The Weldability of Some Arctic-Grade Line-Pipe Steels

Micro-alloy line-pipe steels display good resistance to HAZ cracking and, with respect to weldability, are generally superior to conventional line-pipe steels

BY J. GORDINE

ABSTRACT. The weldabilities of nine Arctic-grade steels were determined and compared to the weldability of a conventional X65 line-pipe steel. Two aspects of welding were examined: the susceptibility to hydrogen-induced heat-affected zone (HAZ) cracking and the HAZ Charpy toughness properties.

It was shown that the majority of the Arctic-grade steels exhibited better HAZ cracking resistance than the conventional X65 steel. It was also confirmed that the conventional carbon-equivalent formula is not valid for the new high-strength low-alloy steels and that the Ito-Bessyo relation is a more appropriate measure to use.

The HAZ toughness measurements made on Gleeble simulated specimens showed a wide variation in behavior of the different steels. All showed severe loss in toughness in the 1300 C (2370 F) peak temperature condition. Some steels also showed severe loss in toughness in the 700 C (1290 F) peak temperature condition.

Introduction

Recent oil and gas discoveries in northern Canada have resulted in several proposals for the construction of pipelines to transport these resources to the southern market areas. The property requirements for the pipeline steels that will be used to build these pipelines are very stringent and cannot be met by the conventional types of line-pipe steels.

Because of the Arctic environment, very high demands have been set on both strength and toughness. For example, the latest Canadian Arctic Gas Pipeline proposal specifies a minimum yield strength of 70 ksi (482 MPa) with a Charpy V-notch toughness criterion of 30 ft-lb (40 J) at design temperatures of −25 C (−15 F) or −5 C (+25 F) in the case of pipe of 0.72 in. (18 mm) wall thickness.

To meet these difficult specifications, new steel compositions have been developed and a number of these so-called Arctic-grade steels are now available. While considerable effort has gone into developing the required mechanical properties in these new steels, much less attention has been given to their welding behavior. It was the purpose of the work described in this paper to examine this aspect.

A number of Arctic-grade line-pipe steels were purchased, and weldability tests were conducted upon them. The two aspects of weldability selected for examination were:

1. Susceptibility to HAZ cracking.
2. HAZ toughness characteristics.

Both aspects will play a major role in the ultimate selection of steels for use in any Arctic-pipeline project. Both characteristics are also very much a function of the steel, itself, rather than of the welding consumables used.

HAZ cold cracking is a problem often encountered in the field welding of conventional line-pipe steels and is likely to be an even more serious problem in the Arctic pipeline projects. Due to the higher operating pressures, pipe wall thicknesses are greater, and this will result in faster cooling rates in the HAZ. Thus, the risk of developing hard crack-susceptible microstructures in the HAZ is much greater. In addition, the greater wall thicknesses will mean that the external stress imposed on the weld during the laying of the pipe will be greater, and this will also increase the risk of cracking.

The maintenance of adequate toughness properties in the HAZ is likely to be a serious problem in the welding of these Arctic-grade line-pipe steels. The improved toughness in these steels is largely a result of a fine-grained microstructure produced by carefully controlled rolling techniques. It can be assumed that welding will alter this carefully regulated microstructure and that accompanying these microstructural changes will be changes in toughness. The extent to which the toughness is changed in the HAZ is not known. It is clear, however, that most Arctic specifications will require minimum toughness levels in the HAZ.

Materials

A general description of the 10 line-pipe steels in the evaluation program is given in Table 1. The chemical compositions and some representative mechanical properties are given in Tables 2 and 3.

All pipes were purchased in the form of a pipe length and all except one (pipe A) were made to defined API specifications using normal commercial practices. The steels listed

Paper presented at the 58th AWS Annual Meeting held in Philadelphia, Pennsylvania, during April 25-29, 1977.

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represent producers from Canada, Japan and West Germany. All, except pipe F, were considered candidate materials for Arctic pipeline applications.

The line-pipe steels in the program can be separated into the following three categories.

1. Conventional line-pipe steel—steel F (included for comparative purposes).
2. Controlled rolled micro-alloy line-pipe steels—steels J, M, N, P, R and Z.
3. Quenched and tempered steels—steels T and S.

Table 1—General Description of Line Pipes

Pipe	Size, in.[a] (O.D. × wall thickness)	API specification	Processing
A	30 × 0.375	Experimental	As-rolled and aged 1 h at 700 C (1300 F)
F	26 × 0.518	5LX-X65	Controlled-rolled
J	48 × 0.562	5LX-X65	Controlled-rolled
M	42 × 0.380	5LX-X70	Controlled-rolled; spiral weld
N	42 × 0.420	5LX-X65	Controlled-rolled
P	48 × 0.720	5LX-X70	Controlled-rolled
R	48 × 0.720	5LX-X70	Controlled-rolled
S	48 × 1.250	5LX-X65	Quenched and tempered
T	48 × 1.250	5LX-X65	Quenched and tempered
Z	42 × 0.540	5LX-X70	Controlled-rolled; spiral weld

[a]1 in. = 25.4 mm.

Experimental Procedures

Assessment of Cold Cracking Susceptibility

Two test methods were used to determine susceptibility to hydrogen-induced HAZ cracking. They were:

1. Controlled Thermal Severity (CTS) testing.
2. Implant testing.

Both are standard weldability tests used to determine the susceptibility of a steel to HAZ cold cracking, and full descriptions are available in the literature.[1,2]

The test welding conditions selected for both tests were based on those used for manual girth welding in pipelines.[3] They are summarized in Table 4. The energy inputs used were slightly higher than are normally used in girth welding, but this was necessary in order to produce suitable test welds.

In the CTS test, all test welds were deposited at the energy input listed in Table 4. After the recommended 48 h holding time at the test temperature, they were sectioned and examined for cracking using the wet fluorescent magnetic particle process. The test severity was varied by lowering the test temperature. This was accomplished by conducting the tests in a cold room in which the temperature could be accurately controlled over a temperature range from room temperature down to −40 C (−40 F).

The implant testing procedures used were identical to those described by Sawhill.[2] In his technique the single-notched implant test bar, as proposed by Granjon, is replaced with a helical notch. This has the advantage that the notch will necessarily pass through the most crack-sensitive region of the HAZ. With Granjon's technique, careful attention must be given to ensure that the notch is accurately located in the critical region of the HAZ.

The implant test weld was deposited manually with Lincoln HYP.E7010 electrodes at a constant energy input. Two minutes after completion of the test weld, the surface of the assembly was inverted in water and cooled for 4 min. The test assembly was then removed from the water and loaded in a modified creep machine exactly 10 min after completion of the implant test weld. Approximately 12 specimens from each steel were tested in order to produce a stress vs. time-to-rupture curve from which could be determined the critical stress value for the steel.

HAZ Toughness Testing

HAZ toughness was assessed by means of the Charpy impact test, using ⅔-size specimens. Earlier work on submerged-arc seam welds showed that extreme variations were obtained in the Charpy impact values in the HAZ due to the difficulty of controlling the precise location of the notch in the HAZ. To overcome this problem the RPI Gleeble was used to simulate specific locations in the HAZ of real welds, and the Charpy impact values of these simulated specimens were determined.

