54	1.37	1.50
Ni	0.03	0.05	0.03	0.67
Cr	0.03	0.05	0.02	0.13
Mo	0.01	0.02	<0.01	0.05
V	0.01	<0.01	<0.01	<0.01
Cu	0.01	0.08	0.02	0.03
Nb	0.018	<0.005	<0.005	<0.005
Ti	0.006	0.005	0.004	<0.01
Al	0.052	0.032	0.025	0.037
B	<0.0005	<0.0005	<0.0005	<0.0005
Sn	<0.01	<0.01	<0.01	<0.01
Co	<0.01	<0.01	<0.01	<0.01
CE[b]	0.43	0.5	0.45	0.52
N_2[c]	80	74	75	86
O_2[c]	21	20	21	34

(a) Material thicknesses and specification nos. as follows: A—13 mm (0.51 in.), BS4360-50D; B—25 mm (0.98 in.) BS1501-224; C—50 mm (1.97 in.), BS1501-224; D—75 mm (2.95 in.), BS4360-50D. (b) CE = carbon equivalent with IIW CE

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

(c) N_2 and O_2 data are averages in ppm.

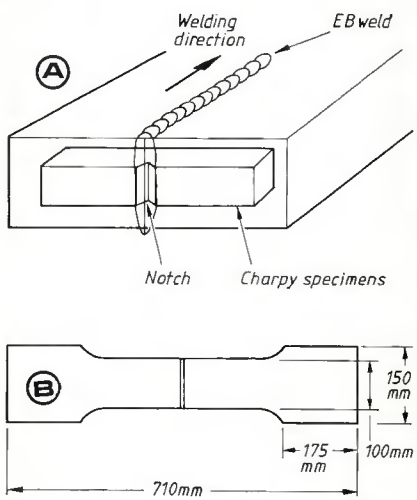

Fig. 1—Specimens used for weld assessment: A—location of Charpy V-notch specimens with respect to EB weld; B—fatigue specimen design

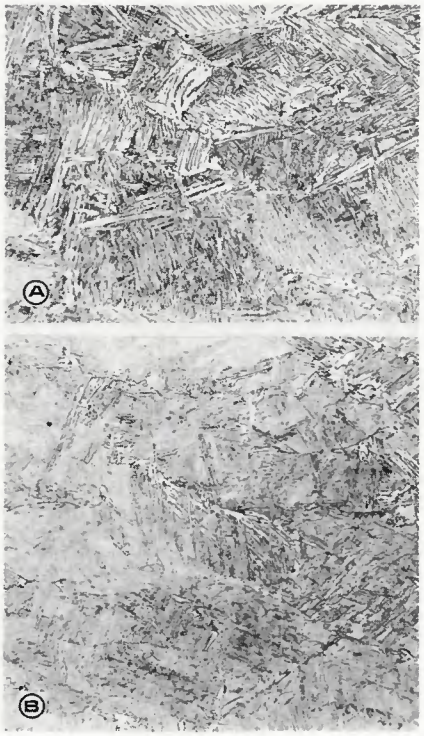

Fig. 2—Weld metal microstructures for EB welds in 12.5 mm (0.49 in.) thick 0.19 C-1.35 Mn steel: A—weld made at 150 mm/min (5.9 ipm); B—weld made at 500 mm/min (19.7 ipm). X630 (reduced 50% on reproduction)

Table 2—Effect of Welding Speed on As-Welded Weld Properties for 12.5 mm (0.49 in.) Thick Plate at Constant 2 mm (0.08 in.) Weld Width

Welding speed, mm/min[a]	27 J Transition temperature, °C[b]	40 J Transition temperature, °C[b]	Fusion zone hardness, (min-max) / mean
150	+20	+30	307-348 / 323
250	+20	+30	326-370 / 330
500	−20	−10	404-444 / 423
750	−30	−15	398-427 / 422

(a) mm/min ÷ 25.4 = ipm.
(b) $°F = \frac{9}{5}°C + 32$

ductile/brittle transition temperature of approximately −10 to 0°C (14 to 32°F). On the other hand, the slower 150 and 250 mm/min (5.9 and 9.8 ipm) welding speeds do not show a transition from cleavage to ductile fracture until approximately +50°C (122°F). These differences are also reflected in comparisons of 27 and 40 J transition temperatures and hardnesses as indicated in Table 2.

One feature apparent from the Charpy testing was that, at a certain test temperature, the fracture path in the Charpy impact test specimens did not stay confined to the weld metal; instead, it deviated into the base metal as described earlier. The results were misleadingly high and not representative of the energy absorbed by the weld metal.

In this work, the temperature at which the fracture path deviates from the weld metal into the HAZ or base metal is classified as T_{DV} (temperature of deviation). Each Charpy specimen was examined on testing to assess when this type

