CRANFIELD REPORT MAT. NO. 1

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BY

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Cranfield Report Mat. No. 1.

April, 1970.

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DEPARTMENT OF MATERIALS

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E. Smith, Ph.D., B.Sc., A.I.M.

Summary

Structures representative of the visible weld heat-affected zone (HAZ) of a semi-killed C - Mn steel and a fully killed and normalised aluminium grain - refined C - Mn steel have been produced by means of a simulation technique. The Charpy V - notch impact properties have been assessed and related to microstructure. The only significant deterioration in impact properties occurred in the coarse grained HAZ of the aluminium grain-refined steel.

The notch - impact properties of the subcritical weld HAZ have been assessed by simulating specimen blanks with and without notches through thermal cycles reported to cause the greatest degree of subcritical embrittlement in these steels. The results showed that no significant embrittlement occurred in either steel in the absence of notches. In the presence of a notch there was evidence of an increase in transition temperature of 10 35°C in the semi-killed steel but no increase in the aluminium grain - refined steel. The slight increase for the semi-killed steel was attributed to the presence of free nitrogen, which was absent in the aluminium grain - refined steel.

CONTENTS

1. INTRODUCTION

The discovery that the class of steels known as "the grain-refined C-Mn steels" possessed not only superior strength but also superior fracture toughness to the plain C - Mn steels considerably broadened the range of uses of C - Mn steels. The grain-refinement is achieved by the addition of a small quantity of a grain-refining element such as aluminium, niobium, titanium, or vanadium to provide many nucleation sites for the ferrite, and also by controlling the final heat treatment temperature. Aluminium has an additional benefit in steels containing large quantities of nitrogen in that it combines with the nitrogen to form AlN and so reduces the harmful effect caused by the presence of free nitrogen. Thus the grain-refined steels can be economically incorporated in structural steelwork for use at higher stresses or lower temperatures than would be possible with plain C - Mn steels.

The assembly of structural steelwork invariably involves welding. This poses the problem as to whether the superior properties of the grain-refined C - Mn steels are maintained in the weld HAZ, including the subcritical HAZ. Irvine et al (Ref. 1) have shown that of the four elements normally used for grain-refinement in steels i.e. aluminium, niobium, titanium, and vanadium, aluminium is the most effective in maintaining good impact properties after the steel has been subjected to high austenitising temperatures such as happens in the coarse grained HAZ of a weld. This is because aluminium produces no dispersion hardening while the other three elements are taken into solution at high temperatures and produce an appreciable degree of dispersion hardening on subsequent cooling with a consequent reduction in impact properties.

Another important benefit of the grain-refining elements is that they raise the grain-coarsening temperature of steels. Chalmers (Ref. 2) has shown, however, that in the aluminium-treated steels once the grain-coarsening temperature is exceeded grain growth occurs at a faster rate than in an aluminium free steel so that the austenite grain size of the aluminium-treated steel tends to approach that of the aluminium free steel.
The overall effect in a weld HAZ would be to produce a narrower region of grain coarsening with a consequent benefit to the weldment as a whole.

One of the objects of the present work was to compare the notch-impact properties of the visible weld HAZ of a semi-killed C - Mn steel and a fully killed and normalised aluminium grain-refined C - Mn steel.

Embrittlement outside the visible weld HAZ has been reported frequently in the literature on C - Mn steels (Refs. 3-9) and this is generally accepted as being due to a dynamic strain-ageing phenomenon enhanced by the presence of free nitrogen. It would thus be expected that in steels susceptible to this form of embrittlement the presence of a notch-like defect in this region would enhance the strain produced there because of the triaxial stress system created at the root of the notch and thus give rise to a greater degree of embrittlement.

The present work was undertaken in an attempt to throw some more light onto this problem by examining two steels, one containing a large quantity of free nitrogen and the other containing little free nitrogen by subjecting specimen blanks, with and without notches, to thermal cycles known to cause the greatest degree of embrittlement in the subcritical HAZ in these steels.

2. EXPERIMENTAL

2.1 MATERIALS

The materials used were as follows:-

U steel - semi-killed, as rolled C - Mn steel.

T steel - fully killed, aluminium-treated, normalised C - Mn steel.

Both materials were in the form of 25mm (1 inch) thick plate. The chemical analyses and mechanical properties are listed in Tables 1 and 2 respectively. Details of the nitrogen determinations have been reported by Saunders (Ref. 10) who worked on the same steels.

Material	C	Mn	Si	S	P	Al	Tot. N	N as AlN	Free N
U Steel	0.14	1.23	0.06	0.034	0.021	0.005	0.0045	-	0.0041
T. Steel	0.17	1.43	0.17	0.033	0.027	0.094	0.0051	0.0032	0.0016

Table 1. Chemical Analyses of U and T Steels expressed in weight percent

Material	Lower yield stress		U.T.S.		El %	R.A.	Av. grain diam.	Hardness
	$MN\,m^{-2}$	tonf. in.$^{-2}$	$MN\,m^{-2}$	tonf. in.$^{-2}$	on 8 in.	%	mm	HV5
U. Steel	232	15.1	430	28.0	34	67	0.024	132
T. Steel	338	22.0	490	31.8	32	62	0.008	152

Table 2. Mechanical properties of U and T Steels

2.2 PROCEDURE

2.2.1 Examination of the visible weld HAZ

The methods used generally in this investigation of the visible weld HAZ have been described in detail in an earlier report, (Ref 11). Briefly, this consisted of subjecting specimen blanks 75mm x 10mm x 10mm (3 in. x 0.4 in. x 0.4 in.) machined from the mid-thickness of the plates, to weld thermal cycles in a weld thermal cycle simulator. The thermal cycles were measured in a submerged arc bead-on-plate weld in 38mm (1.5 in.) thick mild steel plate with a heat input of 4.2 KJ mm^{-1} (108 KJ in.$^{-1}$). The welding conditions used during the temperature measurement programme are listed in Table 3 and full details of the technique are given in an earlier report (Ref. 12). An account of the design and operation of the weld thermal cycle simulator has also been published, (Refs. 13 and 14.)

Current (amps)	390
Voltage (volts)	30
Welding Speed (mm/sec.)	2.75
Preheat Temperature	None
Heat Input (KJ/mm.)	4.2

Four thermal cycles were used to produce structures representative of the four regions of the visible weld HAZ. These are shown in fig. 1. The four regions, with peak temperatures, were as follows:-

(1) the coarse grained HAZ, peak temperature	1350°C	U steel.
	1350°C	T steel.
(2) the fine grained HAZ, peak temperature	950°C	U steel.
	910°C	T steel.
(3) the intercritical HAZ, peak temperature	790°C	U steel.
	770°C	T steel.
(4) the spheroidised HAZ, peak temperature	740°C	U steel.
	730°C	T steel.