The thermal cycles used in the HAZ simulations were based on actual measured thermal cycles from the root pass of a manual girth weld.[4] The thermal cycles experienced in the HAZ during the root pass of a manual girth weld are extremely severe in terms of the rapid heating and cooling rates. The actual measured thermal cycles are shown in Fig. 1.

The Gleeble equipment could not accurately reproduce these heating and cooling rates. Therefore, a decision was made to use the most rapid thermal cycles that the Gleeble could produce. These thermal cycles are shown in Fig. 2. The cooling rate was enhanced by helium gas cooling. The specimens used were ⅔-size Charpy bars 3 in. (76 mm) long and with no notch. After thermal cycling, the specimens were notched at the center where the control thermocouple was attached and then cut to proper length. All Charpy specimens were oriented such that the long direction was perpendicular to the original rolling direction of the plate.

Complete Charpy V-notch transition curves were obtained for each thermal cycle condition. Thermal cycles having peak temperatures of 700, 900, 1100 and 1300 C (1290, 1650, 2010 and 2370 F) were used. Three specimens were tested per temperature, and usually tests were made at seven different temperatures to establish the transition curves.

Experimental Results

Cold Cracking Tests

The results of the CTS tests on the different line-pipe steels are summarized in Table 5. Data were obtained for eight of the ten steels in the program. Plate thicknesses used were determined by the wall thickness of the pipe and ranged from 0.375 to 0.720 in. (9.5 to 18.3 mm) with the corresponding thermal severity number varying from 5 to 7. Also shown in Table 5 are the maximum HAZ hardness values recorded. Selected test welds from the CTS tests were sectioned and polished, and hardness traverses were made across the weld and HAZ using a Vickers hardness tester with a 10 kg (22 lb) load. In most cases, the maximum HAZ hardness corresponded to the region of the HAZ through which the crack traversed.

A typical cracked section through a CTS test weld is shown in Fig. 3. This particular example is the conventional line-pipe steel—steel F—which exhibited a large crack that ran entirely through the coarse-grained region of the HAZ at a test temperature of −16 C (+3 F).

In some specimens, the cracking followed a step-like path running between elongated inclusions; an example is shown in Fig. 4.

In general, the amount of cracking observed in all steels in the CTS tests

Table 2—Manufacturers' Analyses of Line-Pipe Steel Chemical Compositions, %

Line Pipe	C	Mn	P	S	Al	Nb	Si	Cu	Ni	Cr	Mo	Sn	V	N
A	0.04	0.44	0.010	0.015	—	0.04	0.27	1.22	0.91	0.67	0.20	—	—	—
F	0.16	1.25	0.008	0.009	—	0.047	0.35	—	—	—	—	—	—	—
J	0.11	1.28	0.014	0.008	0.033	0.04	0.25	—	—	—	—	—	0.06	0.005
M	0.06	1.65	0.012	0.014	—	0.054	0.04	0.19	0.19	0.09	0.24	0.014	—	—
N	0.07	1.38	0.010	0.009	0.050	0.140	—	—	—	—	—	—	—	—
P	0.10	1.47	0.017	0.004	0.039	0.04	0.27	0.28	0.28	—	—	—	0.09	—
R	0.10	1.48	0.014	0.002	0.033	0.05	0.25	0.31	0.32	—	—	—	0.10	—
S	0.10	1.30	0.008	0.003	0.034	0.03	0.30	—	0.45	—	0.08	—	—	—
T	0.05	0.67	0.015	0.003	0.035	0.04	—	0.93	0.96	0.29	0.30	—	—	—
Z	0.05	1.94	0.007	0.011	—	0.060	0.28	—	—	—	0.49	—	—	—

Table 3—Representative Mechanical Properties of Line-Pipe Steels[a]

Steel	Yield stress, ksi	Ultimate tensile stress, ksi	Charpy absorbed energy
A	84.0	87.0	58 ft-lb at −29 C(−20 F)—2/3 size
F	67.2	91.8	42 ft-lb, 97% shear at −4 C (+25 F)—2/3 size
J	66.7	84.8	38 ft-lb, 100% shear at −29 C (−20 F)—full size
M	72.3	91.3	59 ft-lb at −29 C (−20 F)—2/3 size
N	67.0	81.1	66 ft-lb, 88% shear at −4 C (+25 F)—2/3 size
P	77.2	95.3	40 ft-lb, 90% shear at −60 C (−76 F)—full size
R	74.7	88.5	127 ft-lb, 100% shear at −60 C (−76 F)—full size
S	72.1	94.0	217 ft-lb, 100% shear at −29 C (−20 F)—full size
T	75.1	90.8	187 ft-lb, at −29 C (−20 F)—full size
Z	73.0	91.8	76 ft-lb, 95% shear at −29 C (−20 F)—full size

[a] ksi × 6.89 = MPa; ft-lb × 1.356 = Joules.

Table 4—Welding Conditions Used in Cold Cracking Weldability Tests

Test	Electrode	Energy input
CTS	Lincoln HYP.E7010	780 J/mm (20,000 J/in.)
Implant	Lincoln HYP.E7010	1560 J/mm (40,000 J/in.)

was small and would indicate a good resistance to HAZ cracking. In many specimens examined, small cracks were observed in the weld metal rather than in the HAZ, indicating that in many of the steels hydrogen cracking is more likely to occur in the weld metal.

Implant tests were made on all ten steels included in the evaluation program. In the implant test a relationship of applied stress versus time-to-failure is established. From this relationship a stress level may be determined at which a specimen has a 50% probability of failing in the implant test. This stress is defined as the Critical Stress (CS) and can be used as a fundamental measure of the resistance of a steel to hydrogen-induced cold cracking. A high CS value indicates the steel will have a high resistance to HAZ cold cracking in welding; conversely, a low CS value indicates a low tolerance to hydrogen during welding.