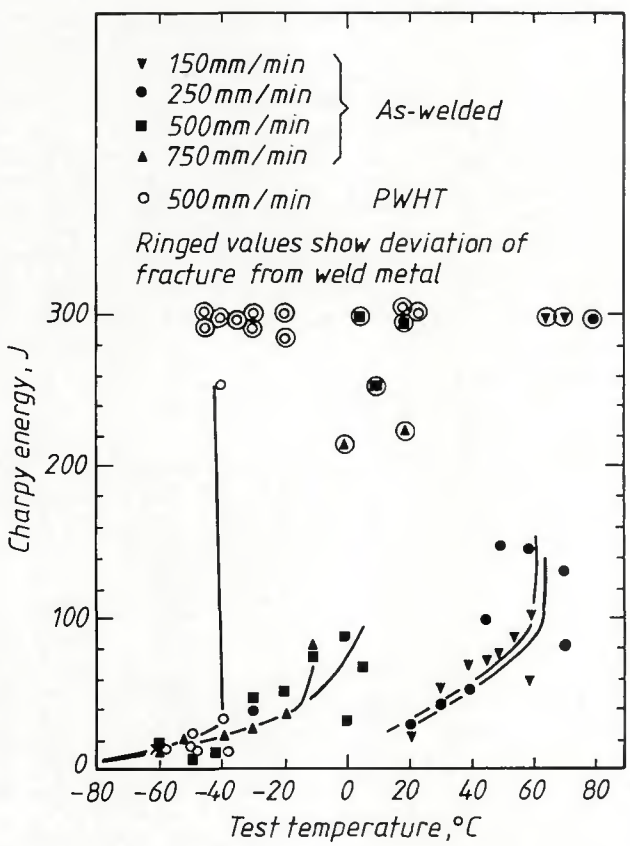

Fig. 3—Effect of welding speed on toughness for EB welds in a 13 mm (0.51 in.) thick C/Mn steel (BS4360:50D with 0.19C — 1.35 Mn). Each symbol represents one CVN sample

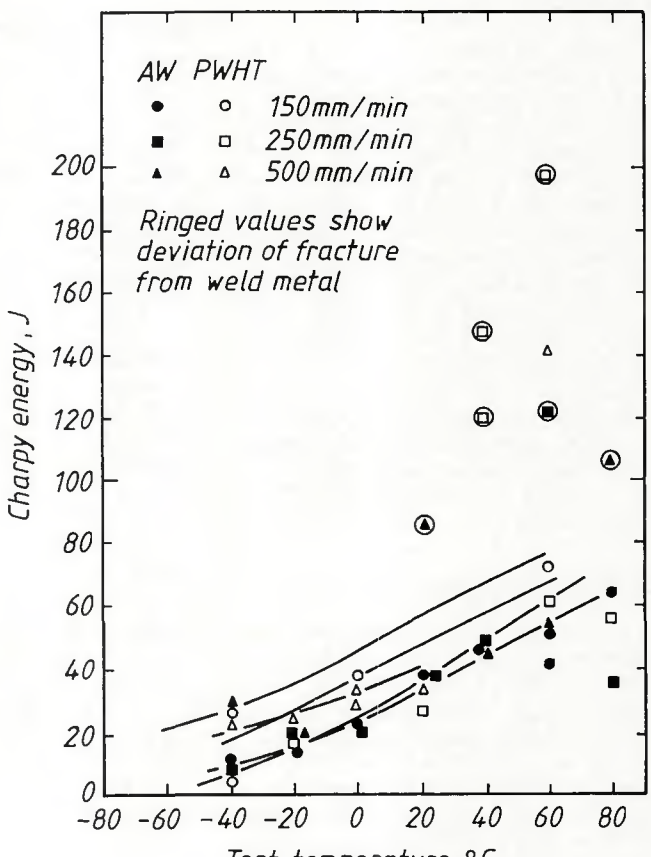

Fig. 4—Effect of welding speed on toughness for EB welds in a 25 mm (0.98 in.) thick C/Mn steel (BS1501-224 with 0.22C — 1.54 Mn)

of failure had occurred. Results, which deviated, are shown as ringed values where they are reported in the illustrations used. The implications for Charpy testing narrow welds resulting from this classification is discussed later in the paper.

25 mm (0.98 in.) Material. No effect of welding speed on Charpy toughness levels of the 25 mm (0.98 in.) thick steel (B) to BS1501 244 was observed when welding at speeds of 150, 250, and 500 mm/min (5.9, 9.8, and 19.7 ipm), and the toughness levels were generally low—Fig. 4. This was reflected in the weld metal microstructure which remained similar over this welding speed range. Also, the microstructure consisted predominantly of areas of fine ferrite with aligned MAC. Little improvement was observed with postweld heat treatment.

50 mm (1.97 in.) Material. With the thicker steel, (C), a change in microstructure was observed when welding at speeds of 100, 150 and 250 mm/min (3.9, 5.9, and 9.8 ipm). This related more to the coarseness of the microstructure produced rather than the transformation products generated, as shown in Fig. 5. The microstructure associated with the slowest welding speed (100 mm/min, *i.e.,* 3.9 ipm) (Fig. 5A) shows predominantly ferrite with aligned MAC with ferrite carbide aggregate. At the higher welding speed (250 mm/min, *i.e.,* 9.8 ipm) the microstructure becomes more varied with coarser colonies of ferrite with aligned MAC. The coarser colony size would be associated with lower cleavage resistance.

The slowest welding speed of 100 mm/min (3.9 ipm) showed the highest

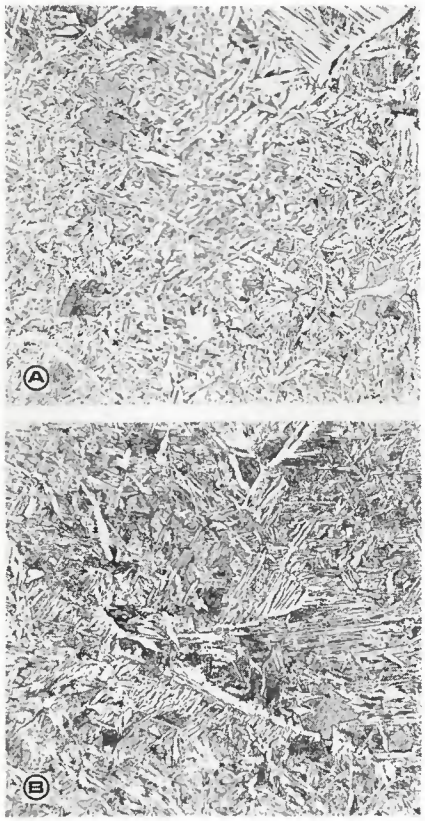

Fig. 5—Weld metal microstructures for EB welds in 50 mm (1.97 in.) thick steel (BS1501-224 with 0.2 C — 1.37 Mn): A—weld made at 100 mm/min (3.94 ipm); B—weld made at 250 mm/min (9.8 ipm). X630 (reduced 50% on reproduction)