In some cases the thermal cycles shown in fig. 1 were modified slightly to give the appropriate peak temperature, but in every case the heating and cooling rates shown in fig. 1 were used.

After thermal simulation one specimen from each category was used for hardness determination and metallographic examination using both optical microscopy and electron microscopy of carbon extraction replicas. The remaining specimens were machined into standard Charpy V-notch impact specimens and a Charpy transition curve determined in the normal manner by testing over a wide range of temperature.

2.2.2 Examination of the subcritical weld HAZ

Previous work at the Welding Institute on these steels has shown that the maximum degree of embrittlement in the subcritical weld HAZ occurs in regions experiencing weld thermal cycles with peak temperatures of 490°C and 670°C for U and T steel respectively. (Ref. 15). Consequently weld thermal cycles with these peak temperatures were used in this stage of the work and these are shown in fig. 2. The specimens were rigidly clamped in the simulator for this work in an attempt to induce thermal strains during the weld thermal cycle as in an actual welding situation. For each steel one set of un-notched specimens were cycled once through the appropriate temperature cycle and three sets of notched specimens were cycled once, twice, and three times respectively through the same temperature cycle. One specimen from each category was used for a hardness survey across the heat treated section and the remaining specimens were used for the determination of a Charpy V-notch impact curve by testing over a wide range of temperature.

3. RESULTS

3.1 THE VISIBLE WELD HAZ

The Charpy V-notch impact transition curves are shown in fig. 3 for U steel and fig. 4 for T steel. Transition temperatures and hardness values are listed in Table 4. The metallographic structures corresponding to each condition are shown in fig. 5 for U steel and fig. 6 for T steel.

Material	Condition	Transition Temperature		Hardness HV5
		20 ft. lbs.	50 % Crystallinity	
U Steel	As received	$-23^{\circ}C$	$+4^{\circ}C$	132
	Spheroidised HAZ	$-17^{\circ}C$	$-1^{\circ}C$	150
	Intercritical HAZ	$-30^{\circ}C$	$+1^{\circ}C$	152
	Fine grained HAZ	$-110^{\circ}C$	$-44^{\circ}C$	162
	Coarse grained HAZ	$-27^{\circ}C$	$+24^{\circ}C$	208
T Steel	As received	$-70^{\circ}C$	$-68^{\circ}C$	152
	Spheroidised HAZ	$-69^{\circ}C$	$-62^{\circ}C$	175
	Intercritical HAZ	$-67^{\circ}C$	$-55^{\circ}C$	182
	Fine grained HAZ	$-98^{\circ}C$	$-89^{\circ}C$	187
	Coarse grained HAZ	$+15^{\circ}C$	$+14^{\circ}C$	245

Table 4. Transition temperatures and hardness values of U and T steels.

In U steel the Charpy curves for the weld HAZ were similar to that of the parent material with the exception of the fine grained HAZ in which a much lower transition temperature was recorded. This was attributed to the beneficial effect of grain refinement produced in this region as shown in fig. 5 (d).

In T steel there was also a lowering of the transition temperature in the fine grained HAZ although not to the same extent as in U steel. Fig. 6 (d) shows that this was due to a further refinement of the grain size. However the most significant result is the considerable increase in transition temperature in the coarse grained HAZ, where the transition temperature is about $80^{\circ}C$ higher than in the parent material. In fact in this region T steel has poorer notch-impact properties than U steel although in the 'as received' condition T steel is vastly superior. Fig. 6 (f) shows that a mixed structure of high temperature transformation products and upper bainite is present in this structure and it is the presence of the upper bainite which is thought to be responsible for the poor notch-impact properties of this region.

An interesting feature of the Charpy results is the lower ductile shelf energy 75 J (55 ft. lbf.) of the aluminium treated steel compared with that of the semi-killed steel, 122 J (90 ft. lbf.). In both steels the test specimens were machined with their length parallel to the rolling direction and notched through the thickness. The most likely explanation for this is that the higher C and Mn contents of the aluminium treated steel have given rise to more pearlite and additional substitutional solid solution hardening of the ferrite than in the semi-killed steel. Both these factors would reduce the energy requirement for crack initiation and crack propagation and hence lower the ductile shelf energy.

3.2 THE SUBCRITICAL WELD HAZ

The Charpy V-notch impact results for the subcritical HAZ are shown in fig. 7 for U steel and fig. 8 for T steel. For convenience the result of all specimens notched before simulation are represented as a scatter band. The results show no marked variation between the different conditions in either steel. However, a closer examination of the energy-temperature curve for U steel does indicate a distinct trend for the specimens notched before simulation to have a 20 ft. lbf. transition temperature some 10 - $35^{\circ}C$ higher than the specimen notched after simulation although there is a good deal of scatter in the results. No such trend is discernible in T steel. Hardness surveys on specimens notched before simulation failed to detect any higher levels of hardness in the region of the notch root in either steel.

4. DISCUSSION

4.1 THE VISIBLE WELD HAZ

The results of the investigation have shown that in the semi-killed C - Mn steel the notch-impact properties of the visible weld HAZ are as good as, or better than, those of the parent material for the conditions used here, i.e. submerged arc welding at a heat input level of 4.2 KJ/mm (108 KJ/in.). The spheroidised and intercritical HAZ structures shown in figs. 5(b) and 5(c) respectively show that the ferrite has undergone little or no transformation and that transformation has been restricted to the pearlite areas. Since the increase in hardness is very slight (see Table 4) it is not surprising that the notch-impact curves for these structures are very similar to those of the parent material. In the fine grained HAZ, fig. 5(d), a considerable degree of refinement of the structure occurred and since there is only a minor increase in hardness the effect on the notch-impact properties has been to lower the transition temperature by 50 - 80°C. Contrary to this, Saunders, (Ref. 10), who measured C.O.D. (crack opening displacement) transition temperatures on specimens cut from an actual weld made at a heat input level of 1.0 KJ/mm (26 KJ/in.) in the same steel, found an increase in transition temperature of 19°C in this region. This discrepancy may arise from the different methods used for measuring transition temperatures, or from the different experimental techniques used in preparing specimens with notches positioned in the fine HAZ. The improvement in notch-impact properties measured in the present investigation is more consistent with the hardness and microstructure since one would expect the grain size effect to predominate and thus give rise to better notch-impact properties.