In Fig. 5, typical stress/time curves obtained from the implant test are

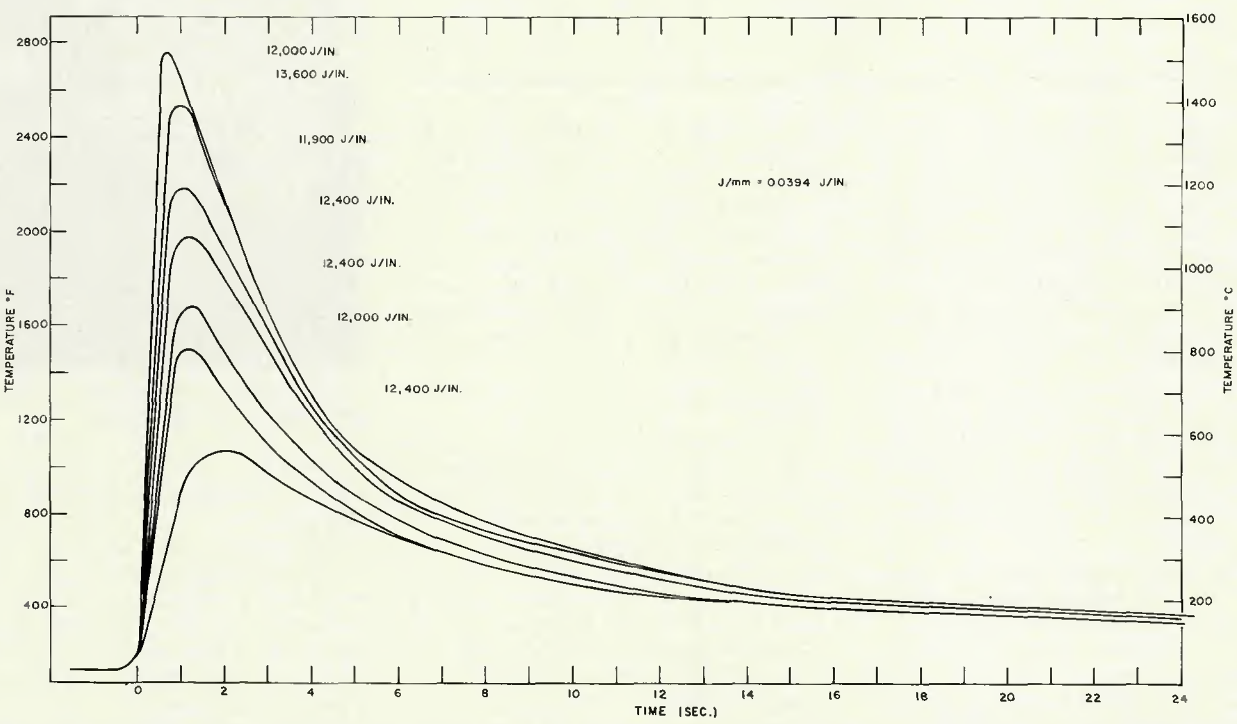

Fig. 1—The thermal cycles experienced in the root pass HAZ for the lowest practical heat input condition used in manual pipeline arc welding

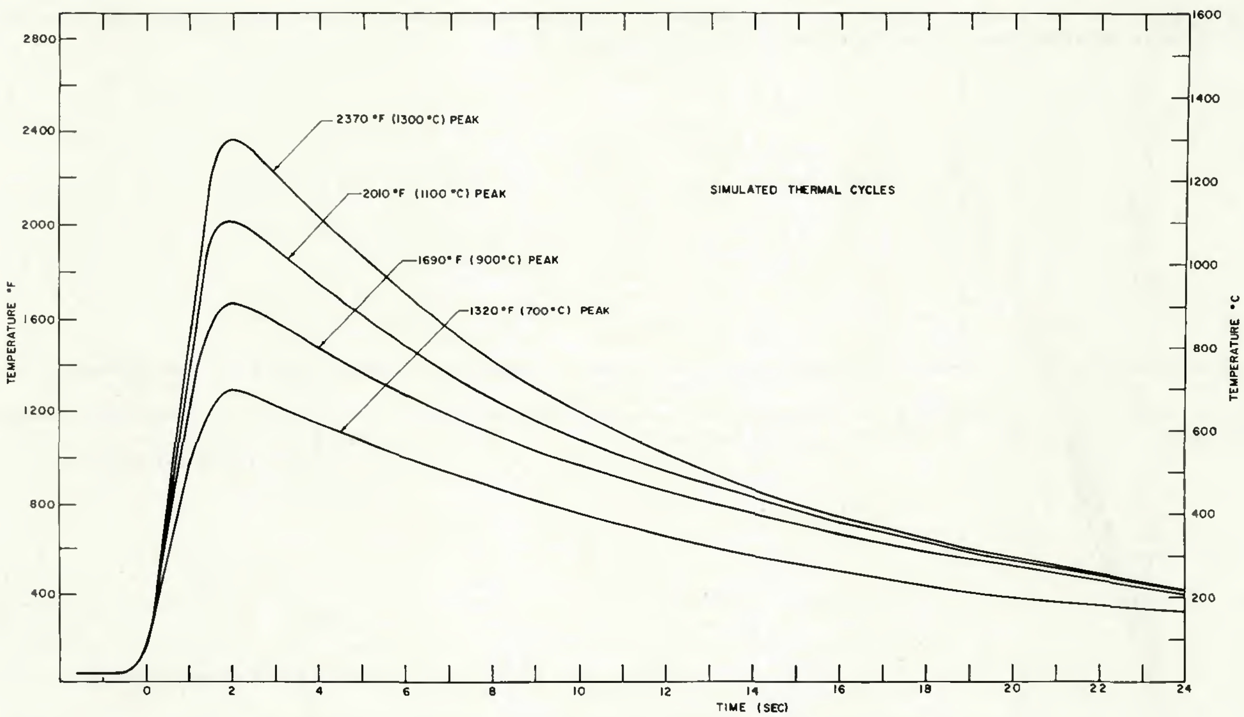

Fig. 2—The best simulation of the actual manual pipeline weld thermal cycles obtainable on the Gleeble

Table 5—Controlled Thermal Severity Test Results for Line-Pipe Steels

Steel	Thermal severity number	Max. HAZ Vickers hardness	Test temperature at which cracking first occurred	
A	5	318	<−36 C	(−32 F)
F	6	435	−16 C	(+3 F)
J	6	352	+21 C	(+70 F)
M	5	274	−27 C	(−16 F)
N	5	255	−36 C	(−32 F)
P	7	442	−17 C	(+2 F)
R	7	422	−17 C	(+2 F)
S	6	429	+21 C	(+70 F)
T	6	352	<0 C	(+32 F)
Z	6	355	<0 C	(+32 F)

shown for three selected steels in the evaluation program. The data in Fig. 5 show the two extremes of behavior. Steel R exhibits a CS value of 47.0 ksi (324 MPa), whereas steel M has a much higher value of 87 ksi (600 MPa). Also shown for comparison is the stress/time curve for the conventional C-Mn X 65 steel, i.e., steel F.

A summary of the entire implant results for all ten steels is given in Table 6. The results show a wide range in CS values for the ten different materials. All steels tested, except for steels P and R, showed CS values higher than that of the conventional line-pipe steel, steel F.

As in the CTS tests, sections were taken through selected fractured implant test specimens and examined metallographically. A typical example is shown in Fig. 6 and shows the fracture is almost entirely in the coarse-grained region of the HAZ. This was not always the case, and some steels (J, M and S) behaved in an anomalous manner and fractured outside of the HAZ. This occurred only at the higher applied stresses.

Hardness measurements using the Vickers Hardness Tester were made across selected implant test specimens, and the maximum values recorded in the HAZ are shown in Table 6.

HAZ Toughness

The variation in the toughness across the HAZ of the line-pipe steels was determined by testing simulated HAZ's produced in the Gleeble. Four locations within the HAZ were selected for simulation, and a sufficient number of specimens were simulated such that complete Charpy impact transition curves could be determined for these four locations. The four locations selected for simulating were those regions that heated to peak temperatures of 700, 900, 1100 and 1300 C (1290, 1650, 2010, 2370 F) during welding.

