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Properties of High-Strength Steels for High-Pressure Containments

edited by
E. G. NISBETT

Properties of High-Strength Steels for High-Pressure Containments

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FOREWORD

Vessels for high pressure containment applications have been in use for varied purposes and industries for many years. For most of these vessels the Rules of the ASME Boiler and Pressure Vessel Code could not be applied because of material and design limitations and commonly special arrangements had to be made with the cognizant jurisdiction of the location where it was to be used. Methods of manufacture broadly included welded layered vessels particularly for the lower end of the high pressure range, wire or strip wound vessels for the higher pressures and monoblock forged vessels, or forged multiwall vessels again for high pressures.

For the forged vessels, some were made from surplus naval battleship guns designed to fire 12 to 16 inch shells. These barrels were often multiwalled, and with the breech closure at one end could be cut to the required void length, equipped with a closure for the opposite end and thus make a pressure vessel. Some vessels have even been made from the 16 inch shells themselves. The large naval guns (12 to 16") were commonly made from nickel steels (2300 series). Field artillery and anti-aircraft barrels utilized Ni-Cr-Mo-V steels and these were commonly treated to yield strengths in the range from about 1000 MPa (145 ksi) to 1240 MPa (180 ksi). These materials, generally containing higher alloy contents than the well known AISI 4300 series steels, were widely known as gun steels, and came to be used for forged high pressure vessel bodies and closures. Similar alloys were also in use for turbine and generator rotor forgings via such specifications as ASTM A470 and A469, but none were available in the form of an ASTM or ASME specification suitable for pressure vessels.

The need to make high pressure vessels with an ASME Code stamp spurred the writing of a pressure vessel material specification to cover the gun steels and in 1975 ASTM published A723 This specification was unusual in that a choice of three broad specification grades could be taken with one of five strength classes to enable the alloy to be matched with the forging size and configuration for a given strength class. This enabled code case number 1873 to be published by ASME for the manufacture of high pressure vessels and this has since been incorporated into Section VIII, Division 2 of the Code.

The rules of Section VIII Division 2 place some appreciable restraint on the manufacture of many high pressure vessel designs, and this, together with the restriction on the use of material to SA-723 to the Classes 1 and 2 strength levels has spurred an on going effort within Section VIII to write new rules specifically for high pressure vessels.

These rules are being drafted by the Special Working Group on High Pressure Vessels, of Section VIII, and a need has been expressed for data on fracture toughness, fatigue crack growth rates and fatigue strength for the materials being considered. The materials of SA-723 are prime candidates for the assessment of this data, and this symposium is intended to present and collect the available information prior to beginning a test program under the auspices of the Materials Properties Council to fill the gaps and enable the design rules for the new Section VIII division to be completed.

This activity will be extended to other materials suitable for use in high pressure vessels and the current program also includes additional information on the well established chromium molybdenum pressure vessel steels. This symposium really marks the beginning of the efforts to gather the fatigue and fracture toughness data for the new high pressure vessel code section. All interested parties are urged to participate both by discussion at this symposium and by contributing material data.

E. G. Nisbett

CONTENTS

REDUCING THE SUSCEPTIBILITY TO TEMPER EMBRITTLEMENT IN 2¼ Cr-1Mo STEELS BY LANTHANIDE ADDITIONS

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ABSTRACT

The temper embrittlement behavior of a series of 2¼%Cr-1%Mo steels with and without lanthanide (rare earth) metal (LnM) additions has been studied. The results of embrittling treatments (conventional step-cooling treatments versus long isothermal hold treatments) have been assessed in terms of their effect on the impact transition temperature and the fracture mode of the steels. Factors such as the extent of carbide precipitation, distribution and types of carbides have been investigated in terms of their role in the resulting temper embrittlement of the steels.

Introduction

It is well recognized that temper embrittlement results when many alloy steels are held within, or cooled slowly through a temperature range of about 371 to 593°C (700 to 1100°F). Temper embrittlement is manifested by a shift in the ductile to brittle transition temperature of a steel to higher temperatures. The major factors that influence the relatonship between microstructure and temper embrittlement are fairly well understood. For example, the combined effects upon temper embrittlement of austenitic grain size and impurities (such as P, Sb, As and Sn) segregating to prior austenite grain boundaries was indicated by Woodfine(1) in 1953.

In recent years considerable interest in the role of impurities on the temper embrittlement of steels has been triggered by the continuing increase of the electric furnace steelmaking process. This process consumes large quantities of scrap, which is believed to be a major source of impurities(2). The presence of impurity and tramp elements tend to promote temper embrittlement in low alloy steels. The mechanism involved in this process is the segregation of impurities to prior austenite grain boundaries(1, 3-6). It has also been shown that there is a concomitant co-segregation of impurities with major alloying elements such as Cr and Mo. This co-segregation is believed to be a major driving force in controlling the rate of segregation of impurities at grain boundaries and at carbide-matrix interfaces.