In the coarse grained HAZ the energy-temperature curve is similar to that of the parent material although there is an increase of about 20°C in the 50% crystallinity transition temperature. This agrees well with the results of Saunders (Ref. 10) for the coarse grained HAZ of this steel, where an increase of 10°C in the C.O.D. transition temperature was reported. Despite a moderate degree of hardening in this region the fine dispersion of carbides in the form of high temperature transformation products maintained the notch-impact properties at a similar level to the parent material.

In contrast to the semi-killed C - Mn steel, the aluminium grain-refined steel showed considerable degradation of notch-impact properties in the coarse grained HAZ, although in the other regions the notch-impact properties were similar or better than those of the parent material.

The structures of the spheroidised and intercritical HAZ shown in figs. 6(b) and 6(c) respectively are very similar and indicate that some degree of grain-refinement of the parent material occurred. This factor would favour an improvement in notch-impact properties but this has probably been balanced by the increase in hardness (see Table 4) so that the notch-impact curves are similar to those of the parent material.

A further refinement in structure occurred in the fine grained HAZ as shown in fig. 6(d) with little further increase in hardness so that the transition temperature has been reduced by 20 - 30°C. This also conflicts with the results of Saunders (Ref. 10) for the fine grained HAZ in this steel where an increase in the C.O.D. transition temperature of 82°C was reported. Once again the hardness and microstructure are more consistent with a reduction in transition temperature.

The degradation in notch-impact properties in the coarse grained HAZ was attributed to the formation of a mixed structure of high temperature transformation products and upper bainite as shown in fig. 6(f). The hardness was considerably higher than in the other structures examined which is probably due to the fairly rapid cooling from a higher austenitising temperature and is consistent with the high carbon and manganese contents of this steel (see Table 1). These results agree with those of Saunders (Ref. 10) who measured an increase in the C.O.D. transition temperature of 107^{o}C for this region of the HAZ.

It is interesting to note that although the aluminium grain-refined steel possessed much better notch-impact properties than the semi-killed C - Mn steel in the unwelded condition, the aluminium grain-refined steel had slightly poorer notch-impact properties in the coarse grained HAZ. This is probably due to the higher carbon and manganese contents of the aluminium grain-refined steel which would give a higher hardenability. This is born out by a comparison of the hardness values of the two structures (see Table 4) and also by the presence of lower temperature transformation products in the coarse grained HAZ of the aluminium grain-refined steel. It would thus appear that if an aluminium grain-refined steel were used with the same carbon and manganese contents as in the semi-killed C - Mn steel, the coarse grained HAZ should have comparable notch-impact properties to the parent material since Irvine et al (Ref. 1) have shown that aluminium produces no dispersion hardening even after a high temperature austenitising treatment.

4.2 THE SUBCRITICAL WELD HAZ

The degree of subcritical weld HAZ embrittlement frequently reported in the literature on C - Mn steels was not realised in the present work. There was, however, a distinct trend in the semi-killed C - Mn for the transition temperature to be increased by about 10 - 35^{o}C when subjected to a subcritical weld HAZ thermal cycle in the presence of a notch. The effect was absent in the aluminium grain-refined steel. Changes of this order of magnitude are rather meaningless when the inherent scatter in the Charpy test is considered although there is some agreement in these trends with the results of Saunders (Ref. 10) who used a technique of preparing welds with and without saw cut notches positioned at different depths in the plate and extracting C.O.D. specimens after the weld had cooled. This work showed an increase in the C.O.D. transition temperature of 10 - 15^{o}C in both steels in the absence o of notches. With a notch present before welding the transition temperature was increased by 40^{o}C *for the semi-killed C - Mn steel and* 25^{o}C *for the aluminium grain-refined steel.* This is consistent with the higher free nitrogen content of the former which would enhance the degree of dynamic strain-ageing occurring during welding. Dolby and Saunders (Ref. 15) have correlated this subcritical weld HAZ embrittlement with variations in hardness and dislocation density in the regions undergoing dynamic strain-ageing. The same workers showed from micro-hardness-strain calibration experiments that these embrittled regions correspond to strains of greater than 20%.

The failure to detect appreciable embrittlement or increases in hardness in the subcritical weld HAZ in the present investigation is probably due to the low degree of restraint imposed on a simulated specimen during the weld thermal cycle. Although the specimens were held rigidly in clamping blocks during the weld thermal cycles to restrain expansion and contraction, they were isolated from the restraining influence of surrounding material which occurs in the actual welding situation so that the long range stresses which can build up in an actual weldment could not do so using this simulation technique. It is, therefore, tentatively suggested that unless a stress can be applied to a simulated specimen during thermal cycling, then the strains produced in the subcritical weld HAZ will be grossly underestimated, in which case the technique is not suitable for assessing subcritical weld HAZ embrittlement.

The position is probably not so critical in investigations of the visible weld HAZ by simulation techniques since the high temperatures involved are probably sufficient to anneal out any strain-ageing that occurs. Even if some strain-ageing does occur the effect is likely to be swamped by the changes due to metallurgical transformations. The agreement between the present simulation results and the results of Saunders (Ref. 10) on specimens taken from actual weldments in the two steels considered here are encouraging and it is felt that the simulation technique is still an excellent method of examining the visible weld HAZ.

5. CONCLUSIONS

(i) The visible weld HAZ of a semi-killed C - Mn steel was shown to have notch-impact properties comparable to, or better than, the parent material.

(ii) The aluminium grain-refined steel suffered marked degradation of notch-impact properties in the coarse grained HAZ due to the formation of upper bainite in the microstructure and this was attributed to the high carbon and manganese contents of the steel. The other regions of the visible weld HAZ had notch-impact properties comparable to the parent material.

(iii) In the subcritical weld HAZ the semi-killed C - Mn steel showed a tendency to embrittlement and this was associated with the presence of free nitrogen. No tendency to embrittlement in the subcritical weld HAZ of the aluminium grain-refined steel was observed.

(iv) It is suggested that the simulation technique grossly underestimates the strains present in the subcritical weld HAZ in an actual weldment and is thus considered not to be a good method of assessing embrittlement in this region.

