The HAZ Impact Toughness of a Mn- Mo-Nb Linepipe Steel

Simulated weld heat-affected zones exhibit good crack initiation resistance, although toughness of the grain coarsened region of the HAZ is not improved by a second thermal cycle, i.e., subsequent weld passes

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ABSTRACT. One of the newly developed high strength low alloy steels (low carbon Mn-Mo-Nb) for gas and oil pipeline applications in arctic regions was investigated to determine the HAZ impact properties. Results of the synthetic specimen technique (Gleeble) showed that the grain refined region of the HAZ has excellent toughness at test temperatures of as low as −60 F, with the upper shelf extending to about −25 F. As was anticipated, the grain coarsened region showed inferior toughness.

However, even at 10% shear fracture appearance the impact energy absorbed was about 50 ft-lb, indicating good crack initiation resistance. The 50% FATT values were −80 F for the 1150 F and the 1600 F peak temperature cycles. For 1350 and 2400 F thermal cycles, the 50% FATT values were 0 and 70 F, respectively. The results of this investigation also showed that beneficial effects are not realized from a second weld thermal cycle (1600 F peak) superimposed on the grain coarsened region produced by a thermal cycle of 2400 F peak temperature. The impact energies absorbed at −10 F were 140, 25, 170, 50, and 160 ft-lb, for the base metal at 2400, 1600, 1350, and 1150 F peak temperatures, respectively.

Introduction

Since the early 1960's there has been a significant commercial effort in the United States and Europe for development of hot-rolled, high strength, low alloy (HSLA) steels with superior mechanical properties. In particular, the goal has been to attain the desirable high strength level with little or no compromise in impact properties. Moreover, due to economic considerations, it is also desirable to produce these structural steels with minimum processing and/or alloying conditions.

The class of hot-rolled, high strength, low alloy steels which best satisfies these requirements is commonly known as grain refined carbon-manganese steels (C-Mn steels). The grain refinement can be accomplished with a variety of alloying elements—namely, aluminum, vanadium, titanium, or niobium.* In addition, an interplay with the hot-rolling practice can also result in grain refinement. Recent investigators have shown that niobium bearing C-Mn steels are probably the most economical of this particular class of commercial structural steels.[1,2]

Until the late 1960's, the existing pipeline steels appeared to offer adequate weldability, strength, and impact resistant properties for the intended end use. However, the discovery of the Prudhoe Bay gas and oil fields in 1969, and a similar discovery of gas fields in the Canadian Arctic Islands, created a new and challenging demand for pipeline steel with optimum properties and stringent specifications. The severe arctic environment and sub-zero temperatures necessitate decreased susceptibility to brittle failure, higher resistance to crack propagation, and improved field weldability as compared to existing pipeline steels. Furthermore, the arctic regions are rather distant from the centers of pipe manufacturing, and by utilizing higher strength grades the pipe section thickness could be reduced, resulting in attractive savings in material, transportation, and welding costs.

In an effort to meet the challenge of this new demand, several companies embarked on development programs to produce improved steels for arctic pipeline usage. The combined efforts of several years of laboratory and industrial research resulted in the commercial introduction in 1972 of a new grade of HSLA steel, commonly referred to as low carbon Mn-Mo-Nb steel. As far as the pipeline application is concerned, the impact properties of the heat-affected zone (HAZ) are of paramount importance because field experience has shown that the fracture path can be parallel to the weld bead.

The object of this research program was to measure and document the impact properties, the transition tem-

**The other name for this element, Columbium (Cb), is widely used in the U.S.*

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Table 1—Composition of Low Carbon Mn-Mo-Nb Steel[a]

Chemical analysis (wt.%)	
C—0.046	Ni—0.03
Mn—2.1	Cu—0.77
P—0.005	Cr—0.05
S—0.003	Al—0.07
Si—0.07	Co—0.011
Mo—0.29	N—0.0124
Nb—0.07	

(a) Armco Steel Heat 51850.