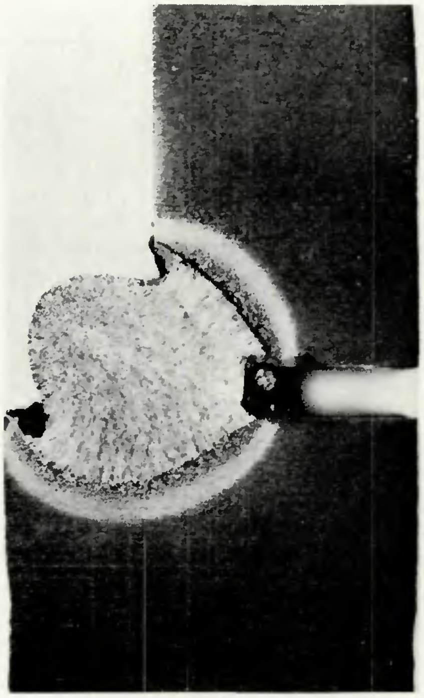

Fig. 3—Example of HAZ cracking in CTS test specimen of line-pipe F. ×6

Only six of the ten steels in the evaluation program were tested in this manner. The volume of data generated from this type of experimental work was large and is merely summarized here in an attempt to show general trends in behavior.

Figure 7 shows the Charpy impact transition curves for the simulated

HAZ of conventional steel F. A significant loss in Charpy impact toughness can be seen for the 1300, 900 and 700 C (2370, 1650 and 1290 F) peak temperature conditions. Only the 1100 C (2010 F) condition maintains impact toughness equivalent to the parent steel.

The microstructures that correspond to the four peak temperature conditions are shown in Fig. 8. No significant microstructure changes could be observed in the 700 C (1290 F) peak temperature condition that could explain the drastic loss in toughness shown in Fig. 7. Only the 1300 C (2370 F) condition exhibited a microstructure that correlated with the toughness levels obtained.

Steels P and M exhibited a similar type of behavior with severe losses in toughness properties in both the 700 C (1290 F) and 1300 C (2370 F) peak temperature conditions.

The pattern of behavior observed in the conventional steel was not common to all the steels tested. Steel A showed a different behavior as shown in Fig. 9. In this case only the 1300 C (2370 F) peak temperature condition caused degradation of impact toughness. The other peak temperature conditions showed impact toughness values equivalent to, or better than, the parent steel. The microstructures corresponding to the four peak temperature conditions for steel A are shown in Fig. 10. Steel N showed a similar type of behavior to that of steel A.

The final pattern of behavior was that observed for the quenched-and-tempered steel S. The results are shown in Fig. 11. In this case, the 700 C (1290 F) peak temperature condition has impact toughness levels close to the parent steel. The 900, 1100 and 1300 C (1650, 2010 and 2370 F) thermal cycles, however, produced a significant drop in toughness properties. The distinction between the behavior of this steel and the other steels is that, in relative terms, the loss in toughness is much greater. The microstructures corresponding to the four peak temperature conditions in steel S are shown in Fig. 12; they correlate well with the observed toughness properties.

An overall summary of the results for the six steels tested is given in Table 7. Here the data have been presented in terms of Charpy absorbed energy values at −29 C (−20 F) and −62 C (−80 F) for the four peak temperature conditions and also for the parent metal.

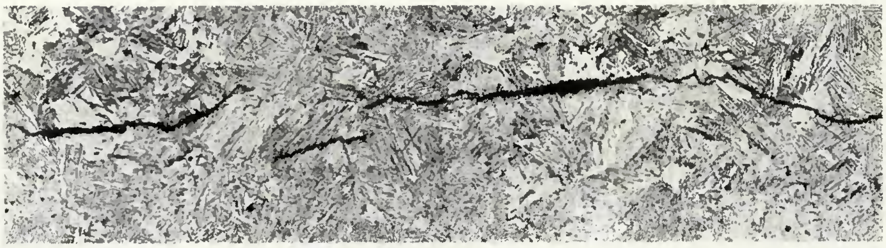

Fig. 4—Example of HAZ crack in line-pipe M in CTS test made at −37 C (−35 F) showing association of crack with inclusions. ×100

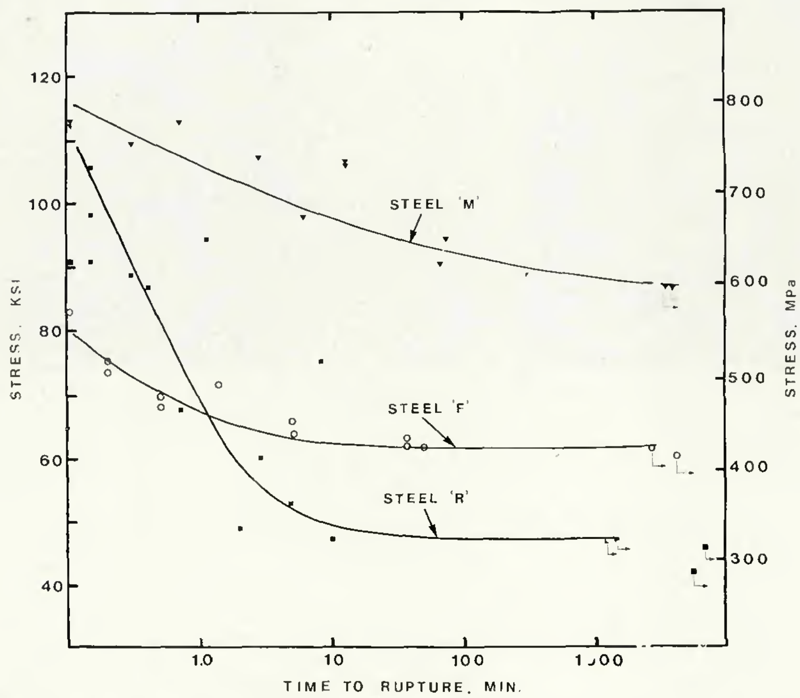

Fig. 5—Implant test results for steels M, F and R

Table 6—Summary of Implant Test Results on Line-Pipe Steels

Steel	Carbon Equivalent: Conventional	Carbon Equivalent: Ito-Bessyo	Lower critical stress, ksi	Maximum HAZ Vickers hardness
A	.43	.19	82.0 (565 MPa)	303
F	.37	.23	61.2 (421 MPa)	442
J	.34	.19	87.0 (600 MPa)	326
M	.43	.18	87.0 (600 MPa)	274
N	.30	.14	86.6 (595 MPa)	294
P	.40	.21	59.2 (407 MPa)	422
R	.41	.21	47.0 (324 MPa)	360
S	.36	.19	71.0 (490 MPa)	360
T	.41	.18	76.0 (524 MPa)	391
Z	.47	.18	65.0 (448 MPa)	336

Discussion

The weldability tests conducted on the nine Arctic-grade line-pipe steels to determine their susceptibilities to hydrogen-induced HAZ cracking confirmed that the majority had lower levels of susceptibility than the one conventional line-pipe steel that was also tested. Only steels P and R showed a greater susceptibility to cold cracking than the conventional steel.