Charpy energy values in the as-welded condition—namely, a 27 J transition temperature of −35°C (−31°F) compared to −15°C (+5°F) for the faster welding speed of 250 mm/min (9.8 ipm). The situation was reversed with PWHT when the faster welding speed of 250 mm/min (9.8 ipm) produced the highest Charpy values. Again, difficulties were found in producing a complete curve for welds in the PWHT condition because of deviation of the fracture path from the weld metal at relatively low temperatures.

75 mm (2.95 in.) Material. When EB welding thick section material, the range of practical welding speeds is limited. This is because the increased restraint arising from plate thickness encourages the tendency for solidification cracking to occur at fast welding speeds, which are associated with severe thermal cycles. The 75 mm (2.95 in.) thick plate (D) was therefore welded at only two speeds (100 and 150 mm/min, *i.e.,* 3.9 and 5.9 ipm) at a constant weld width of 3 mm (0.12 in.). No difficulty was found in Charpy testing these welds, since the fracture path remained in the weld metal over the complete range of test temperatures.

The as-welded Charpy results are similar for the two welding speeds. This is indicated in Fig. 6, but a substantial improvement in weld metal toughness is obtained with postweld heat treatment. The weld metal microstructures are similar for the two welding speeds, consisting mainly of a fine ferrite carbide aggregate with some islands of martensite and ferrite laths.

Effect of Weld Width on Toughness

The results reported in this section are confined to the thicker materials investigated in this study (25–75 mm, *i.e.,* 0.98 to 2.95 in.) because of the difficulties of producing a wide range of weld widths in thin materials. The variation of Charpy toughness with weld width for the 25 and 50 mm (0.98 and 1.97 in.) thick plate

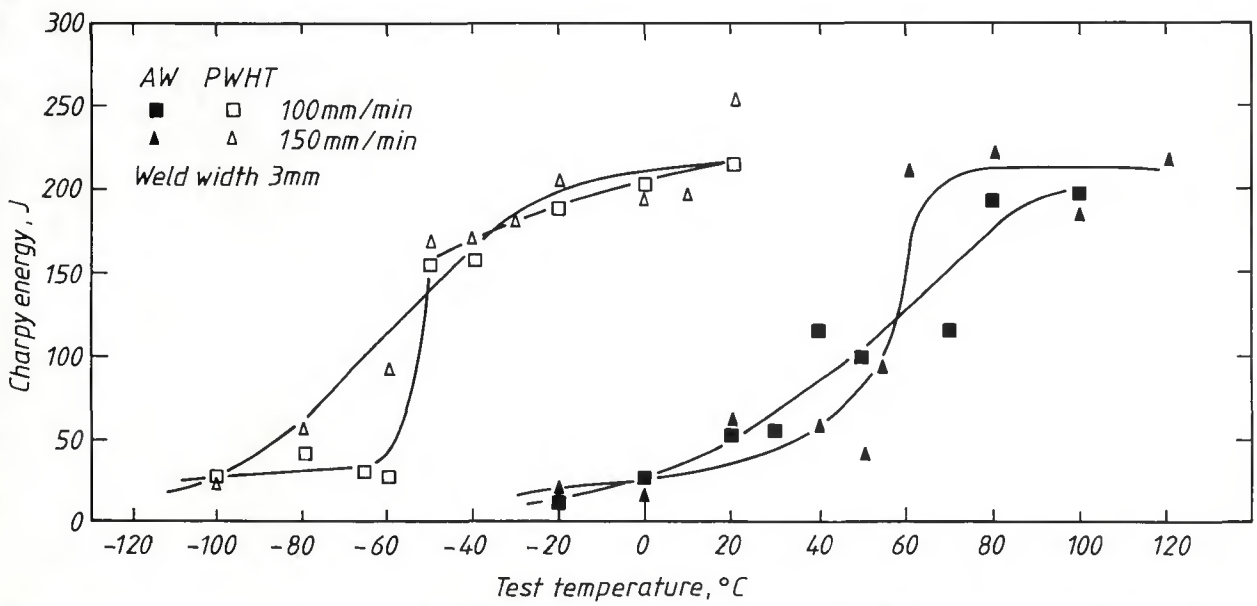

Fig. 6—Effect of welding speed on toughness of EB welds in a 75 mm (2.95 in.) thick C/Mn steel (BS4360 50D with 0.18 C — 1.50 Mn)