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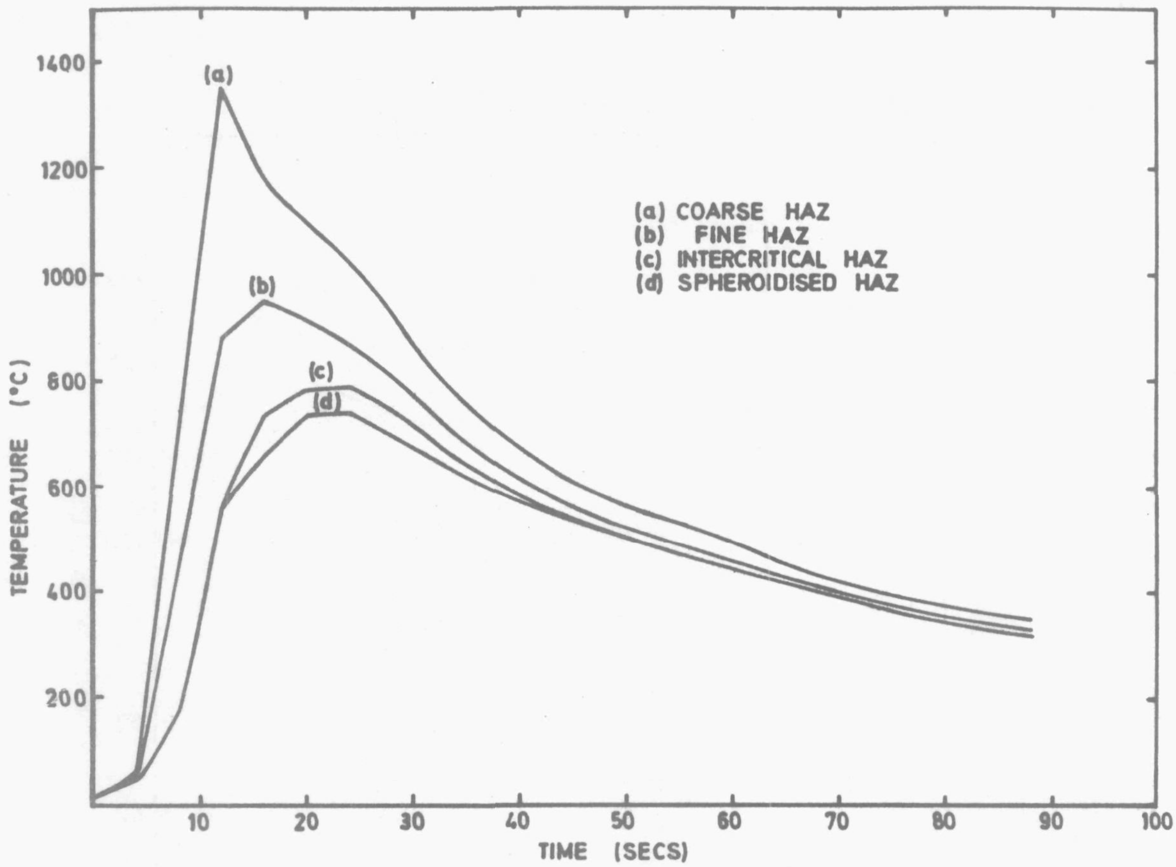


FIG. 1 THERMAL CYCLES PRODUCED IN THE WELD HAZ FOR A HEAT INPUT OF 4·2 KJ./mm (108 KJ./IN.)

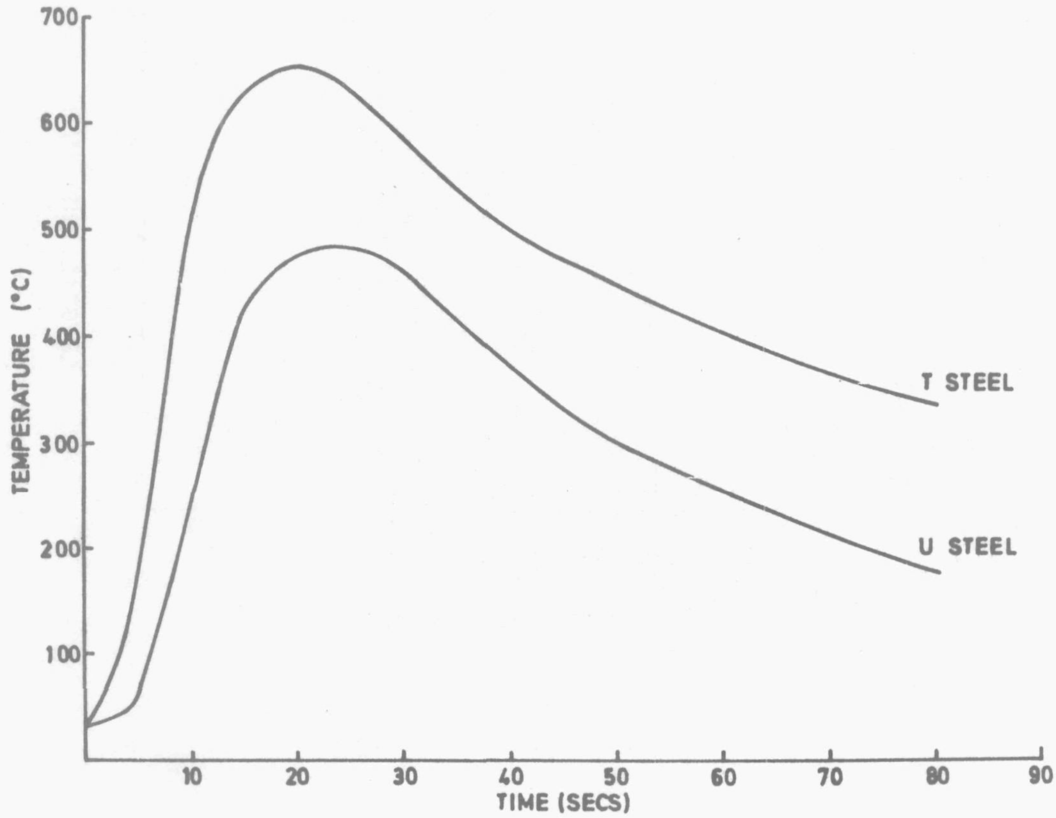


FIG. 2 THERMAL CYCLES PRODUCED IN THE SUBCRITICAL WELD HAZ FOR A HEAT INPUT OF 4·2 KJ/mm. (108 KJ/ in.)