Table 2—Mechanical Properties of Low Carbon Mn-Mo-Nb Steel in As-Rolled Condition[a]

Mechanical property	Climax API-2 in. ga. length, transverse	Armco ASTM-8 in. ga. length, transverse
Yield strength, psi	73,200[b]	85,400-84,900[c]
Tensile strength, psi	103,700	104,000-103,200
Y/T ratio	0.71	0.82-0.82
Elongation, %	37	20-18
Reduction of area, %	60	61-62

(a) Thickness, 0.667 in.; Armco Steel Heat 51850.
(b) 0.5% strain.
(c) Duplicated testing.

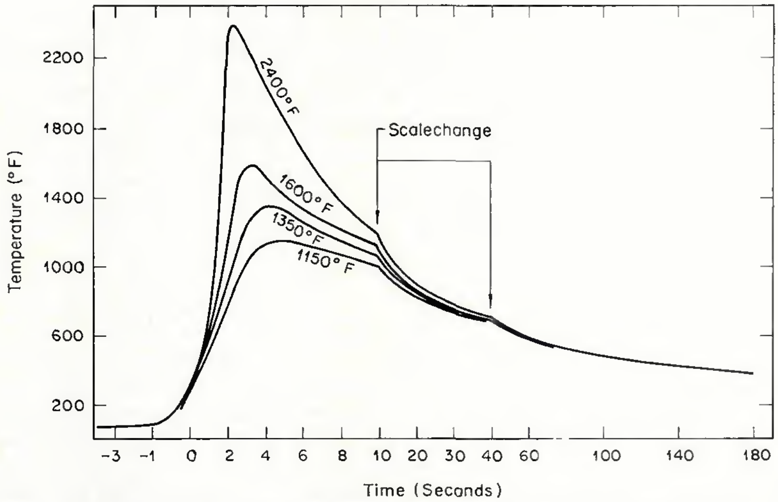

Fig. 1—Weld thermal cycles selected to correspond to heat input of 40 kJ/in., 4 in. plate, no preheat

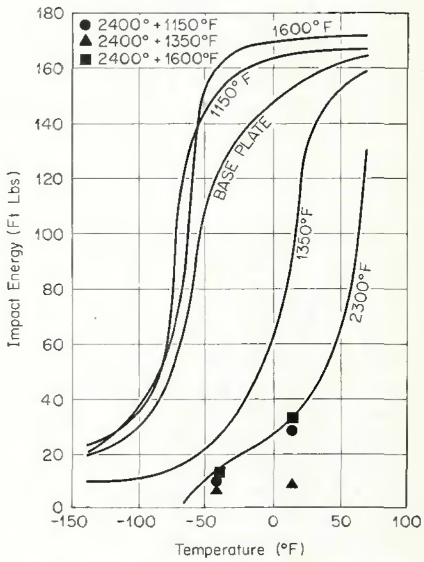

Fig. 2—Charpy V-notch impact energy vs. test temperature for HAZ and base metal (peak temperature of HAZ thermal cycle as indicated). Solid points represent double thermal cycled specimens

perature behavior, and the fracture characteristics in the weld HAZ region of one of these newly developed low carbon Mn-Mo-Nb steels.

Procedure

Material

The chemical composition and the mechanical properties of the heat of steel used in this investigation are shown in Tables 1 and 2.

Specimen Preparation

The Gleeble specimens were of longitudinal orientation (long axis of the blanks was parallel to the rolling direction). The tests blanks were thermal cycled using the Gleeble technique described by Savage.[3] The notch was subsequently machined perpendicular to the original plate surface.

The welding parameters for the development of the four desired weld thermal cycles (2400, 1600, 1350, and 1150 F peak temperatures)** were selected to correspond to ½ in. plate with 40 kJ/in. (1575 kJ/m) energy input and no preheat. These conditions approximate girth welding parameters for pipeline welding. The four weld thermal cycles are shown in Fig. 1.