Of the two test methods used to determine the susceptibility to cold cracking, the implant test was found to give a more sensitive measure of cracking susceptibility, and more reliance was placed upon these results than those from the CTS test. The CTS results shown in Table 5 indicate little real difference between the cracking susceptibilities of the ten steels tested. Only steels A, M and N showed a cracking tendency lower than the conventional line-pipe steel. They also showed much lower HAZ hardness levels. The remaining steels showed performances equal to or, in some cases, worse than the conventional steel.

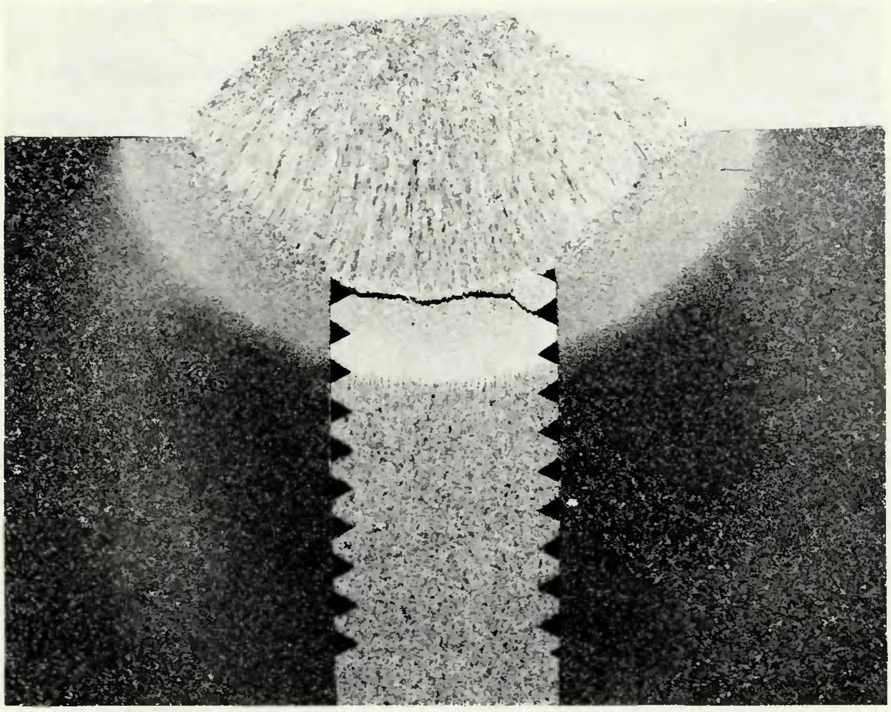

Fig. 6—Cross section through fractured implant specimen of line-pipe R. ×5 (reduced 14% on reproduction)

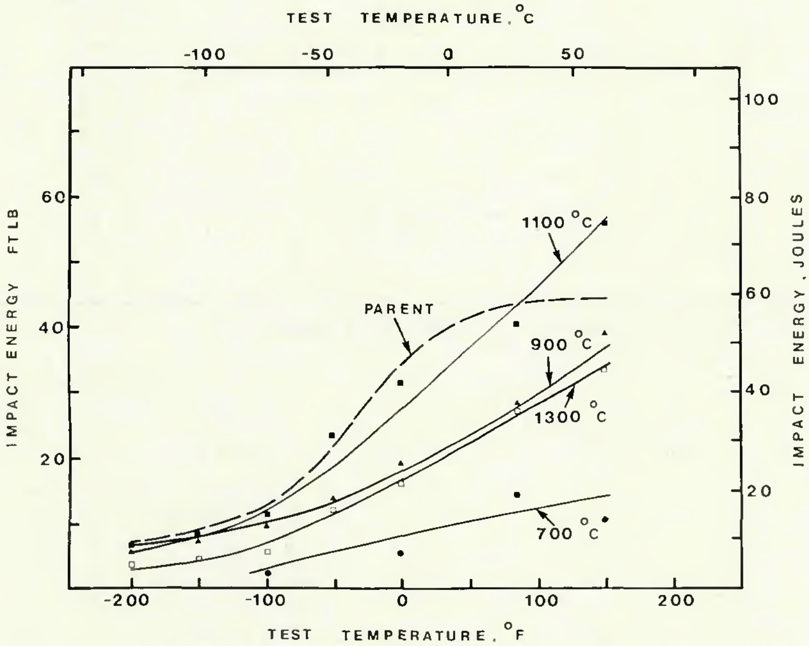

Fig. 7—Charpy V-notch impact properties of Gleeble-simulated heat-affected zones in line-pipe F (⅔-size specimens)

The CTS test is better suited to thicker-gauge materials where the test severity can be controlled by varying the specimen thickness. In this work, the test severity was increased by lowering the test temperature. This introduced experimental difficulties which reduced the accuracy of the test. In fact, considerable scatter in results was obtained.

On the positive side, the CTS does represent a real restrained weld condition similar to that produced in a circumferential girth weld. The joint configuration is, of course, completely different, being a fillet weld rather than the single-V butt joint in a circumferential pipe weld. However, attempts to use a weldability test with a butt configuration, such as the Tekken test and others, were not successful since cracking occurred almost exclusively in the weld metal rather than the HAZ.

It was concluded that these tests were evaluating weld metal cracking susceptibility rather than HAZ cracking; they were, therefore, discontinued. In the CTS test, cracking could be more readily produced in the HAZ. In most steels tested, however, cracking also occurred in the weld metal, and it must be concluded that in many of these materials hydrogen-induced cracking of the weld metal is likely to be a more serious problem than cracking in the HAZ.

The implant test gave a much better differentiation between the cracking susceptibilities of the different line-pipe steels. The results in Table 6 show a wide range in the CS values for the different steels. This, in turn, will mean a wide range in the relative cold-cracking susceptibilities for the different steels. Steels A, M and N showed high resistance to cold cracking as evidenced by their high CS values. This is in good agreement with the CTS results obtained on these three steels. Steel J also showed a high CS value but did not perform as well in the CTS tests. Of the remaining steels (except for P and R), all showed higher CS values than the conventional line-pipe steel F. Steels P and R had CS values below that of the conventional steel.

Also shown in Table 6 are the carbon equivalents (C.E.) for the ten different steels. The carbon equivalent is commonly used as an indicator of the tendency of a steel to hardening in the HAZ and susceptibility to hydrogen-induced cracking. Most pipeline specifications impose upper limits upon the C.E. values of the line-pipe steels. The carbon equivalent formula used in most pipeline specifications is as follows:

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

Although this formula works very well for the conventional C/Mn line-pipe steels, it has been recognized by many workers that it is not as appropriate for the newer micro-alloy steels. The results in Table 6 confirm this very clearly. For example, steels A, M and Z have high C.E. values as determined by the above formula and yet show quite low HAZ hardnesses and high resistances to HAZ cracking. Conversely, conventional line-pipe steel F has one of the lowest C.E. values and yet has high HAZ hardness and low tolerance to HAZ cold cracking.