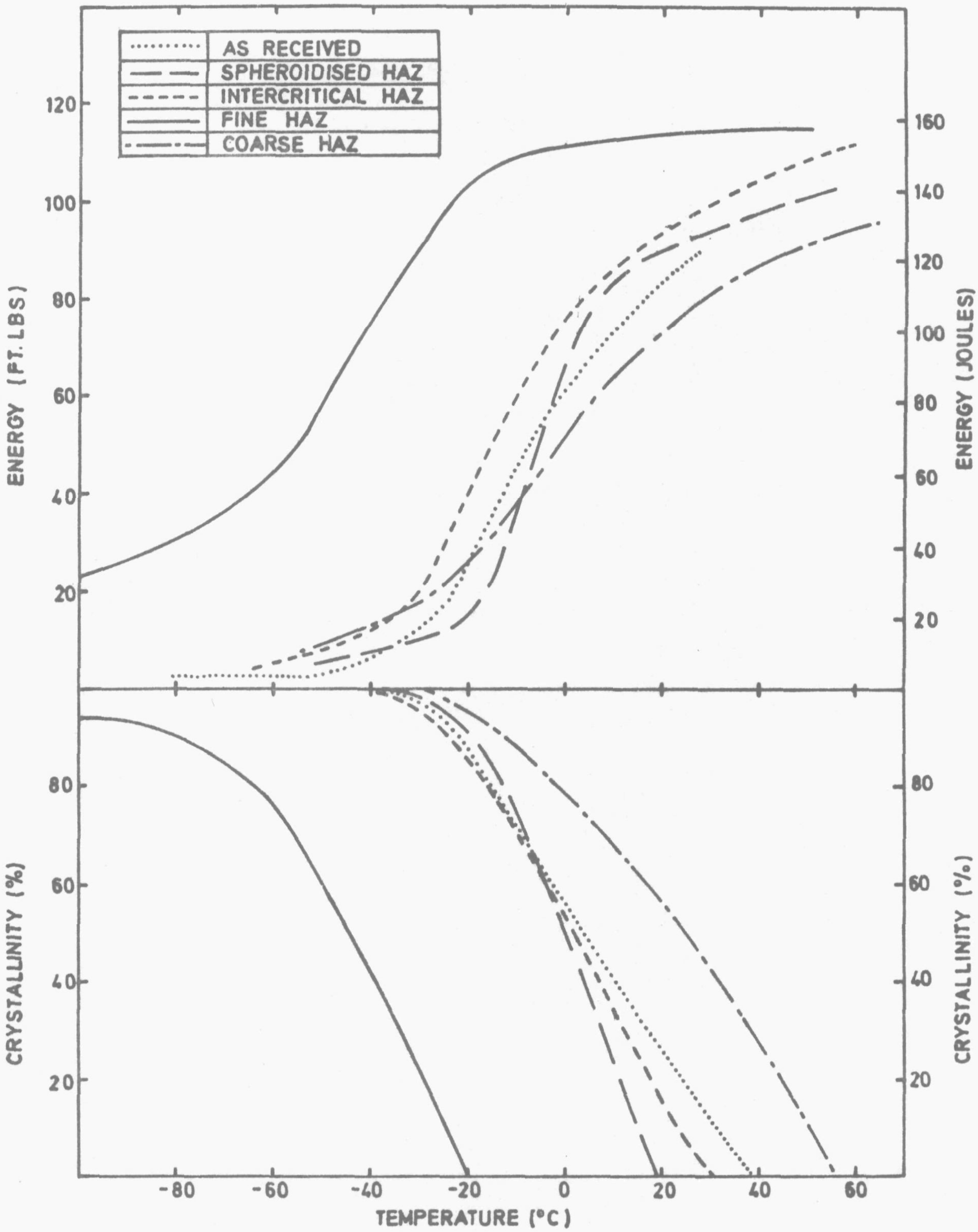


FIG. 3 CHARPY V-NOTCH IMPACT RESULTS FOR U STEEL

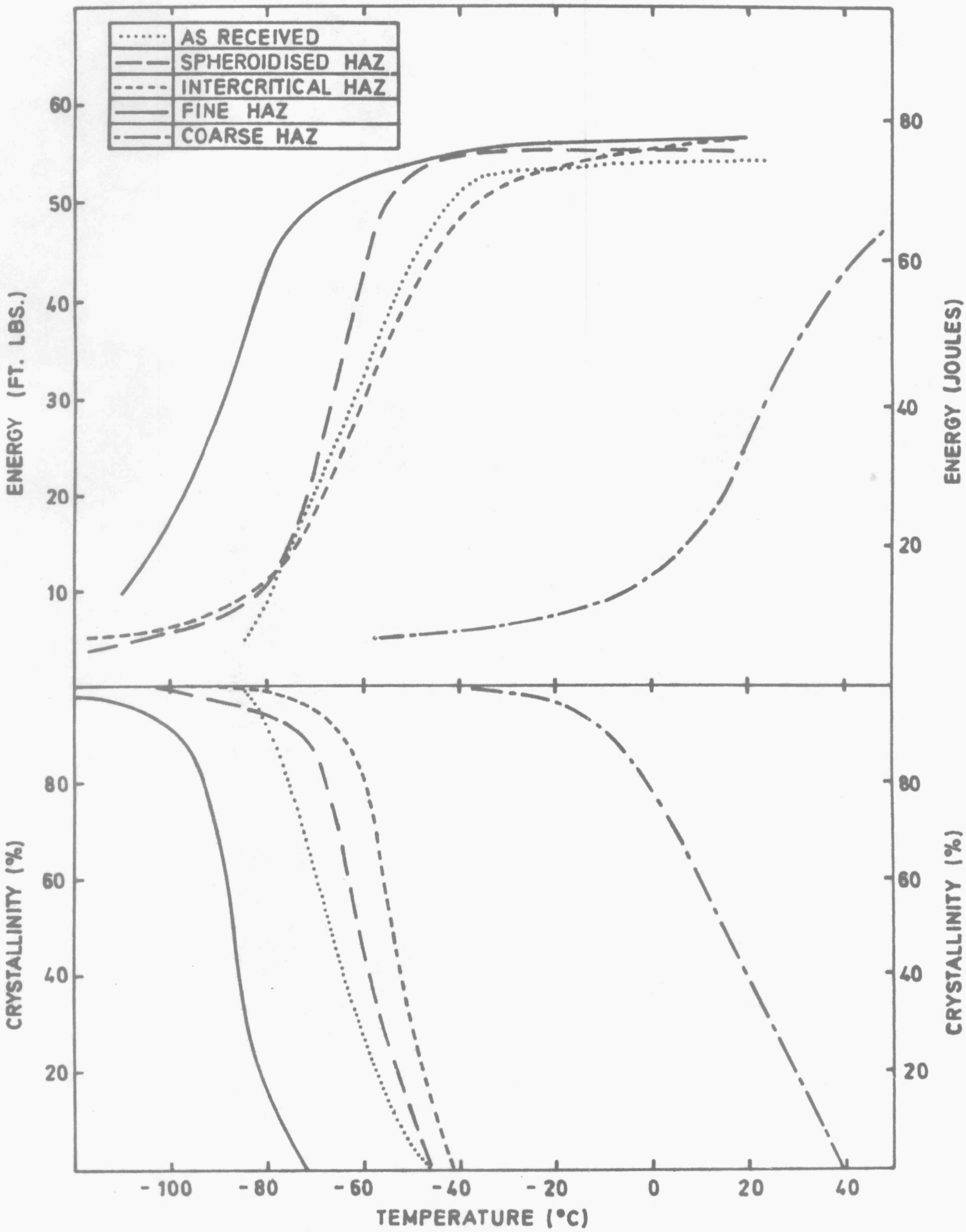
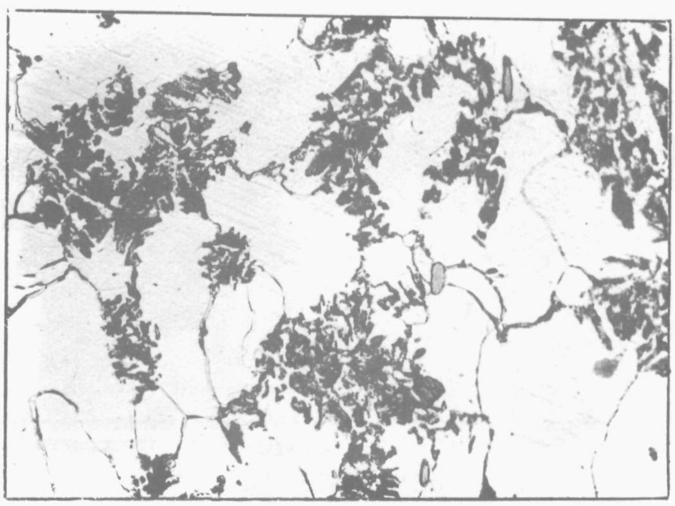