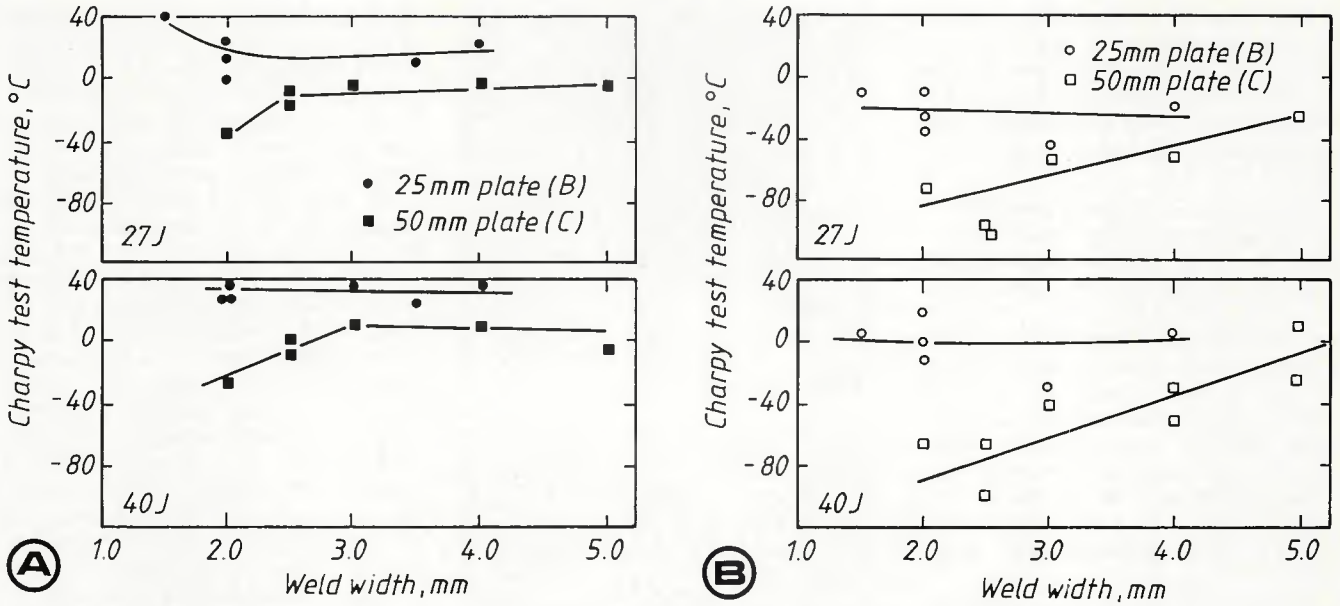

Fig. 7 – Variation of 27 and 40 J Charpy test temperatures with weld width for weld in 25 and 50 mm (0.98 and 1.97 in.) thick plates (BS1501-224): A – as-welded; B – PWHT

is summarized in Fig. 7 as variations of 27 J and 40 J transition temperatures with weld width.

25 mm (0.98 in.) Thick Material. Weld widths covering a range 1.5 to 4 mm at a constant welding speed of 250 mm/min (9.8 ipm) were tested. Substantial variations in microstructure were achieved – Fig. 8. However, these changes were not reflected in the Charpy results obtained (Fig. 7), either in the as-welded or PWHT condition. The reason for this is not clear, although neither of the microstructures shown in Fig. 8 are associated with particularly high toughness. Perhaps this combined with the relatively high carbon level of this steel (0.2% C) would prevent any differences in toughness being observed.

50 mm (1.97 in.) Thick Material. Welds covering a range of weld widths from 2.5 to 5.0 mm (0.098 to 0.197 in.) were produced at a welding speed of 150 mm/min (5.9 ipm) for Charpy testing. The microstructural variations produced as a result of change in weld width are shown in Fig. 9. The microstructural constituents are basically similar consisting of ferrite laths associated with some ferrite carbide aggregates, but the wider weld produces a generally coarser microstructure.

The effect on Charpy energies of this microstructural difference is relatively small in the as-welded condition – Fig. 7. However, toughness is significantly improved with postweld heat treatment for the narrower welds – that is to say, a 27 J transition temperature of <–80°C (–112°F) for the narrow weld compared with –20°C (–4°F) for wide welds.

The improved toughness with postweld heat treatment for the narrow weld compared to the wide is related to the relatively fine microstructure associated with the narrow weld – Fig. 9A. In the as-welded condition, the wider weld is softer than the narrow weld (240 HV compared to 260 HV). This could account for the slightly higher toughness observed for the 5.0 mm (0.197 in.) wide weld. However, the microstructure,

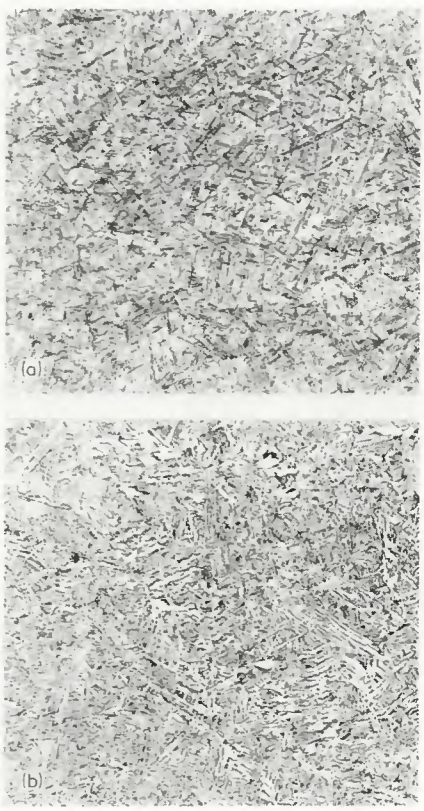

Fig. 8 – Weld metal microstructures for EB welds in 25 mm (0.98 in.) thick steel (BS1501-224 with 0.22 C – 1.54 Mn): A – 1.5 mm (0.06 in.) weld width; B – 4 mm (0.16 in.) weld width. X630 (reduced 50% on reproduction)