Thermodynamic (7,8) and kinetic (9,10) models have been formulated to account for the influence of alloy-impurity interactions on the degree of segregation and embrittlement. Although these models (7-10) provide a strong basis for the fundamental understanding of temper embrittlement, they are not readily applicable to explain temper embrittlement in more complex multicomponent systems such as commercial steels. That is, during temper embrittlement treatments (either conventional step- cooling or long isothermal holding), several microstructural changes can occur, thus inducing a continuous alteration of the thermodynamic equilibrium state conditions. Recently, Garcia et al (11) have shown that the degree of embrittlement, and the accompanying fracture mode in a 2¼%Cr-1%Mo steel, are strongly associated with microstructural changes such as the precipitation and transformation of carbides and impurity segregation phenomena. These observations are in good agreement with other studies.(6,12,13)

Several efforts to reduce the effect of impurities on temper embrittlement have been reported in the literature. Some studies are based on the use of scavenger elements that combine with the impurities, thus preventing their segregation to grain boundaries. Of particular interest is the work by Seah et al(14), Morrison(15) and others (11,16-18), using lanthanide (rare earth) metal additions to reduce the susceptibility to temper embrittlement in 2¼%Cr-1%Mo steels. In these studies, the amount of lanthanide metal (LnM) additions was based on calculations to satisfy stoichiometric conditions.(14) Although LnM additions calculated from theoretical considerations reduced the susceptibility to temper embrittlement in low-alloy steels, the shift in the impact transition temperature (TT) to lower temperatures was attended by a decrease in the upper-shelf energy of the steel. Recent work (11) has shown that additions of LnM levels well below those dictated from theoretical calculations were sufficient to trap the impurities. Thus, improvements in the resistance to temper embrittlement without a significant loss in both hot ductility and the upper-shelf energy of the steels can be obtained.

The present study was undertaken to assess the influence of tempering (embrittling) treatments (step cooling versus long isothermal) on the temper embrittlement behavior of 2¼%Cr-1%Mo steel with LnM additions.

Experimental Procedure

Four 227 kg (500 lb.) experimental steel heats were prepared by air-induction melting with an argon-protecting atmosphere. The chemical composition (in wt.%) of the laboratory-melted 2¼%Cr-1%Mo heats are given in Table I. Steel 1 was a vacuum-melted base heat, whereas steels 2 through 4 were induction-melted heats. Two ingots 108 mm x 108mm x 266 mm (4.25 inch x 4.25 inch x 10.5 inch) were cast for each heat. One as-cast ingot from each heat of steel was sectioned at top, center, and bottom locations for macrostructural and segregation studies. The second ingot from each heat of steel was heated at 1200^{o}C (2192^{o}F) for three hours and hot rolled to 13 mm (0.5 in.) plates and then air cooled to room temperature. The as-hot rolled plates were then cut into 150 mm x 120 mm x 13 mm (6 inch x 5 inch x 0.5 inch) blanks and austenitized at 950^{o}C (1750^{o}F) for 1 hr and water quenched. These quenched steel blanks were then tempered to obtain a tensile strength of approximately 690 N/mm^2(100 ksi). At this point, the steel blanks were divided in two groups. One group represented the quenched and tempered (QT) condition. The other two groups were subjected to an embrittling treatment (TE) that consisted of the following usual step-cooling treatment: 595^{o}C (1103^{o}F) 1 hr + 540^{o}C (1000^{o}F) 15 hr + 525^{o}C (975^{o}F) 24 hr + 495^{o}C (952^{o}F) 48 hr + 470^{o}C (875^{o}F) 72 hr, furnace cooled to 315^{o}C (600^{o}F) 1 hr and then air cooled to room temperature. These samples represented the temper embrittled (TE - SC) condition. In addition to this step-cooling embrittling treatment, additional blanks in the QT condition were subjected to a relatively long isothermal tempering treatment. These samples were held 2000 hr at 520^{o}C (968^{o}F) and then water quenched (TE-LIT).

Table I. Composition of Steels Evaluated,wt%*

Steel No.	Alloy Addition and/or Variations	P	Sn	As	Sb	Total LnM	La
1	B + P	.015	—	—	—	—	—
2	High P + T	.024	.015	.015	.005	—	—
3	High P + T +MM	.020	.016	.023	.007	.06	—
4	P + T + La	.016	.016	.019	.006	—	.03

*In adlition, all steels contained C-0.0/0.12, M-0.50/0.55, S-0.004/0.008, Si-0.20/0.26, Al-0.001/0.022, Cr-2.19/2.25, Ni-0.10/0.12, Mo-0.94/0.99 and N$_2$-0.005/0.008.

NOTE: B denotes base; T denotes tramp elements (Sn, As, Sb); LnM denotes lanthanide (rare earth) metal; MM denotes mischmetal (50% Ce, 30% La, 15% Pr and the balance other lanthanides).

The assessment of the microstructures, non-metallic inclusions and precipitates present in the steels was conducted using optical and electron microscopy techniques.

Thin-foil and carbon extraction replica specimens were prepared from the heat-treated 2.25 Cr-1 Mo steels. These samples were examined in a JEOL 200CX Temscan at 200 kV. Alloy carbides were identified by electron diffraction and STEM qualitative energy dispersive x-ray analysis.