FIG.4 CHARPY V-NOTCH IMPACT RESULTS FOR T STEEL

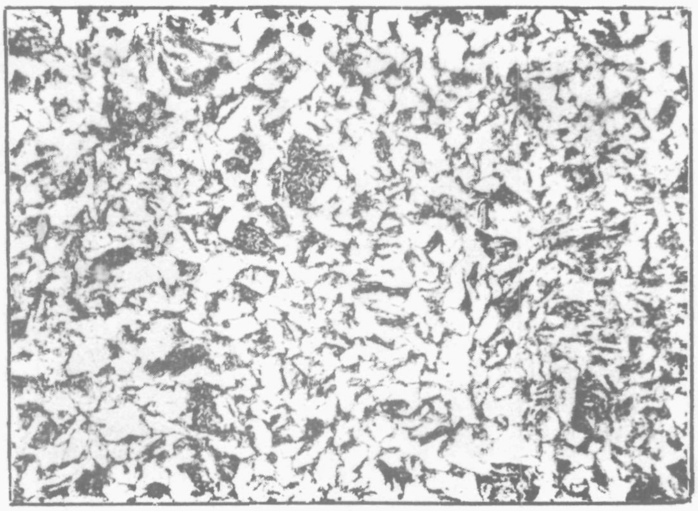
(a) As received x 500.

(b) Spheroidised HAZ x 500.

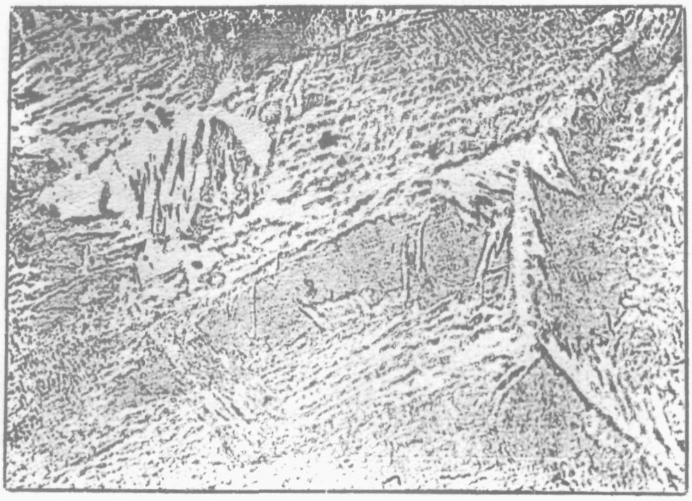


(c) Intercritical HAZ x 500.

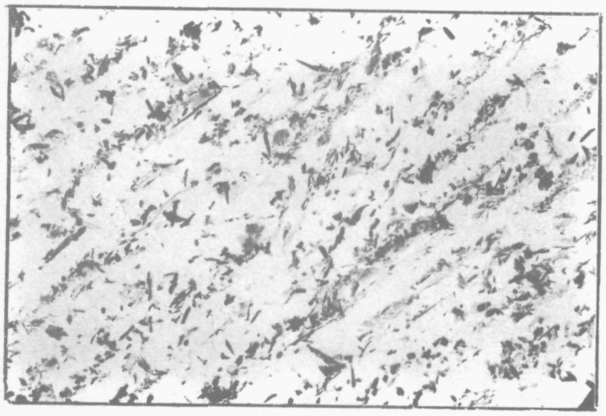
Fig. 5 Metallographic structures in U steel.



(d) Fine HAZ x 500.

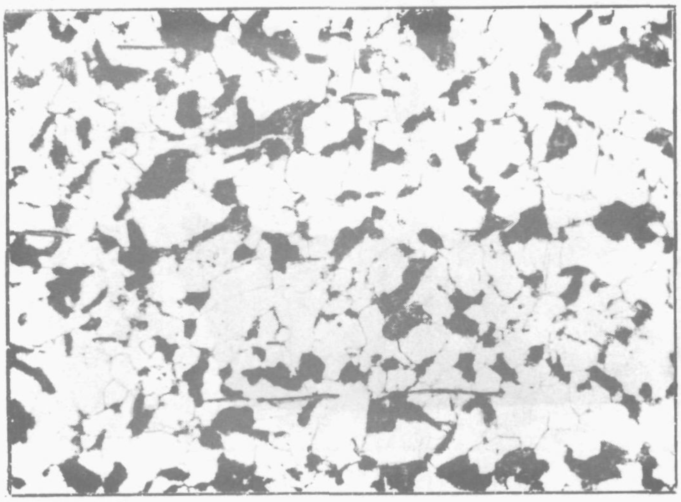


(e) Coarse HAZ x 500.

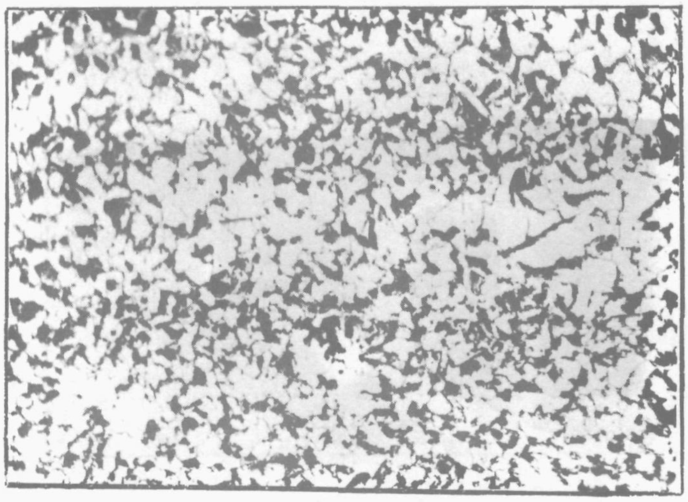


(f) Coarse HAZ, carbon extraction replica x 3000.