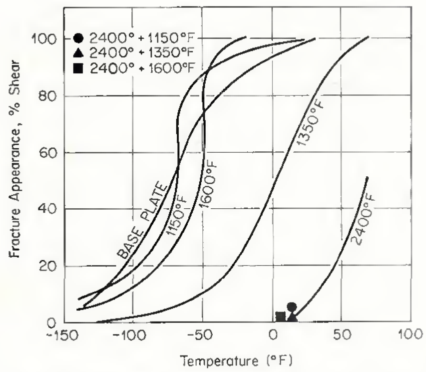

Fig. 3—Fracture appearance vs. impact test temperature for HAZ and base metal (peak temperature of HAZ thermal cycles as indicated). Solid points represent double thermal cycled specimens

Metallography, Fractography, and X-ray Analysis

Longitudinal and transverse cross sections of the base metal and synthetic HAZ specimens were examined metallographically. The etchant used for revealing the microstructure was 2% nital. Other etchants, such as 10% ammonium persulfate, picral, and super picral (picral plus 5% HCl) were also investigated. The most appropriate etchant for this material was determined to be nital.

The fracture surfaces were examined with the scanning electron microscope (qualitative X-ray analysis was obtained with an ORTEC energy dispersive X-ray analyzer).

Results and Discussion

Material Toughness

Figures 2 and 3 illustrate the results of the Charpy impact testing. They show the impact behavior of both the HAZ and the unaffected base metal as defined by energy absorbed and by the fracture surface appearance, respectively. The transition temperatures of the four HAZ regions and that of the base metal are shown in Fig. 4. As indicated in Table 3, the criteria for the evaluation of the transition temperature in this study are as follows:

1. That temperature at which the fractured specimens show approxi-

**To convert F degrees to C degrees, subtract 32 and multiply by 5/9.

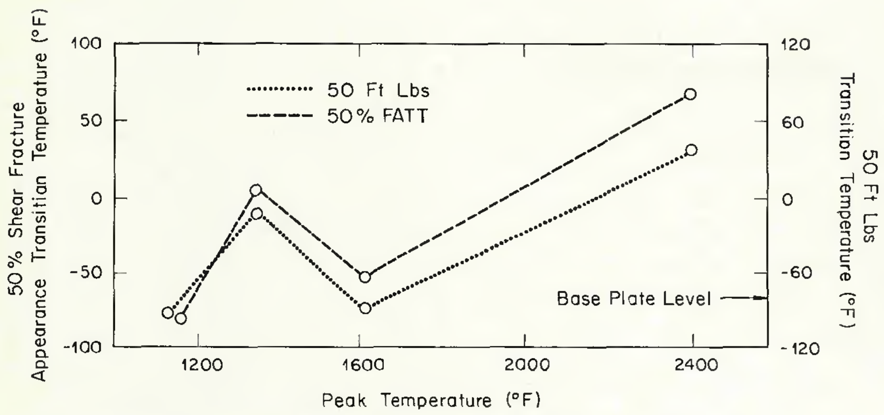

Fig. 4—Transition temperature vs. peak temperature of HAZ thermal cycle

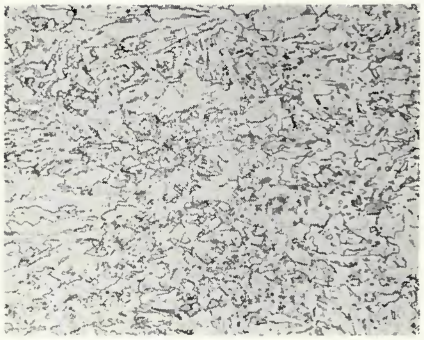
Fig. 5—As-received plate, longitudinal section. 2% nital etch. ×500 (reduced 41% upon reproduction)

mately 50% ductile fracture characteristics (shear). This temperature is commonly referred to as 50% fracture appearance transition temperature (FATT)

2. That temperature at which the impact energy absorbed is one-half of the upper shelf value

3. The 50 ft-lb† transition temperature

4. The 15 ft-lb transition temperature

The microstructure of the as-received plate is shown in Fig. 5. The microstructure of the four HAZ regions are shown in Figs. 6 and 7. The hardness and grain size data corresponding to these microstructures are recorded in Table 4.