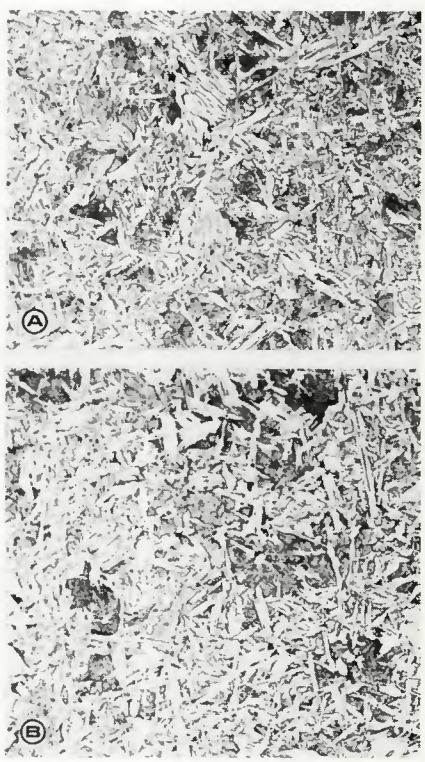

Fig. 9 – Weld metal microstructure for EB welds in 50 mm (1.97 in.) thick steel (BS1501-224 with 0.2 C – 1.37 Mn) made at constant 150 mm/min (5.9 ipm) welding speed: A – 2.5 mm (0.098 in.) weld width; B – 5.0 mm (0.197 in.) weld width. X630 (reduced 50% on reproduction)

which is predominantly one of ferrite laths combined with polygonal ferrite and areas of ferrite carbide aggregate, is not generally associated with good toughness. The finer scale and greater uniformity of the microstructure in the narrow weld would lead to a tempered microstructure more resistant to cleavage than the coarse microstructure associated with the wide weld. This would account for a reversal in trends between as-welded and heat-treated microstructures.

75 mm (2.95 in.) Plate. Again, thick plate poses limitations on the range of weld widths practical to achieve with EB welding. The relatively high heat input necessary to achieve complete penetration eliminates the possibility of making very narrow welds.

Only two weld widths – 3 and 5.5 mm (0.12 and 0.22 in.) – were investigated at this plate thickness at a welding speed of 100 mm/min (3.9 ipm). The wide weld produces a significantly coarser microstructure than the narrow weld – Fig. 10; again, the finer weld metal microstructure is reflected in higher levels of Charpy toughness – Fig. 11. A significant improvement was observed for both these welds with postweld heat treatment.

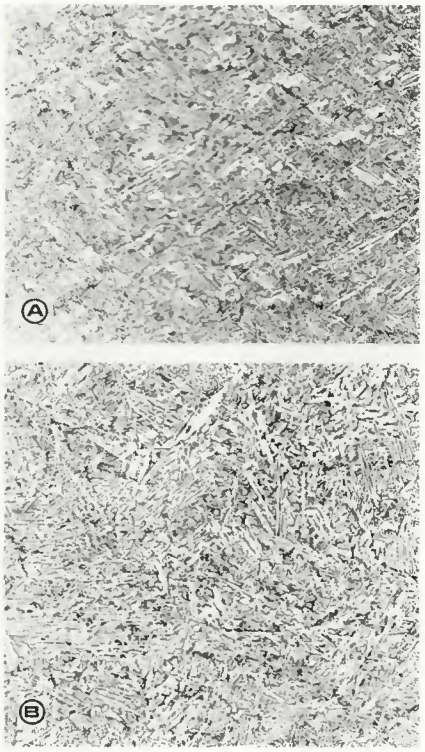

Fig. 10 – Weld metal microstructures for EB welds in 75 mm (2.95 in.) thick steel (BS4360-50D with 0.18 C – 1.50 Mn): A – 3 mm (0.12 in.) weld width; B – 5.5 mm (0.22 in.) weld width. X400 (reduced 50% on reproduction)

Use of the Charpy Test for Assessing Narrow Welds

As indicated in the results presented in this paper, the usefulness of Charpy testing narrow welds is limited because of the tendency of the fracture to deviate away from the weld metal. The test temperature at which fracture deviated away from the weld was found to vary between welds.

A graphical representation of the test temperature at which fracture deviated from the weld metal (T_{DV}) against weld width for these results (Fig. 12), shows that for weld widths greater than 3 mm (0.12 in.), the T_{DV} is <+80°C (176°F). This illustrates that Charpy test specimens should be carefully inspected if they are used to test weld widths less than 3 mm (0.12 in.) to ensure that the fracture path remains in the weld. A comparison between Figs. 12A and 12B shows that in general, postweld heat treatment tends to reduce the value of T_{DV} for a given weld width.

This trend can be explained in terms of the previously mentioned model described by Goldak and Nguyen (Ref. 9). Here the PWHT would tend to increase the cleavage resistance of weld metal so that the adjacent base metal would be more likely to exceed the yield stress before the fracture stress of the weld metal was reached, *i.e.*, the "plastic hinge" mechanism of failure would be more likely to operate at lower test temperatures, in preference to brittle failure through the weld metal. Thus, the T_{DV} seems to vary inversely with weld metal toughness.

A similar tendency can be seen when T_{DV} is plotted against weld metal hardness – Fig. 13. Here the trend is for T_{DV} to increase with increasing hardness, *i.e.*, decreasing cleavage resistance and increasing yield stress. Only welds in the thinnest material (A) do not follow this general trend. Here the trend is reversed in that the T_{DV} tends to decrease with increasing hardness. The implication is that the cleavage resistance of the Nb treated C/Mn steel weld metal is not so low as the untreated C/Mn steel weld metal of equivalent hardness (420 Hv).