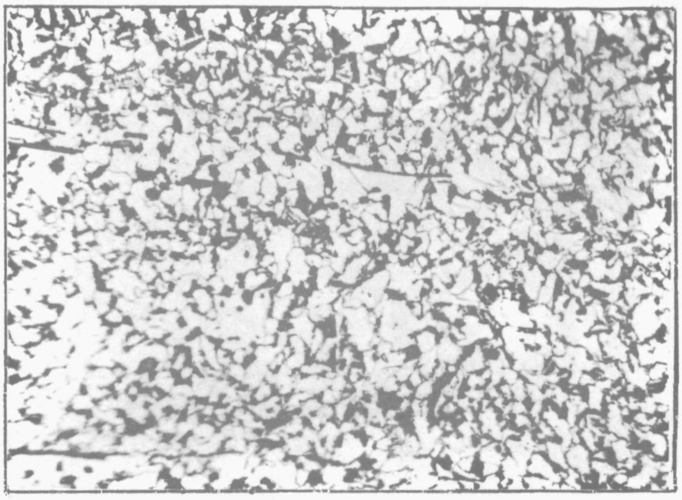
Fig. 5 Continued.



(a) As received x 500.

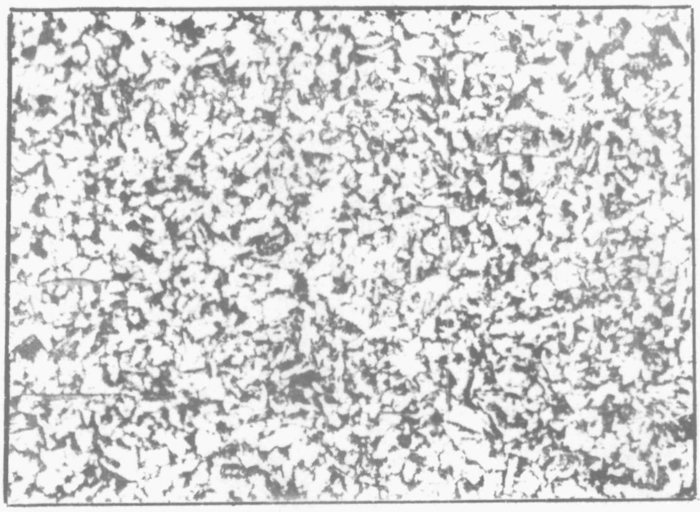


(b) Spheroidised HAZ x 500.

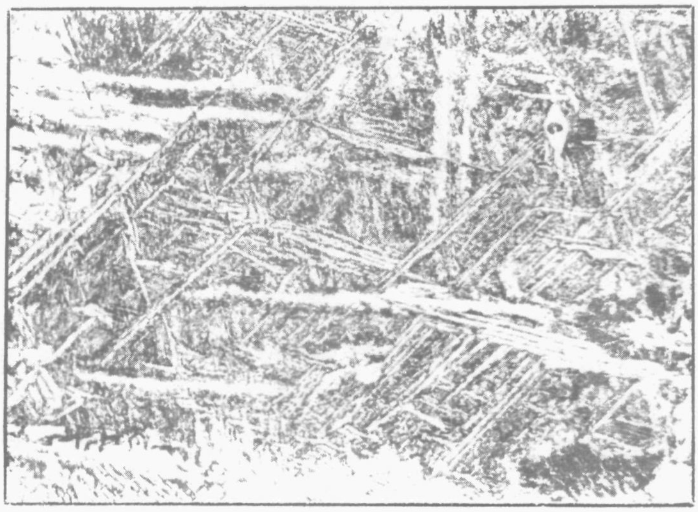


(c) Intercritical HAZ x 500.

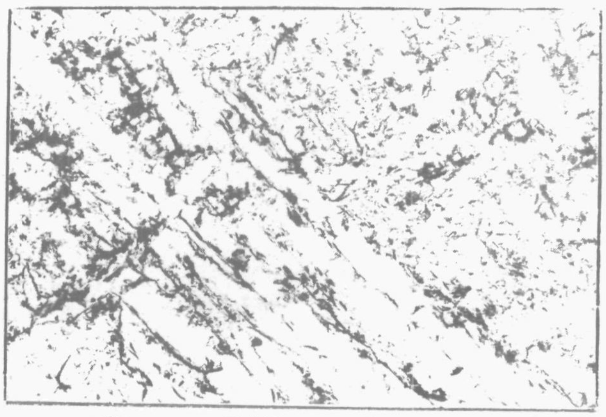
Fig. 6 Metallographic structures in T steel.



(d) Fine HAZ x 500.



(e) Coarse HAZ x 500.



(f) Coarse HAZ, carbon extraction replica x 3000.

Fig. 6 Continued.