Table 3—Charpy V-Notch Impact Testing Results, Transition Temperatures

Material	50% FATT,[a] F	½ Upper shelf trans. temp. F	15 ft-lb trans. temp. F	50 ft-lb trans. temp. F
Base plate	−80	−76	−160	−92
2400 F peak temp.	68	40	− 40	32
1600 F peak temp.	−56	−70	−172	−88
1350 F peak temp.	4	10	− 74	−12
1150 F peak temp.	−86	−71	−232	−82

(a) Fracture appearance transition temperature.

Discussion

As noted in Figs. 2 and 3, the results of this study indicate that at 80 F the impact strength of the HAZ is generally comparable to that of the base metal. The impact energies vary from 171 ft-lb for the 1600 F peak temperature (grain refined zone) to 130 ft-lb for the 2400 F peak temperature (grain coarsened zone).

However, impact strength per se is not a totally satisfactory criteria for evaluation of toughness behavior because, as noted in Fig. 3 at the 68 F testing temperature, the fracture of 2400 F thermal cycled specimens showed approximately 50% cleavage characteristics. The percent shear fracture appearance in the transition region is a parameter indicative of the material propensity for arrest of a propagating brittle crack. In the case where the fracture appearance is mostly cleavage (e.g., 90%), the impact energy absorbed is an indication of the ease of brittle crack initiation in a material. This criterion is referred to as "ductility transition temperature", and is generally evaluated at a given percent of brittle fracture surface appearance (normally 90% cleavage).

For example, the absorbed energies, at impact testing temperatures where fracture surfaces exhibit approximately 90% cleavage characteristics, are shown in Fig. 8. As noted, the lowest impact strength is 24 ft-lb, corresponding to the 1150 F peak temperature, and the highest is 51 ft-lb, related to

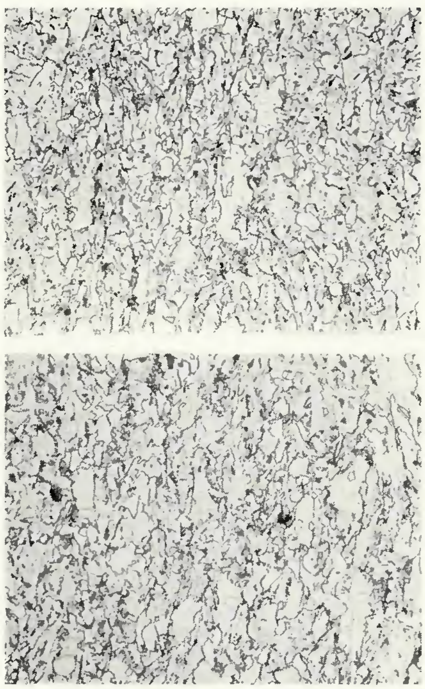
Fig. 6—Microstructure of the synthetic regions, longitudinal section, 1150 F (A—top) and 1350 F (B—bottom) peak temperatures. 2% nital etch. ×500 (reduced 41% upon reproduction)

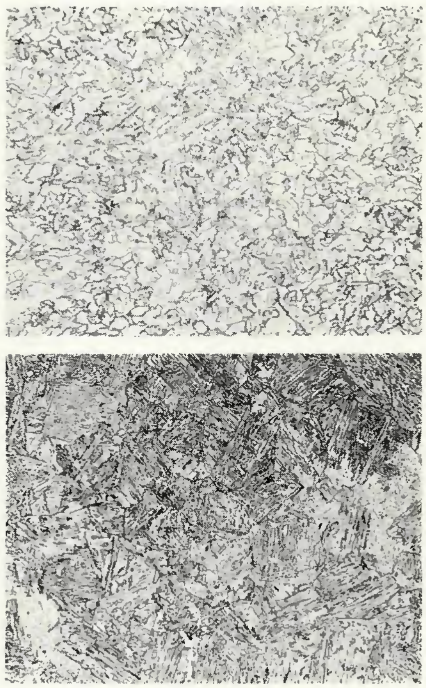
Fig. 7—Microstructure of the synthetic regions, longitudinal section, 1600 F(A—top, ×750) and 2400 F peak (B—bottom, ×250) temperatures. 2% nital etch; reduced 41% upon reproduction