Further investigations are required to confirm this trend. However, these results indicate that, even if the fracture path does deviate away from the weld metal in a Charpy test, the temperature at which this occurs can provide an indication of relative toughness (*e.g.*, low T_{DV} indicates high cleavage resistance).

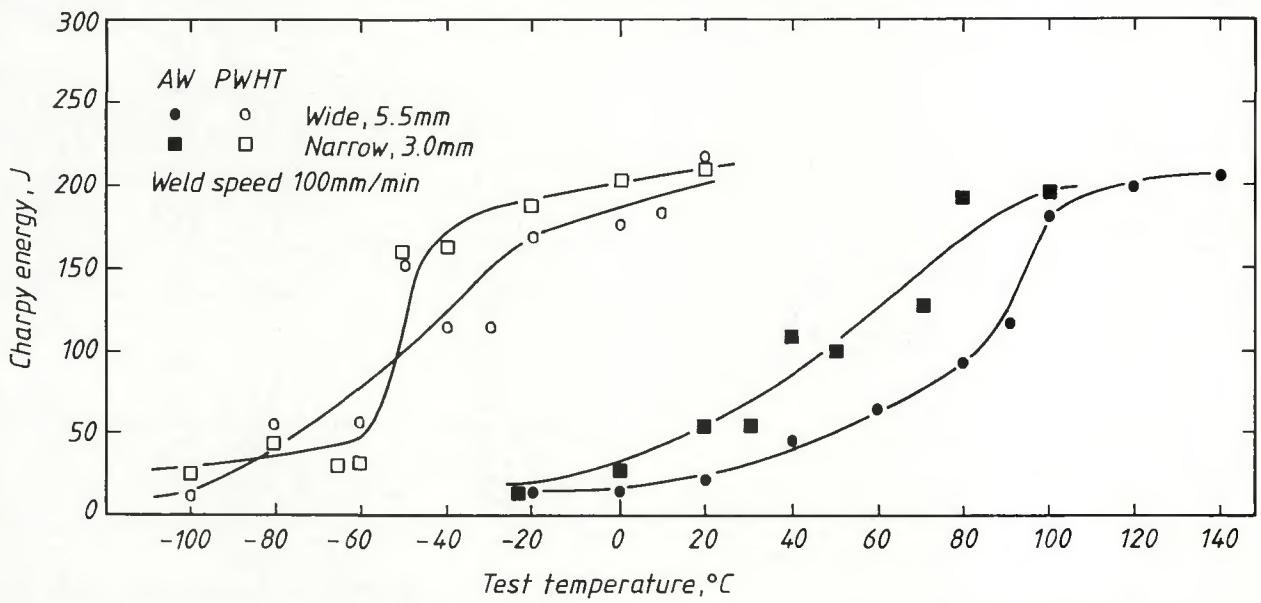

Fig. 11 – Effect of weld width on toughness for EB welds in a 75 mm (2.95 in.) thick C/Mn steel (BS4360-50D with 0.18 C – 1.50 Mn)

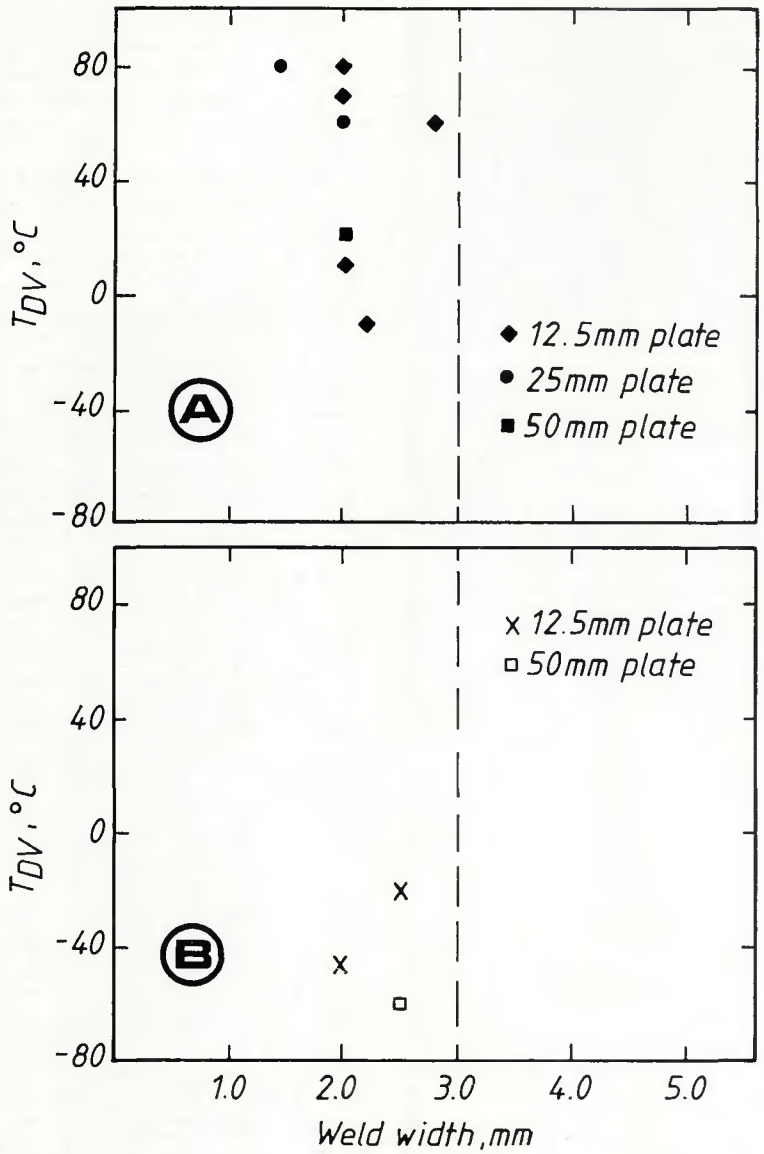

Fig. 12—Variation of temperature at which fracture deviates from the weld metal (T_{DV}) with T_{DV} shown as a function of weld width: A—as-welded; B—after PWHT