Needle-shaped specimens for atom probe field-ion microscopy were electropolished using a 2% $HC10_4$ - butoxyethanol solution at 18V. Microchemical analyses were conducted at 50 K in 2 X 10^{-7} torr Neon.

Standard Charpy V-notch specimens, transverse to the rolling direction, were used to measure the impact energies on the steels from the QT and TE conditions. The Charpy V-notch impact tests were conducted over the temperature (5)range -200^{o} to + 100^{o}C (-328 to 212^{o}F). The fracture surfaces were examined by scanning electron microscopy (SEM) to determine the fracture modes.

Results

The results of the Charpy V-notch impact energy testing obtained for steels 1, 2, 3 and 4 in the QT and from both TE (step cooled and long isothermal) conditions are shown in Figures 1 and 2. The results illustrated in Figure 1, showed that when steel 1(base + 0.015%P) was temper embrittled either by conventional step cooling (TE - SC) or long isothermal tempering (TE-LIT), a reduction in the upper-shelf energy (USE) and a shift in the impact transition temprature (TT) to higher temperatures was observed when compared to the same steel in the QT condition for a given impact energy value. The reduction in the USE and shift in the TT was more severe in the TE-LIT than in the TE-SC condition, Figure 1.
These results on 227 kg (500 lb) heats are in good agreement with results from a previous investigation made on 23 kg (50 lb) heats of steel with similar compositions.(11)

A comparison of the embrittlement behavior of steels 2 (0.024% P + tramps), 3 (0.020% P + tramps + .06%LnM) and 4 (0.016% P + tramps + 0.03% La) from this investigation, after step-cooling and long time isothermal hold treatments, are shown in Figure 2. As expected, the degree of embrittlement was less in the steels with LnM additions than in the steel without it (steel 2). In addition, pure La additions (steel 4) were the most beneficial in reducing the susceptibility to temper embrittlement.

The adverse effects of impurity and tramp elements on the promotion of temper embrittlement in low alloy steels after a step-cooling treatment from the previous study,(11) is shown in Figure 3. Note the embrittlement behavior of steels with increased P (steels A and B) and the presence of tramp elements with P (steels C and D).

The results in Table II(11) clearly indicate the beneficial effects of LnM (6)additions in reducing the shift in TT to higher temperatures between the as-tempered (AT) and as-embrittled (TE) condition.

Table II. 54J (40 ft-lb) Transition Temperature (T_T) of Steels in Tempered (QT) and Embrittled (TE) Conditions (11)

Steel	Alloy Addition and/or Variation	T_T C QT	T_T C TE	T_T,abs
A	B + P	-122	-105	+17
B	High P	-110	-85	+22
C	T	-107	-55	+52
D	High P + T	-60	+140	+200
E	High P + T + 0.06 LnM	-97	-75	+22
H	P + T + 0.03 La	-112	-93	+20
L	P + T + 0.03 Nd	-105	-90	+15

Note: B denotes base; T denotes tramp elements (Sn,As,Sb); LnM denotes lanthanide (rare earth) metal

The beneficial effect of LnM (rare earths), added as mischmetal or added as pure individual lanthanide elements (La, Ce, Nd), is illustrated in Figure 4. It was found(11) that LnM combined with oxygen, sulfur and the impurities present. First, lanthanide (rare earth) oxysulfide $(LnM)_2O_2S$ inclusions were formed, Figure 5. Second, these inclusions were associated with impurities and/or tramp elements such as P, As, Sn and Sb. The inclusions shown in Figure 6 exhibit a "layer" type of configuration that became increasingly prevalent at relatively high lanthanide (rare earth) contents. This layer type of morphology suggests that $(LnM)_2O_2S$ inclusions form first and then the impurities precipitate and coat the oxysulfides. Similar results were found in all the LnM-containing steels, added as mischmetal or as individual elements (La, Ce or Nd).(11) A possible explanation for this behavior is that once the $LnM)_2O_2S$ is formed, this compound acts as a nucleation site (host structure) for precipitation of the impurities (probably in the form of Fe-P, Fe-Sn, Fe-As and Fe-Sb) on th surface of the oxysulfide. Thus, a complex compound is formed with the layer type morphology.

The resulting fracture mode of the steels embrittled under conventional step cooling conditions (TE-SC) was predominantly cleavage, and no significant volume percent of intergranular fracture was observed. Larger amounts of intergranular (7)fracture were detected in the steels embrittled during the relatively long isothermal treatments (TE-LIT). The volume percent of both cleavage and intergranular fracture resulting from the embrittling treatments is illustrated in Table III. The results from Table III and Figures 1 and 2 indicate that more severe temper embrittlement results from long isothermal tempering than from a conventional step cooling treatment.

Table III. Summary of Fracture Mode Observed After Testing At - 100°C

STEEL		AREAL PERCENT(%)-FRACTURE* STEP COOLED C	STEP COOLED I	LONG ISOTHERMAL C	LONG ISOTHERMAL I
1	0.015%P	99	1	93	7
2	0.024% + Tramps	95	5	65	35
3	0.20%P + Tramps + 0.06LnM	97	3	87	13
4	0.016%P + Tramps + 0.03La	97	3	90	10

*EXPERIMENTAL ERROR ± 3%
C = Cleavage fracture
I = Intergranular fracture

Transmission electron microscopy (TEM) examination revealed extensive carbide precipitation on both lath and prior austenite grain boundaries for the steels subjected to the step-cooling treatment (TE-SC). Intralath precipitation was also observed. Three types of carbides were identified: M_3C, M_7C_3, and M_2C. With the exception of relatively large complex inclusions (La or Ce-rich) which were observed in the LnM treated steels, the microstructures of all four steels were similar.