Attempts have been made to modify the C.E. formula. One relationship that has been reported to work much better for the micro-alloy steels is the formula proposed by Ito and Bessyo.[5] Their formula is as follows:

$$C.E. = C + \frac{Si}{30} + \frac{Mn}{20} + \frac{Cu}{20} + \frac{Ni}{60} + \frac{Cr}{20} + \frac{Mo}{15} + \frac{V}{10} + 5B$$

The C.E. values obtained using this relationship are also listed in Table 6. In this case the correlation of C.E. value to HAZ hardness and cracking susceptibility is much better although still not perfect. It would appear that the Ito-Bessyo formula is, however, a better measure of the C.E. of the micro-alloy steels and is a more reasonable guide to the weldability of the steel.

The results obtained for steels P and R indicate danger in placing too much reliance upon the C.E. values. Both steels have reasonably low C.E. values and yet show a low resistance to hydrogen cracking. No clear explanation for this was determined. However, both steels have heavily banded microstructures with long, heavily rolled out inclusions. Both factors have an important bearing upon the susceptibility to hydrogen cracking. Unfortunately, no C.E. formula can take into account these factors.

The determination of HAZ toughness of the line-pipe steels is a difficult problem due to the narrow width of the HAZ and the wide variation in microstructure across its width. Most Arctic pipeline specifications still call for minimum toughness values in the HAZ and usually specify the test procedure to be the Charpy impact test. The Charpy impact values obtained when testing the HAZ of the weld will be a function of the location of the notch within the HAZ and, unfortunately, this is often not specified.

One solution to obtaining more meaningful Charpy impact values in the HAZ is the approach used in this work—namely to test Gleeble-simulated HAZ specimens. This also has its limitations. It does not represent the true condition of the HAZ, since at least part of the HAZ is reheated by subsequent weld passes. Also, the Gleeble could not simulate exactly the most severe thermal cycles experienced in the HAZ of the root pass of the girth weld. This is not as important as it would appear since the remaining weld passes are deposited using higher energy inputs that would produce slower HAZ cooling rates which would approximate those produced in the Gleeble. The thermal cycles used were, therefore, quite representative of the overall thermal cycle conditions in the HAZ of the girth weld.

The results obtained from the Gleeble-simulated HAZ specimens have confirmed that there is a wide variation in the impact properties across

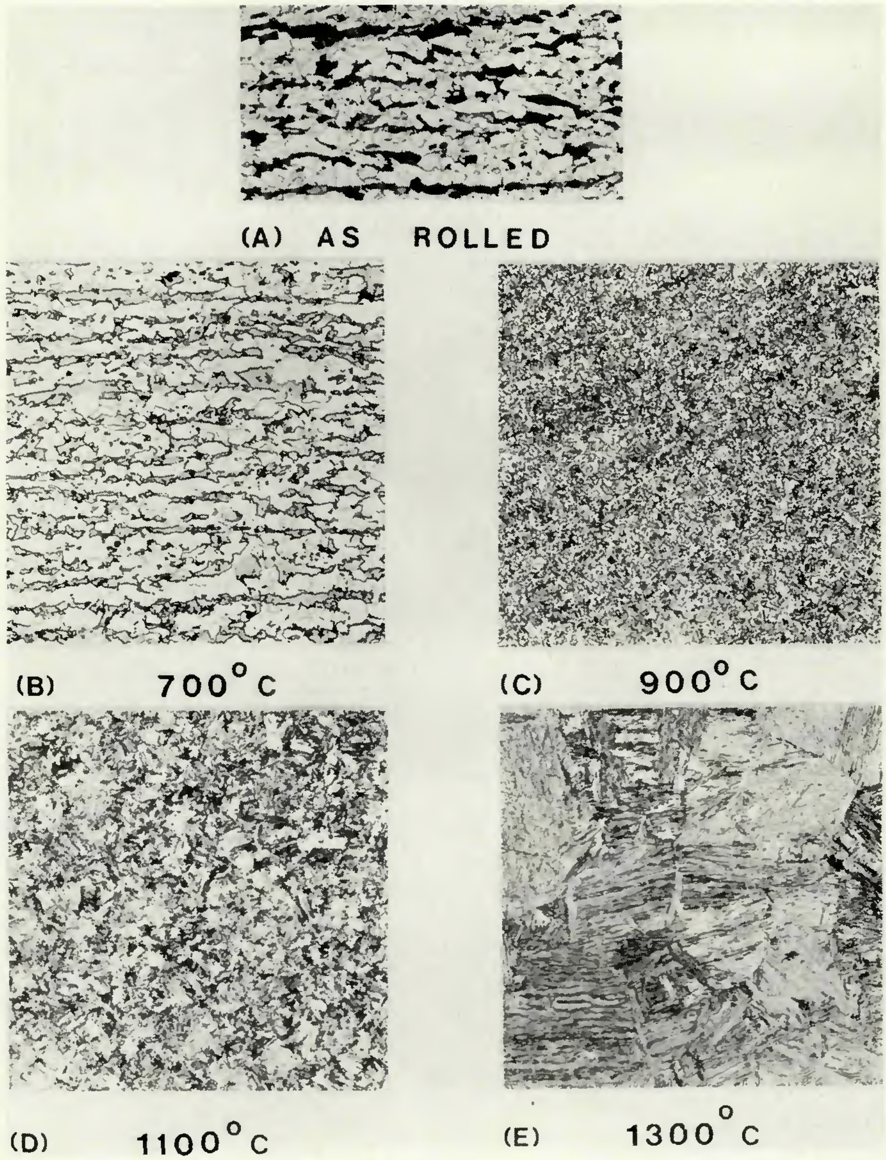

Fig. 8—Simulated HAZ microstructures in line-pipe F. ×250 (reduced 7% on reproduction)

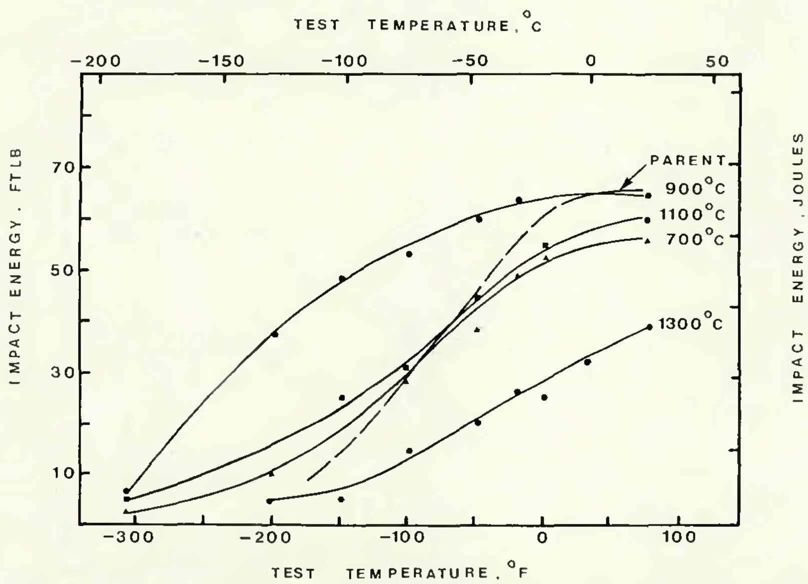

Fig. 9—Charpy V-notch impact properties of Gleeble-simulated heat-affected zones in line-pipe A (⅔-size specimens)