Fatigue Results

Fatigue test results for welds with a good bead profile and those with an undercut top bead are presented in the form of S/N diagrams in Fig. 14. All specimens failed from fatigue cracks initiating at the plate surface. The as-welded specimens failed from cracks initiating at the edge of the top bead and propagating across the section, while the undercut specimens failed from cracks initiating at the undercut. The presence of undercut on the top bead reduces the fatigue life significantly below the scatter band for good as-welded specimens. Typical sections showing failed welds with and without undercut are shown in Fig. 15.

A scatter band drawn to encompass the as-welded test results from good welds (*i.e.*, no undercut) shown in Fig. 14, is compared in Fig. 16 with published scatter bands (Refs. 12, 13) for transverse double welded butt joints made by arc welding processes. It should be noted that, although the scatter band for EB welds is much narrower than the others, this may just be a function of the relative data populations.

The mean line drawn by the eye through the EB test results is almost coincident with the mean line through the data obtained for transverse butt welded joints made by the shielded metal arc process. It is also just below the upper scatter band for joints made by submerged arc welding. Thus, from the limited number of tests carried out, it seems that transverse butt welded joints made by EB welding possess similar fatigue strength to joints made by SMA and better fatigue properties than joints made by submerged arc welding. Undercut of the top bead severely reduces fatigue life—Fig. 14. However, undercut of this severity (approximately 1.1 mm in a plate thickness of 12.5 mm or 0.04 in. in 0.49

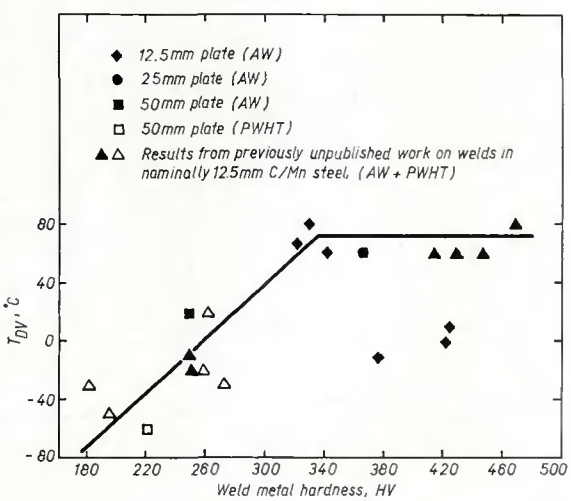

Fig. 13—T_{DV} vs. weld metal hardness

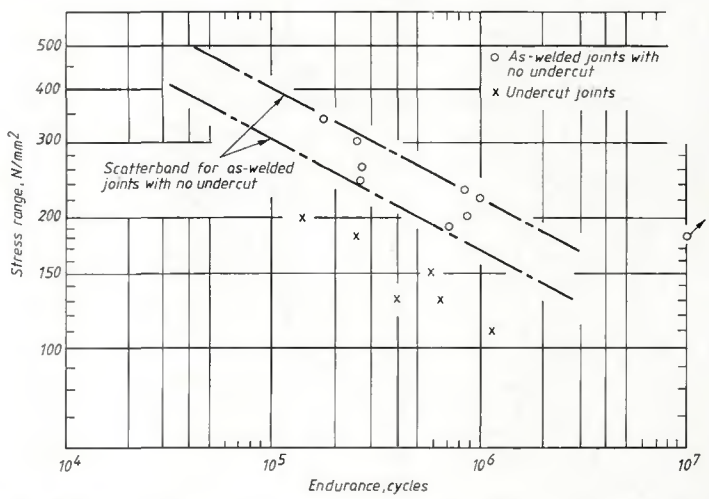

Fig. 14—Fatigue results for EB as-welded transverse welded butt joints, with and without undercut

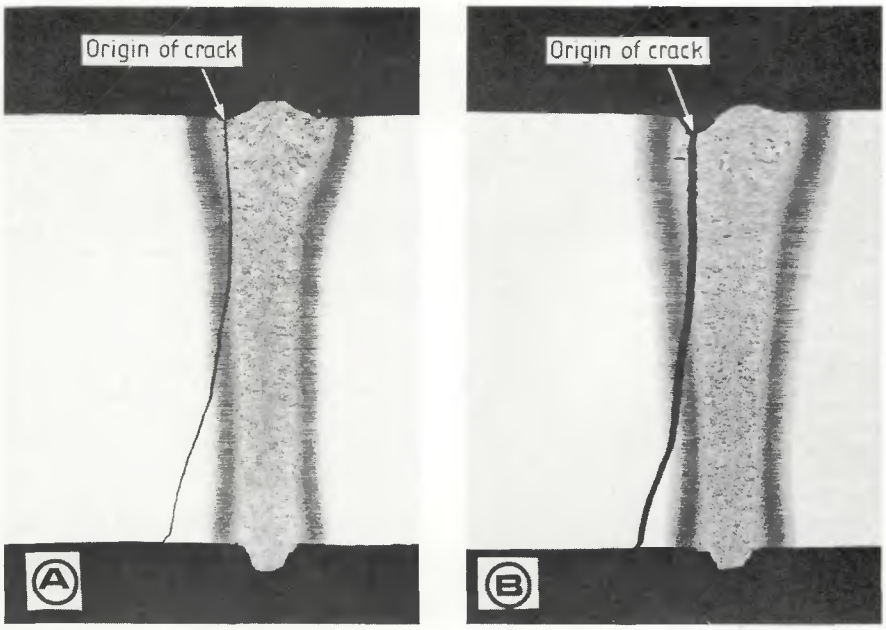

Fig. 15 — Typical failed welds: A — as-welded "good" joint; B — failed joint with severe top bead undercut. X5 (reduced 10% on reproduction)

in.) would undoubtedly be considered unacceptable by most fabrication standards. The low fatigue strengths obtained should, therefore, not be surprising.