The microstructure of the isothermally aged steels (TE-LIT) was characterized by the presence of Fe-rich M_3C and M_7C_3 carbides. With the isothermal temper embrittlement treatment, an increase in the relative amount of Mo-rich M_2C carbide was observed.

In both temper-embrittled conditions (TE-SC and TE-LIT), EDS qualitative microanalysis of the Mo-rich carbides revealed a slight broadening of the Mo_L peak, which may signify the presence of P at these precipitates.

A summary of the electron microscopy analyses for steel 2 (0.024%P + Tramps)is shown in Table IV. The various alloy carbides were identified by electron diffraction and STEM qualitative energy dispersive X-ray analysis. Based upon the analysis of 150 idividual precipitates, the relative proportions of the carbides were found to vary with embrittling heat treatment.

Table IV. Summary of AEM Microstructural Study for Steel 2 (0.024%P + Tramps)

TYPE CARBIDE	COMPOSITION	RELATIVE PROPORTION STEP COOLED	RELATIVE PROPORTION ISOTHERMAL
M_3C	Fe-Cr-(Mo)	35%	10%
M_7C_3	Cr-Fe-(Mo)	50%	55%
M_2C	Mo-Cr	15%	30%
M_6C	Mo-Fe-Cr	---	---
$M_{23}C_6$	Fe-Cr-Mo	---	5%

Experimental Error ±10

Discussion

It is well known that temper embrittlement behavior, and the accompanying fracture mode in low alloy steels are strongly associated with microstructural changes such as precipitation and transformation of carbides and impurity segregation phenomena (3-6,12,13).

The extent of precipitation, distribution and type of carbides and segregation of tramp elements observed was not as severe in the microstructures resulting from conventional step-cooling treatments as compared with that observed in the microstructures after long isothermal treatments. Therefore, the degree of temper embrittlement would be expected to be less in the step-cooled steels. It is believed that the difference in microstructural features of the two temper embrittling treatments may be responsible for the difference in embrittlement behavior and fracture mechanisms observed,

Table III. It appears that the variation in extent of M_3C, M_7C_3, M_2C, and $M_{23}C_6$ carbides was responsible for the increase in inter-granular fracture in the isothermally treated steels, thus the increase in temper embrittlement susceptibility.

The results of this study suggest a strong relationship between the degree of embrittlement and the nature of the carbide present in the steel and the segregation of impurities to the carbide and/or to the prior austenite grain boundaries. For example. in the steels with LnM additions (3 and 4), the impurities (P, Sb, Sn, and As) are combined with lanthanide-oxysulfides forming stable compounds, thus preventing their segregation to carbide/matrix interface and grain boundaries. Therefore, the embrittlement observed in these steels was most likely due to the nature of the carbide formed during the step-cooled and long isothermal treatments, Tables II and III. In the case of the steels without LnM additions (1 and 2) the combined effect of impurity segregation and the type of carbide present produced a more severe embrittlement, as measured by the shift in the TT, Figures 1 and 3.

The contribution of the types of carbides to temper embrittlement in 2¼%Cr-1%Mo steels is still under investigation. The results obtained to date indicate that tramp element segregation is not uniform, i.e. tramp elements have not been detected at all precipitates. The presence of Sb and Sn has been observed at M_7C_3 carbides. The results of several APFIM analyses of M_3C indicate that some, but not all, precipitates are associated with tramp element segregation. Although additional data must be acquired, the present results are consistent with the observed mechanical behavior of the steels in this investigation and the previous study.(11).

Conclusions

1. A beneficial scavenging mechanism of lanthanide (rare earth) metals with respect to impurities in 2¼%Cr - 1%Mo steel was observed.

2. Additions of pure lanthanum have the most beneficial effect in reducing temper embrittlement; however, additions of lanthanides as mischmetal are also beneficial in reducing temper embrittlement.

3. Fracture mode in temper embrittled 2¼%Cr - 1%Mo steels depends on impurity segregation and the type of alloy carbides present in the steel.