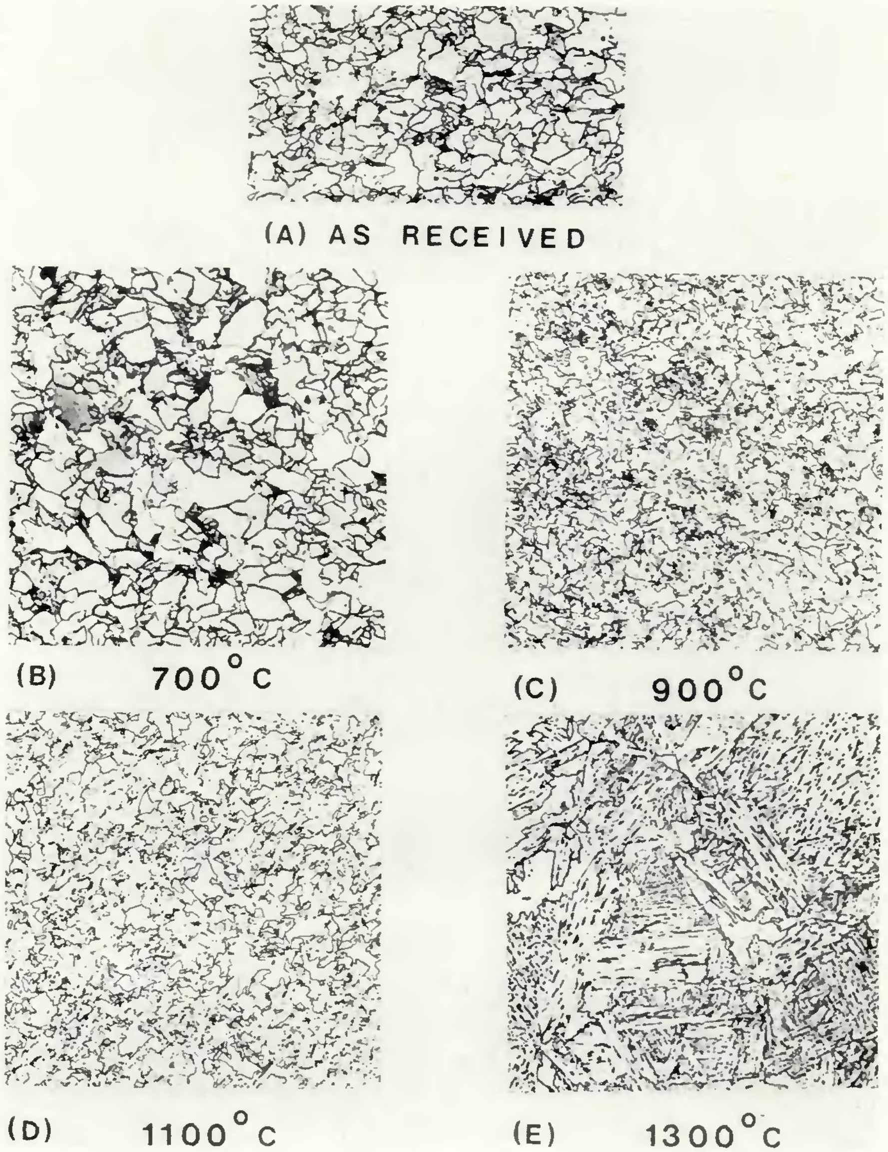

Fig. 10—Simulated HAZ microstructures of line-pipe A. ×400 (reduced 7% on reproduction)

the HAZ in all steels tested. Furthermore, the behavior of each steel is different. In general the steels can be separated into three groups of behavior. First are those such as steels F, P and M which showed large losses in toughness in the 1300 C (2370 F) and 700 C (1290 F) peak temperature thermal cycle treatments. The second group (steels A and N) only showed a toughness loss in the 1300 C (2370 F) peak temperature condition with the remaining thermal cycles having little effect or actually improving toughness. The third type of behavior was that depicted by steel S; this showed a large loss in toughness for all thermal cycles except the 700 C (1290 F) peak temperature thermal cycle. Of the three types of behavior, the second type shown by steels A and N is most desirable.

The exact significance of having bands of low toughness material in the HAZ is not fully understood. It can be assumed that it will cause an overall lowering of the ductile fracture resistance of the HAZ and, therefore, should be minimized. All steels tested showed very low toughness corresponding to the 1300 C (2370 F) region, and it is unlikely that much can be done to prevent this. It is a very narrow band within the HAZ and, provided it is surrounded by material of good toughness, is probably not very harmful.

For steels that develop poor toughness in other regions in the HAZ, however, a more careful appraisal of the overall properties of the HAZ should be made. Examination of the results in Table 6 shows that several of the steels tested have low Charpy impact values across most of the HAZ. It is unlikely that they could meet the minimum low-temperature Charpy impact requirements for the HAZ that will be demanded in the Arctic pipeline specifications.

Conclusion

The purpose of this work was to compare the welding characteristics of a number of so-called Arctic-grade line-pipe steels. Two aspects of weldability were considered. They were the susceptibility to HAZ hydrogen crack-

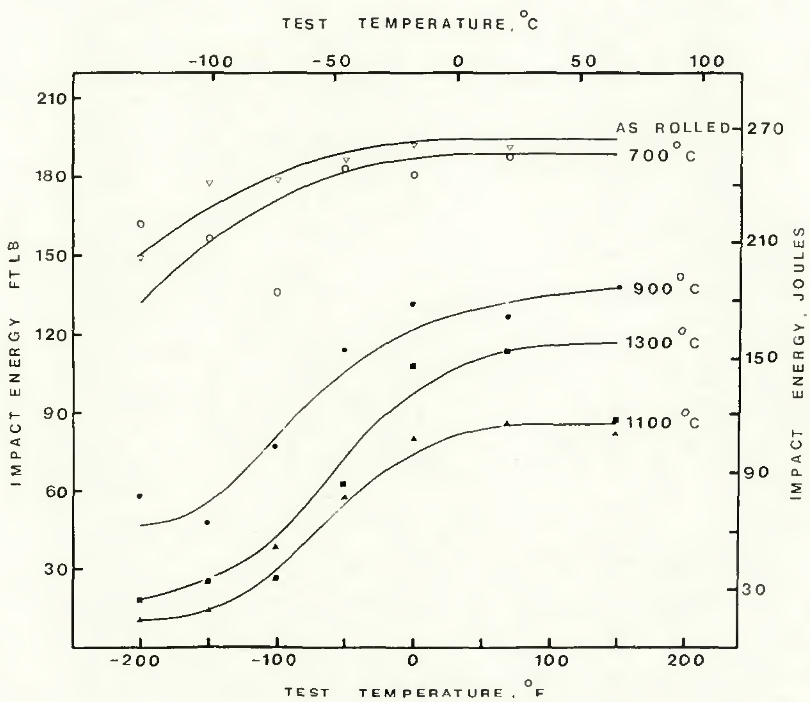

Fig. 11—Charpy V-notch impact properties of Gleeble-simulated heat-affected zones in line-pipe S (⅔-size specimens)

ing and the variation in Charpy toughness in the HAZ. Both are critical factors that will qualify the selection of pipeline steels to be used in the Arctic. This does not mean that these two factors are the only welding considerations to be made. Of equal importance will be the development of the required properties in the weld metal. This is, however, more a function of the welding process and the welding consumables used than of the steel being welded, and was therefore not considered in this work.