†*To convert foot-pounds to joules, multiply by 1.356.*

Table 4—Microhardness, Grain Size, and Impact Energy

Material condition	Microhardness DPH	ASTM Grain Size No.	−10 F impact energy ft-lb
Base plate	230	11.2	140
2400 F peak temp.	234	—	25
1600 F peak temp.	242	12.3	170
1350 F peak temp.	238	10.9	50
1150 F peak temp.	220	11.4	160
2400 F + 1600 F peak temp.	245	11.2	25
2400 F + 1350 F peak temp.	266	6.0	25
2400 F + 1150 F peak temp.	233	—	25

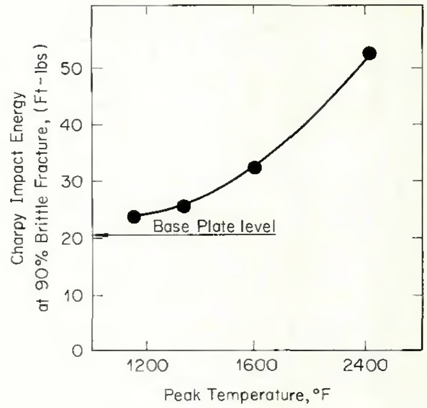

Fig. 8—FATT (10% shear) vs. peak temperature of HAZ thermal cycle

2400 F peak temperature. These values are quite good for high strength, low alloy steel, and clearly suggest that all the regions of the HAZ examined in this study exhibit excellent potential to resist a brittle crack initiaion.

Figures 2 and 3 show that the transition temperature behavior of the 1150 F and 1600 F peak temperature regions in the HAZ are similar or superior to that of the base metal for the entire impact testing temperature spectrum. This observation is clearly demonstrated by the transition temperatures shown in Fig. 4 and Table 3.

In general, four major metallurgical parameters determine the Charpy impact behavior of the HAZ in ferritic-pearlitic steels. These are: (1) microstructural changes (e.g., grain refining or coarsening), (2) precipitation and/or growth of second phase particles, (3) solid solution hardening, and (4) the pearlite fraction in the microstructure. Previous investigators have shown that a quantitative relationship does exist between these parameters and the transition temperature behavior of the ferritic-pearlitic high strength, low alloy steels. The following are some of the relationships which have been determined:[4,5]

$$50\%\ \text{FATT (F)} = 145 + 78\ (\%\ \text{Si}) - 4.2\ (d^{-1/2}) + 4.0\ (\%\ \text{Pearlite}) \qquad (1)$$

$$50\%\ \text{FATT (F)} = 108 - 86\ (\%\ \text{Mn}) + 2700\ (\%\text{N}) + 4.3\ (\%\ \text{Pearlite}) - 3.2\ (d^{-1/2}). \qquad (2)$$

The above equations were developed by various linear regression analysis methods, however, precipitation hardening or overaging phenomena were not directly incorporated in the equations. For the prediction of the transition temperature behavior in low carbon Mn-Mo-Nb steels, the thermal history and therefore magnitude of the precipitation effect of niobium carbonitride is of prime significance. Furthermore, for fine grained ferritic-pearlitic steels (ASTM No. 10-14), the term containing the mean ferrite intercept ($d^{-1/2}$) becomes the dominant term, and this necessitates a very accurate grain size measurement for reproducible and dependable results.