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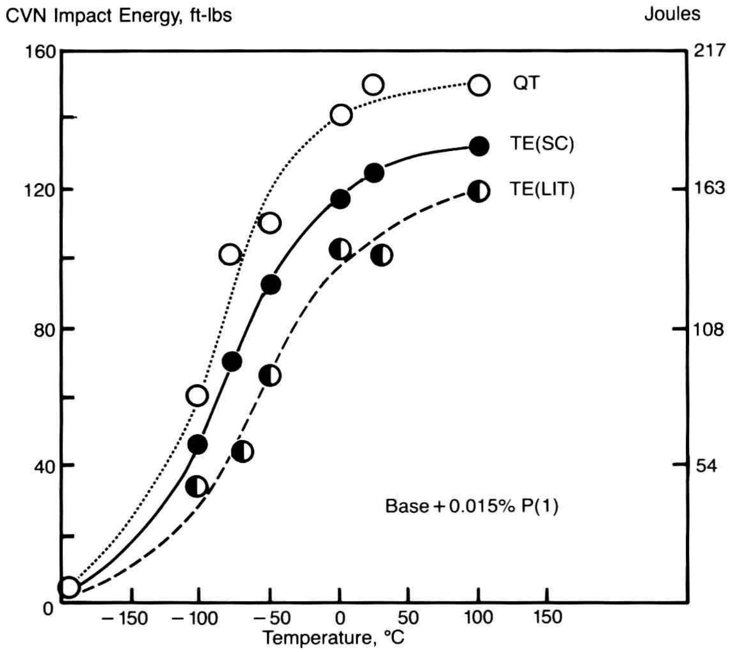

Figure 1. Impact transition curves of $2\frac{1}{4}$%Cr - 1%Mo steel (1) in the quenched and tempered (QT), step-cooled temper embrittled (SC) and long isothermal temper embrittled (LIT) conditions.

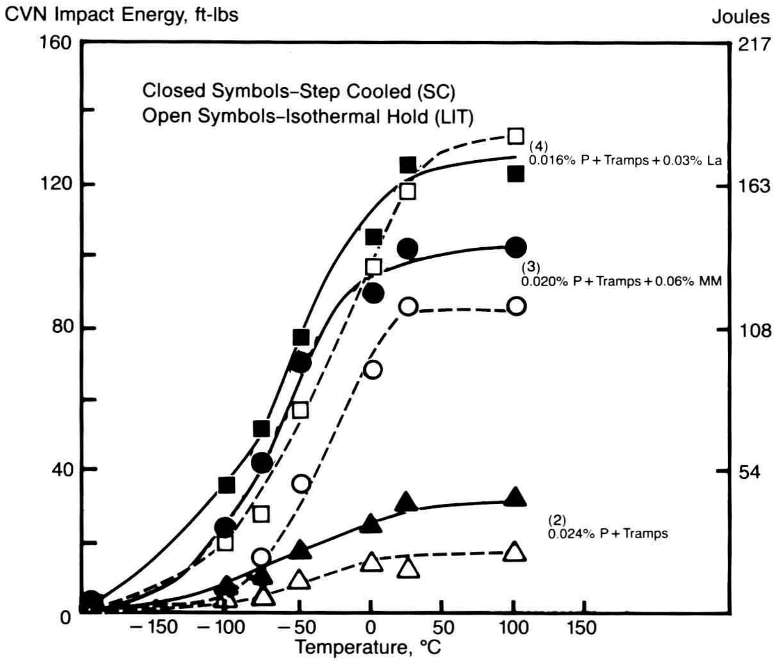

Figure 2. Impact transition curves of $2\frac{1}{4}$%Cr, 1%Mo steels with and without lanthanide (rare earth) metal additions. Temper embrittled by step cooling (SC) and by long isothermal hold for 2000 hours at 520°C (968°F)

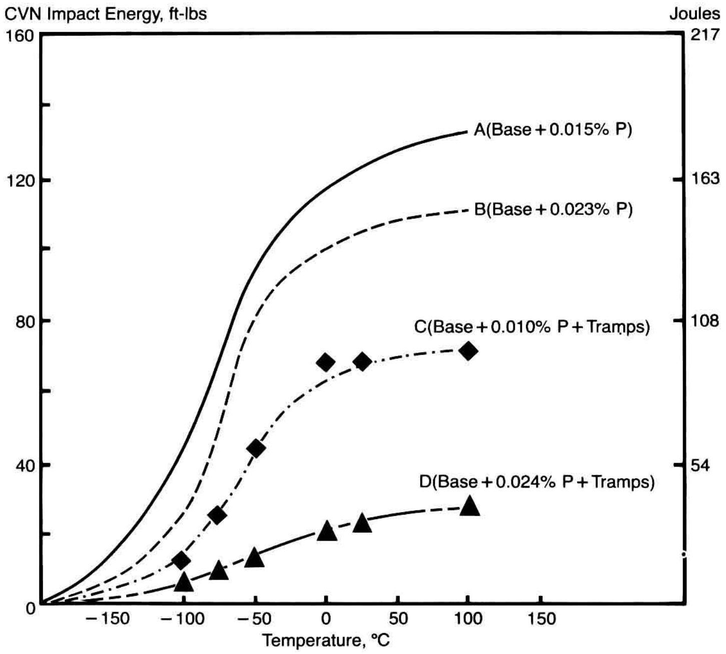

Figure 3. Impact transition curves of $2\frac{1}{4}$%Cr-1%Mo steels in the temper embrittled (TE) condition after a step-cooling treatment showing the effect of phosphorus and tramp elements (Sn, As, Sb).(11)