Conclusions

1. The work described in this paper has shown that the choice of EB welding conditions of speed and weld width can significantly affect the resultant weld metal microstructure. This may result in differences in toughness as measured by the Charpy test, depending on plate thickness and composition. It is, therefore, important to take into account these effects when developing EB welding conditions for a critical application rather than considering weld bead appearance and defect occurrence alone.

2. It has been shown that for the materials studied, the fracture path does not deviate away from the weld metal for weld widths >3 mm (0.12 in.) when using the Charpy V-notch impact test. The test temperature, at which deviation of the fracture path from the weld metal occurs (T_{DV}) at weld widths less than this, appears to increase with weld metal hardness for carbon manganese steels with no microalloying elements. The T_{DV} can, therefore, be considered to provide an indication of relative toughness values for welds.

3. Transverse butt welded joints made by EB welding show a fatigue strength comparable to joints made with the shielded arc processes, and superior to those made by submerged arc welding.

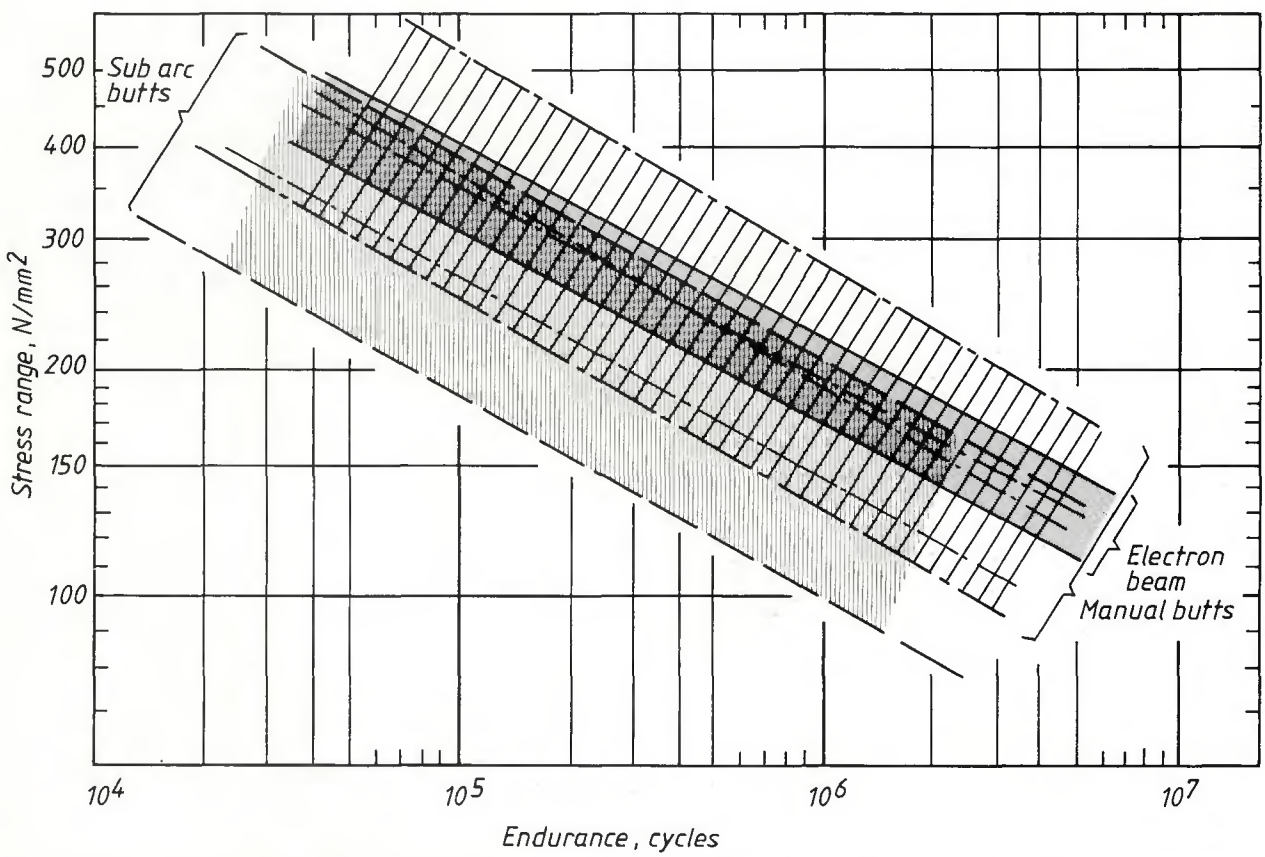

Fig. 16 — Comparison of results obtained with EB welding for transverse as-welded joints vs. results obtained using SMAW and submerged arc welding

Top bead undercut produces a significant reduction in fatigue strength amounting to approximately 40% at an endurance level of 10^6 cycles.

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