For microstructures resulting from low temperature transformation products (see Figs. 5–7) such measurements are quite difficult to perform. The analysis of the data obtained from this study showed that none of the above equations gave a close approximation of the 50% FATT value. However, it should be noted that the first equation predicted a somewhat reasonable 50% FATT value for the base metal and for the 1150 F peak temperature (see Table 5). Discrepancies are attributed to the following:

1. These regression analyses are generally valid when performed for a specified steel. Since the amount of the alloying elements (for solid solution strengthening) is varied for different steels, the contribution of the solid solution term is also a variable in the equations. For low carbon Mn-Mo-Nb, this analysis has not been performed.

2. If the total amount of niobium in the steel is more than 0.020 w/o, the precipitation hardening effect becomes quite significant in increasing the transition temperature. However, it is not possible to predict the magnitude of this effect from the analysis of the thermal history alone. The kinetics of the reaction in this system are quite complex and not well understood at the present time.

In the final analysis, the regression analysis equations predicting the transition temperatures are not readily applicable to low carbon Mn-Mo-Nb steels. However, the 1150 F peak temperature region of the heat-affected zone has a microstructure and grain size similar to that of the base metal (see Figs. 5 and 6 and Table 4). This is because, with the 1150 F thermal cycle, no transformation has occurred. Furthermore, the rather low peak temperature (1150 F) and short time at temperature (as contrasted to mill practice) precludes a chance of ferritic grain growth. Thus, the grain size, which is the most dominant factor in affecting the transition temperature (see equations above), remains basically unchanged when compared with that of the base metal. Any significant precipitation hardening or solid solution hardening (e.g., dissolution of second phase particles) should also be ruled out because of the rather low temperature. In short, the transition temperature behavior of the HAZ region subjected to 1150 peak temperature should be quite similar to that of the base metal.

Table 5—Transition Temperatures (50% FATT) Calculated vs. Measured

Material	Measured F	Calculated Equ'n 1, F
Base metal	−80	−115
1150 F peak temp.	−86	−128
1350 F peak temp.	4	− 97
1600 F peak temp.	−56	−171

Actually, the 1150 F peak temperature may have slightly superior impact transition behavior when compared with the base metal. This is based upon the decrease in hardness noted for this thermal cycle (220 DPH for 1150 F vs. 230 DPH for the base metal) as shown in Table 4. This decrease in hardness is most likely due to the tempering of the low temperature transformation products present in the as-rolled plate. Transmission electron microscopy has shown patches of martensite in this particular grade of steel.[6]

The impact properties of the 1600 F peak temperature region of the HAZ are also similar to that of the base metal and the 1150 F peak temperature. However, the microstructure of the 1600 F cycle shows a finer grain size than that of the base metal (Fig. 7A). The grain size resulting from the 1600 F cycle is ASTM number 12.3, while that of the base metal is 11.2. Grain refinement has occurred because during the 1600 F thermal cycle, the Ac_3 temperature had been exceeded, and transformation had occurred. Since the initial ferritic structure was fine grained, the resultant austenite which nucleates at the ferrite grain boundaries is fine grained also. Moreover, at 1600 F there is little chance for austenite grain growth and upon cooling, the fine grained austenite transforms to a finer grained ferritic structure.

Ordinarily, grain refinement would improve the impact transition behavior (decreasing the transition temperature). For example, if the mean linear ferrite intercept ($d^{-1/2}$) from Table 4 is applied to equation 1, the predicted 50% FATT should be approximately −171 F. However, the experimental value is −56 F (Table 5). Since Figs. 2 and 3 show that the impact transition temperature behavior of the 1600 F thermal cycled material is similar to the base metal, it is therefore quite apparent that the anticipated improvement from grain refinement has not been fully realized. The reason for this may be due to the following:

1. The grain size of the 1600 F cycle and that of the base metal are ASTM No. 12.3 and ASTM No. 11.2, respectively. This difference in grain size may be rather small for differentiation of impact transition behavior by the Charpy V-notch impact test.
2. Any beneficial affect arising from slight grain refinement may have been offset by the precipitation of niobium carbonitride particles; if these precipitate particles are small (less than 20 angstroms) and coherent with the matrix, they may deleteriously affect the impact properties (as indicated by Gross[7]).