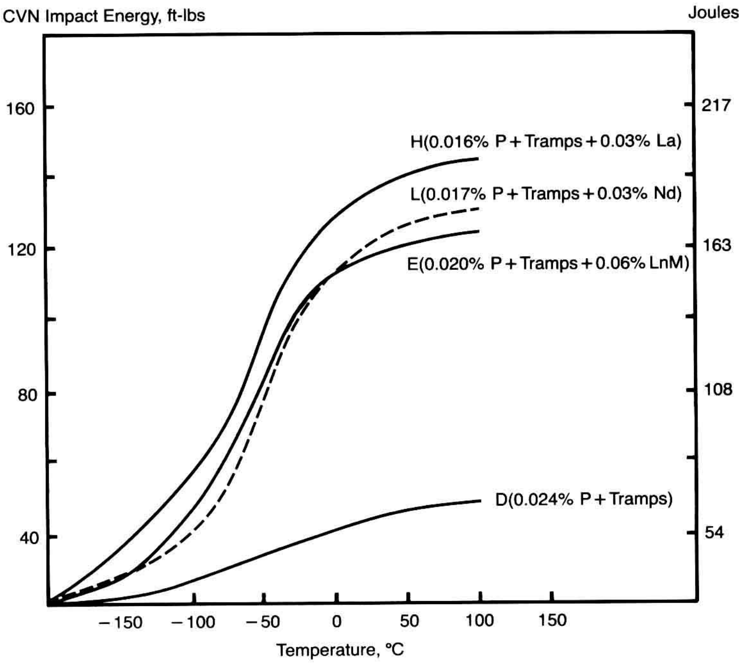

Figure 4. Impact transition curves of $2\frac{1}{4}$%Cr-1%Mo steels in the temper embrittled (TE) condition showing the beneficial effect of lanthanide additions.(11)

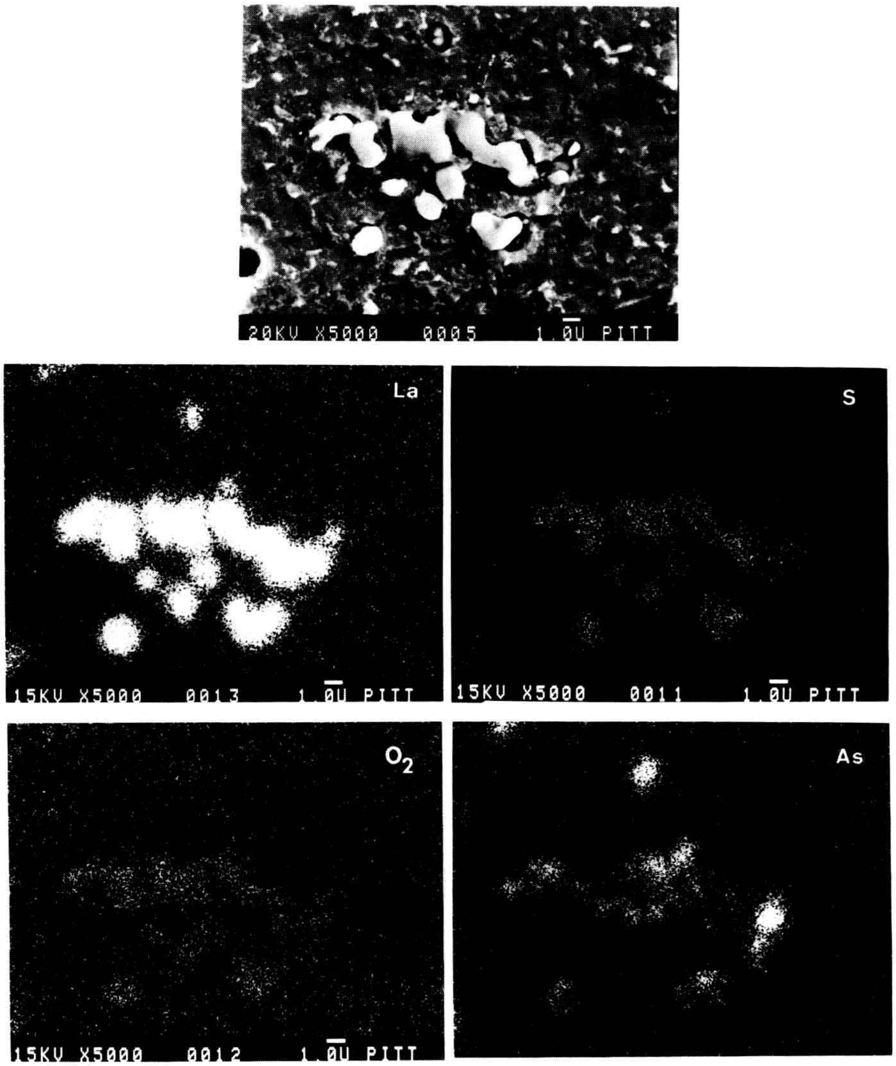

Figure 5. Typical nonmetallic inclusions and corresponding X-ray maps for La, O_2, S and As observed in as-cast $2\frac{1}{4}$%Cr - 1%Mo steel treated with lanthanides (rare earths).