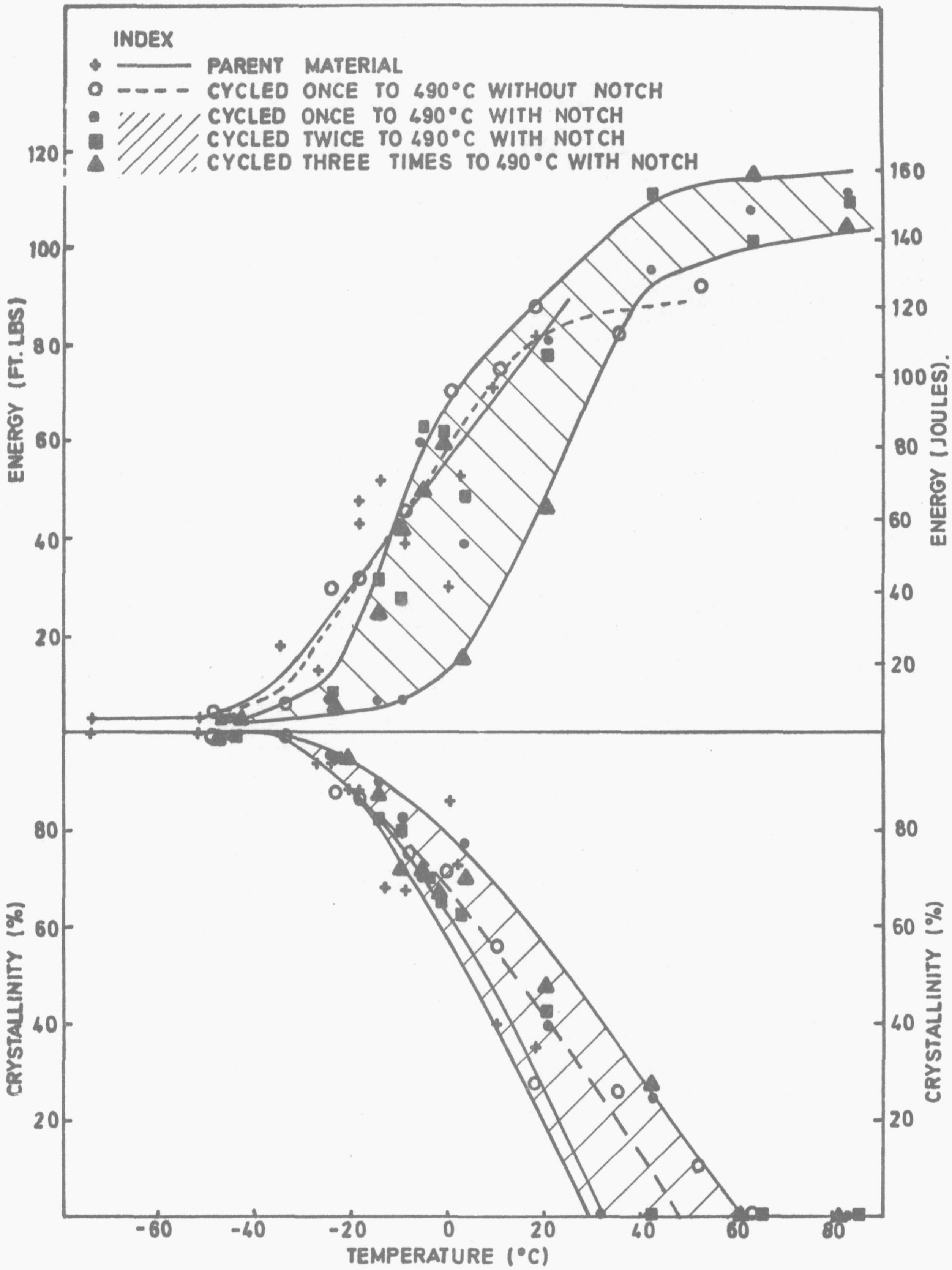


FIG.7 CHARPY V-NOTCH IMPACT RESULTS FOR THE SUBCRITICAL HAZ IN U STEEL.

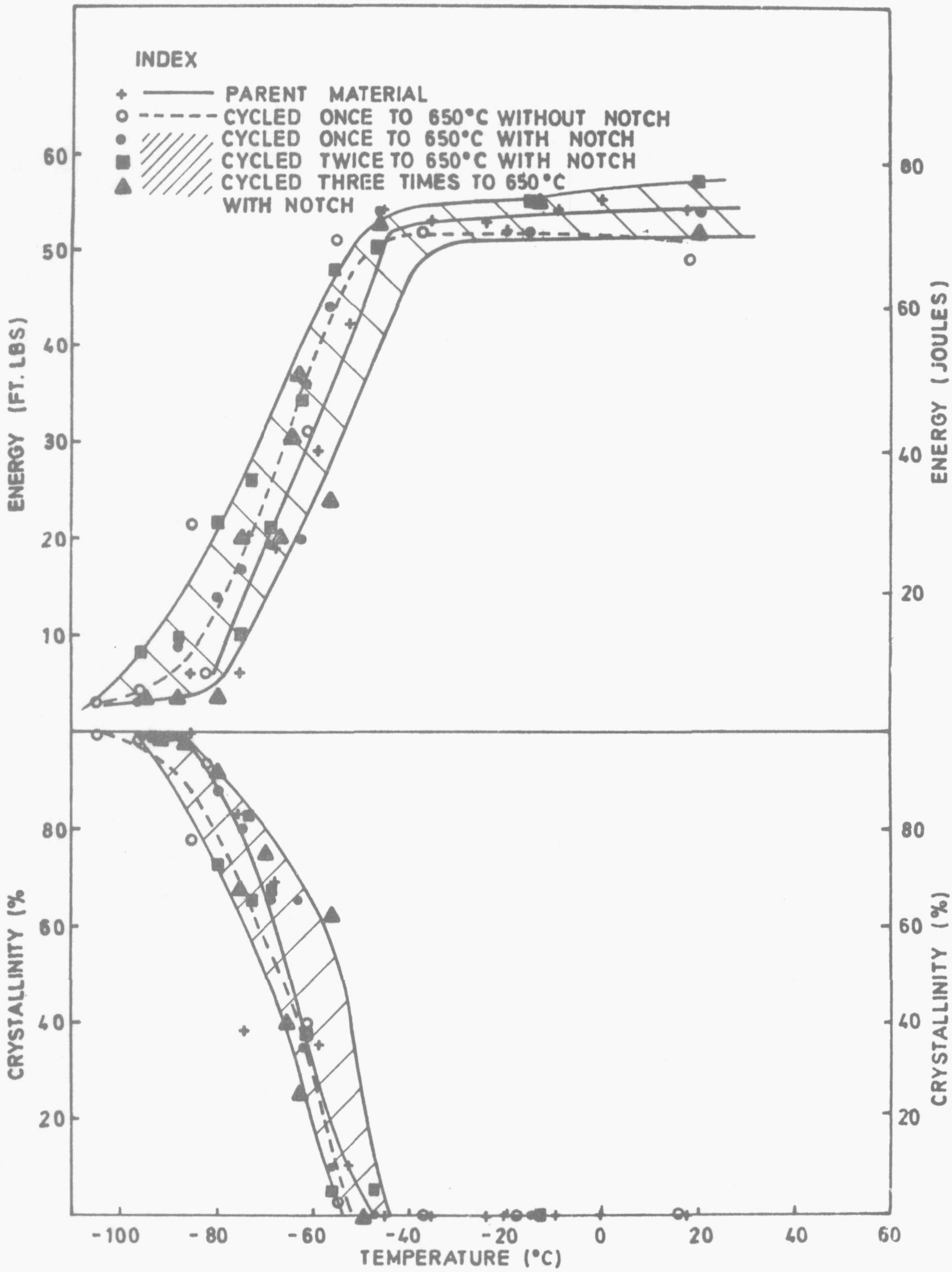


FIG. 8 CHARPY V-NOTCH IMPACT RESULTS FOR THE SUBCRITICAL HAZ IN T STEEL.