It should be pointed out that the increase in hardness due to the fine precipitate particles may not be detectable because overaging of preexisting precipitate particles has also occurred during the thermal excursion. These simultaneous reactions have been reported to occur in niobium treated carbon-manganese steels. It is interesting to note that the hardness after the 1600 F cycle is 234 DPH, while that resulting from the 1150 F cycle is 220 DPH. This indicates that overaging of the precipitated particles for 1600 F cycle had not been of a sufficient magnitude to compensate for any loss in toughness due to the precipitation of finer particles.

A review of the Charpy impact data shown in Figs. 2-4 reveals that exposure to thermal cycles of 1350 F and 2400 F results in a considerable loss in toughness. Figure 4 indicates that the transition temperature (50% FATT) is 68 F and 4 F for the 2400 F cycle and 1350 F cycle, respectively. These values are significantly higher than the corresponding values for 1600 F cycle (−56 F), 1150 F cycle (−86 F), and the base metal (−80 F).

Examination of the microstructures

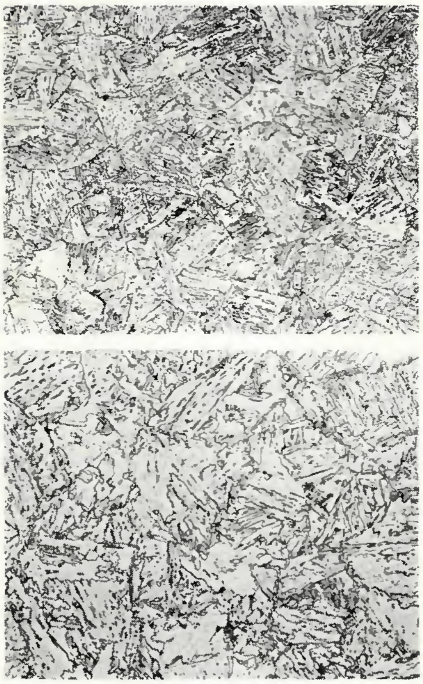

Fig. 9 — Microstructures of the material subjected to double thermal cycles at peak temperatures indicated: A (top)— 2400 and 1150 F; B (bottom)— 2400 and 1350 F. 2% nital etch; ×250, reduced 41% upon reproduction

shown in Figs. 6 and 7 reveals no significant difference in the microstructure or grain size of the 1350 F cycle (nor the 1150 F cycle, for that matter) with that of the base metal. Thus, the loss in impact properties is not attributable to grain size or microstructural changes revealed by optical microscopy. It appears that precipitation of niobium carbonitride has occurred resulting in an increase of the (50% FATT) transition temperature. These results are somewhat similar to those obtained by Aronson[8] in a study performed on a niobium modified high strength low alloy steel with the following compositon:

C-0.065	P-0.016
Mn-1.06	S-0.021
Nb-0.041	Si-0.37

The above composition is similar to

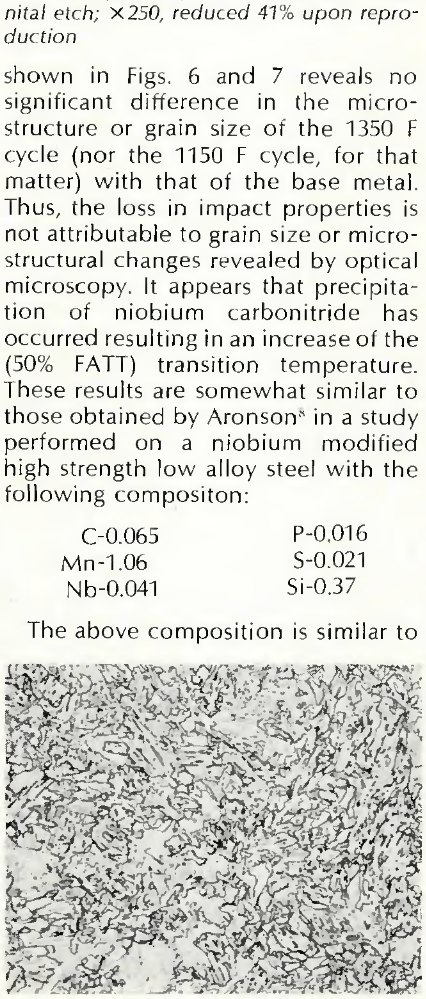

Fig. 10—Microstructure of the material subjected to 2400 + 1600 F peak temperatures. 2% nital etch; ×500 (reduced 41% upon reproduction)

the steel in study, except for silicon and manganese content (refer to Table 1).

As far as the 2400 F thermal cycle is concerned, the loss in toughness is probably primarily due to the grain coarsening and coarse transformation products produced by the high peak temperature. As shown in Fig. 7, the microstructure resulting from this cycle is representative of the type of lower temperature transformation products common in low carbon steel. Less than desirable impact behavior is generally attributed to this type of microstructure.

To evaluate the effect of the subsequent weld passes on the coarse grained region of the HAZ, several 2400 F peak temperature specimens were double cycled (2400 + 1150, 2400 + 1350 F, and 2400 + 1600 F). Figures 2 and 3 show that the superimposing of these thermal cycles did not improve the impact transition properties and actually caused further loss in toughness in the case of the 1350 F cycle. The lack of improvement for the 1150 F cycle is explained by the rather insignificant changes occurring in the microstructure (Fig. 9). Apparently, no precipitation of niobuim carbo-nitride had occurred, as shown by the comparison of hardness values of 2400 F cycle with that of the 2400 + 1150 F cycles in Table 4.

Figure 10 shows that the microstructure of the coarse grained region of the HAZ has been altered to a rather fine grained structure as a result of a subsequent 1600 F cycle. This grain refinement should have considerably improved the toughness; however, the improvement was not realized. Precipitation hardening is probably partially responsible for lack of improvement as shown by hardness values in Table 4 (as a result of the subsequent 1600 F cycle, the hardness of 2400 F cycle had increased from 234 DPH to 245 DPH). However, the primary reason for lack of improvement is not completely clear.

The microhardness, grain size, and impact energy values (at −10 F) shown in Table 4 clearly illustrate that based on the grain size (ASTM No. 11.2) the 2400 + 1600 F should have exhibited impact behavior somewhat similar to the 1600 F thermal cycle. It is possible that the prior thermal history and/or microstructure has had a significant effect on the impact test results. Because relevant data have not been completely documented, this interesting result deserves further investigation. The authors feel that this was one of the most interesting, although unexpected, results of the study. Needless to point out, a conclusive study in this area will be useful in predicting the HAZ impact toughness in multipass welding of this new steel.

Conclusion

The results of this Gleeble study indicate that the impact properties of the HAZ in low carbon Mn-Mo-Nb steel is primarily determined by the grain size of the transformation products and/or precipitation hardening, or tempering. The HAZ impact properties can be separated into two distinct categories as compared with the base metal.

The first category includes two regions of the HAZ which have experienced the 2400 F or the 1350 F thermal cycles. Both of these regions exhibit decreased thoughness with the magnitude of the loss being greater for the 2400 F cycle. The other category is the 1600 F and the 1150 F cycles, which exhibit toughness similar to the base metal. The improvement realized by the 1150 F cycle was due to tempering of isolated areas of low temperature transformation products, and for the 1600 F cycle was due to grain refinement.

Finally, this study shows that the toughness of the grain coarsened region of the HAZ cannot be improved by subsequent weld passes. In fact, the subsequent weld thermal cycle may deleteriously affect the impact properties.

Acknowledgements

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