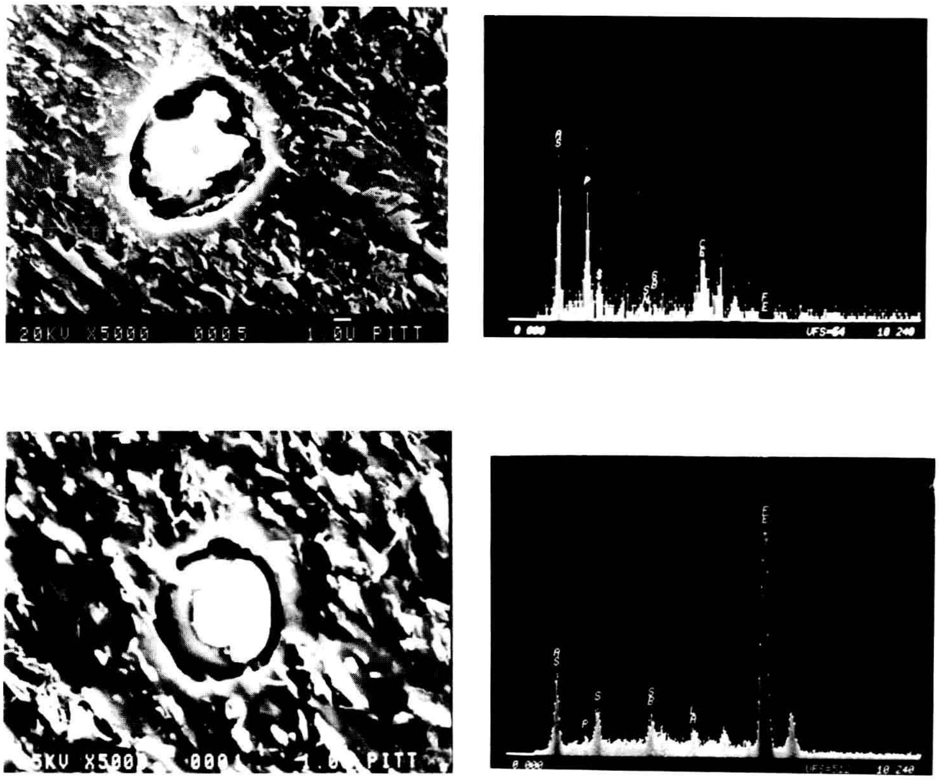

Figure 6. Typical layer-type of morphology of nonmetallic inclusions observed in as-cast 2$\frac{1}{4}$%Cr - 1%Mo steels treated with lanthanide (rare earth) metals.

MOLYBDENUM-CONTAINING ULTRA LOW-CARBON BAINITIC STEELS FOR HEAVY PLATE APPLICATIONS

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ABSTRACT

A new family of molybdenum-containing ultra low-carbon bainitic (ULCB) structural steels has been investigated. These new steels were developed to replace conventional quenched plus tempered (Q+T) high yield strength steels. The results to date have indicated that a proper choice of alloy design and thermomechanical processing can lead to a very attractive family of steels with good mechanical properties in the as-hot rolled condition in sections up to 25.4 mm (1 in.) thick. Studies in progress have indicated that these steels can be used in heavy plate sections up to 100mm (4 in.) with different molybdenum and nickel combinatons than those reported in this paper. Another major advantage of the ULCB steels is that the weldability of these steels and their resistance to underbead cracking should both be considerably improved because of the very low carbon content in these molybdenum-containing ULCB steels. Preliminary weldability studies currently in progress indicate that these new steels have good weldability without the need for pre- and post-heating as that required for Q+T steels.

Introduction

One of the principal types of structural steels used for heavy gage applications has been the HY(high yield) strength family of alloy steels.(1) The HY steels are available in thicknesses up to 200mm (8 in.). These HY steels are hot-rolled to their final thickness, air cooled to room temperature, and then reaustenitized and quenched and tempered to their final microstructure (tempered martensite) and desired mechanical properties. Typically, the HY steels contain a relatively large amount of alloying elements. These elements are required to ensure a level of hardenability adequate to promote the formation of martensite at the midthickness during the quenching of thick plates. Unfortunately, the large amounts of alloying elements required to obtain the proper hardenability also can lead to high carbon - equivalent values. Because it is well known that both the overall weldability and weldment toughness are inversely related to carbon-equivalent values, especially at high carbon contents, the weldability of the HY steels is relatively poor. Hence, obtaining weldment and heat affected zone (HAZ) properties in the HY steels has been a difficult and expensive proposition.

During the last two decades, efforts have been undertaken to provide alternatives to the conventional quenched and tempered HY types of steels for applications which require very high strength, good toughness and good weldability properties. As early as 1967, the potential benefits of a very low carbon (0.03%) bainitic steel were being explored. These early results showed that yield strengths well over 690 MPa (100 ksi) with excellent notch toughness could readily be achieved in an as-rolled 0.03 percent C, Ni-Mo-Nb bainitic steel.(2) Since that time, other low carbon steels have been developed and evaluated. These included the Mn-Mo-Nb(3-5) and the four percent Mn (FAMA) steels.(6) In the late 70's, Nippon Steel Corporation (NSC) announced an ultra-low carbon bainitic steel for large diameter, (2)high pressure linepipe for Arctic applications.(7) The typical composition of this steel (in wt%) was: 0.02C- 2 Mn - 0.4 Ni - 0.3 Mo - 0.04 Nb - 0.02 Ti and 0.001 B. This ULCB steel was developed to exhibit a yield strength in the range 480 MPa to 550 MPa (70 to 80 ksi) in 20 mm (0.8 in.) plate with good low temperature toughness and good weldability.