The results have shown that there is a wide variation in behavior for the different steels tested despite their similar compositions and microstructures. In general, it was concluded that the weldability characteristics of the Arctic-grade steels were better than those of the conventional steel.

In the HAZ cracking susceptibility tests it was shown that all but two steels tested—steels P and R—had a better resistance to HAZ hydrogen cracking than the conventional steel. The poor cracking resistance of steels P and R was thought to be related to their heavily banded microstructure.

The modified implant test used in this work was found to be a sensitive test for determining susceptibility to hydrogen-induced HAZ cracking. The CTS test gave much less reproducible results and was not sensitive enough to differentiate between the steels tested.

Examination of the ten steels on the basis of their C.E. values confirmed that the conventional C.E. quoted in most specifications is not valid for these new high-strength low-alloy steels. The Ito-Bessyo relationship was found to correlate quite well with the implant test results and is a more appropriate formula to use for the micro-alloy steels.

In the study of the HAZ toughness using Gleeble-simulated specimens, it was shown that a wide variation in toughness will exist across the HAZ and that the extent of this variation is different in each steel. The only common factor to all steels was the severe loss of toughness that is produced in those regions of the HAZ that are heated through a weld thermal cycle having a peak temperature of 1300 C (2370 F).

In some steels (F, M, and P) a similar loss in toughness occurs in the 700 C (1290 F) peak temperature regions of the HAZ, but with no observable change in microstructure. In the one quenched-and-tempered steel tested, the Charpy toughness was reduced across the entire HAZ except for the 700 C (1290 F) region.

It must be concluded that a careful appraisal of the low-temperature HAZ toughness should be made before selecting a line-pipe steel for application in an Arctic environment. The results of this work show that the wide variation in toughness properties that can occur across the HAZ are not predictable on the basis of composition or microstructure. The Gleeble approach used in this work is a useful

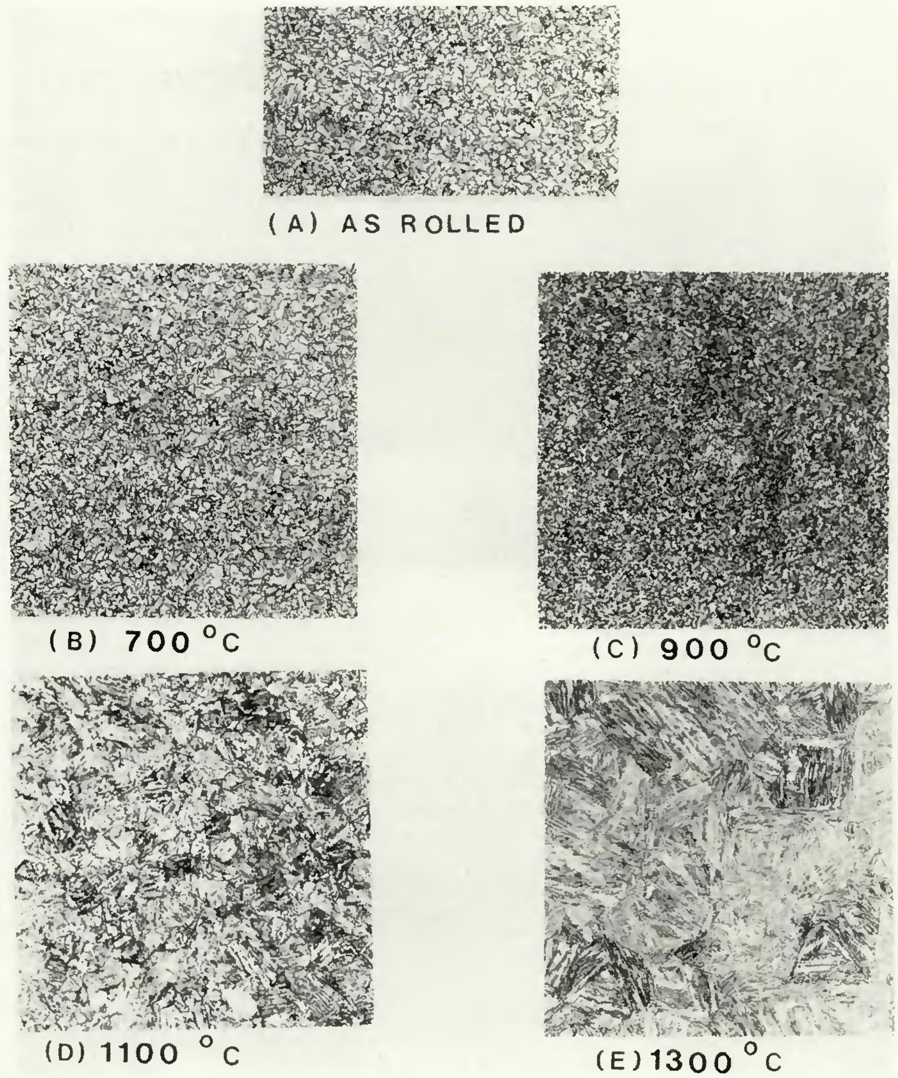

Fig. 12—Simulated HAZ microstructures of line-pipe S. ×250 (reduced 15% on reproduction)

Table 7—Charpy Absorbed Energy Values for Gleeble-Simulated Heat-Affected Zones of Line-Pipe Steels—2/3-Size Specimens Tested at −29 and −62 C (−20 and −80 F)

	Peak temperature condition				
Steel	As-rolled	700 C (1290 F)	900 C (1650 F)	1100 C (2010 F)	1300 C (2370 F)
	—Absorbed energy at −29 C (−20 F)—				
A	57	48	64	50	26
F	22	8	14	27	11
M	60	10	95	72	6
N	50	44	78	35	1
P	35	12	23	14	8
S	192	186	120	68	50
	—Absorbed energy at −62 C (−80 F)—				
A	35	34	58	36	16
F	—	4	8	20	11
M	22	4	65	15	6
N	—	14	63	18	1
P	25	7	13	7	8
S	185	176	100	42	50

technique to obtaining a fuller understanding of the behavior of the HAZ on welding these materials.

Acknowledgments

Thanks are expressed to Dr. R. J. Cooke for his significant contribution to this work. Appreciation is also expressed to M. Letts for his assistance in the experimental work. Finally, thanks are expressed to Dr. K. Winterton for interest and input to the work.

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