A major difference between the Q+T steels and the currently used linepipe ULCB steels are section thickness and yield strength. The Q+T steel is produced in thicknesses exceeding 50 mm (2.0 in.), whereas the present ULCB steels rarely exceed 18 mm (0.75 in.) in thickness. Thus, ULCB steels, modified in composition to produce the required mechanical properties in heavy sections, would be desirable. Because ULCB steels can develop high strength and toughness with low carbon content without additional heat treatment, these steels should exhibit good weldability and HAZ toughness.

The work presented in this paper summarizes the initial studies made on a new family of molybdenum-containing ULCB steels as a substitute for Q+T HY steels for heavy gage applications up to 25.4mm (1 in.) thick. Other studies are currently in progress to develop ULCB compositions in plate thicknesses up to 100 mm (4 in.) and greater.

Experimental Procedure

A series of molybdenum-containing ultra low-carbon bainitic steels were studied. The chemical composition (in wt%) of the steels evaluated in this investigation are shown in Table I.

Table 1. Composition of steels evaluated (wt%)*

Steel	C	Mn	Cr	Ni	Mo	Ti	Nb	N	B
1	.018	.93	--	1.03	.97	.011	.051	.0055	.0004
2	.021	.94	1.33	--	.98	.014	.049	.006	.001
3	.021	.99	--	1.41	1.49	.016	.052	.006	.001
4	.018	.98	--	2.03	1.95	.016	.054	.0008	.0013
5	.017	1.01	--	3.15	3.02	.013	.055	.001	.0011

*Other elements: S, P< .005; Si = .20; All steels were Al-killed.

These compositions were selected to insure that a fully bainitic structure could be attained in 25.4mm (1 in.) plates.

These steels were vacuum melted in an induction furnace. The ingot weight and size was 225 kgs (500 lbs) and 200 mm x 200mm x 675mm (8 in. x 8 in. x 27 in.), respectively. The ingots were reheated to 1150°C (2102°F) and deformed 50 percent. Then, three equal sections from these slabs were cut. Each section was reheated at different temperatures prior to final processing: (1) above the grain coarsening temperature (T_{GC}), (2) below the T_{GC} and (3) below the dissolution temperature of niobium carbonitrides, Nb(CN). After the reheating treatment, the sections were hot rolled at the same temperature range, subjected to the same amount of deformation and then air cooled to room temperature. Figure 1 is a schematic representation of the thermomechanical processing used.

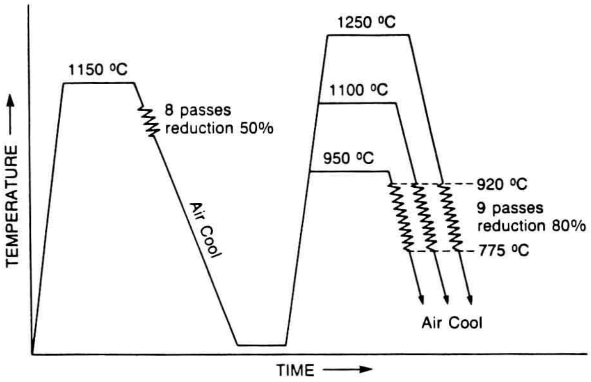

Figure 1. Thermomechanical processing of ULCB steels.

The hardenability of these steels was sufficient to attain a fully bainitic structure in the plate thickness after air cooling. The bainitic start transformation temperature is nearly independent of cooling rate.

Standard mechanical properties (strength, hardness and impact toughness) of the processed steels were determined. Sub-size flat tensile specimens with a 25.4mm (1 in.) gage length from the transverse direction were prepared and tested. Also, full size Charpy V-notch specimens transverse to the rolling direction were tested. The position of the V-notch was perpendicular to the rolling plane.

Steel specimens were tempered for two hours in the temperature range 500°C (1022°F) to 675°C (1247°F) at 25°C (45°F) intervals, and the specimens were then air cooled to room temperature to assess the influence of tempering treatment on the mechanical properties.

Results

The tensile properties of the steels investigated are given in Table II. Only the properties of the steels that met the objectives of this study are included for brevity of the paper.

All the steels evaluated in this investigation were designed to achieve a fully bainitic structure after thermomechanical processing and air cooling. However, the resultant chemical composition for steel 1 showed a lower B content and Ti/N ratio than those required, (Table I).

The data from Table II revealed that excellent tensile strength values were obtained for steels 2 through 5 (818/1073 MPa) when compared to that of a HY-80 steel (699 MPa). Steel number 1 (reheated at 1100°C) had a slightly lower strength value (671 MPa) than HY-80 steel. A possible explanation for the behavior for steel 1 was found in the microstructure observed in this steel. The micro-structure of steel 1 consisted of a mixture of bainite plus pro-eutectoide ferrite.