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An Understanding of HSLA-65 Plate Steels

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HSLA-65 plate steels can be produced using one of five plate manufacturing techniques: normalizing, controlled rolling (CR), controlled rolling followed by accelerated cooling (CR-AC), direct quenching and tempering (DQT), or conventional quenching and tempering (Q&T). The HSLA-65 steels are characterized by low carbon content and low alloy content, and they exhibit a low carbon equivalent that allows improved plate weldability. These characteristics in turn (a) provide the steel plate with a refined microstructure that ensures high strength and toughness; (b) eliminate or substantially reduce the need for preheating during welding; (c) resist susceptibility to hydrogen-assisted cracking (HAC) in the weld heat affected zone (HAZ) when fusion (arc) welded using low heat-input conditions; and (d) depending on section thickness, facilitate high heat-input welding (about 2 kJ/mm) without significant loss of strength or toughness in the HAZ. However, application of this plate manufacturing process and of these controls produces significant differences in the metallurgical structure and range of mechanical properties of the HSLA-65 plate steels both among themselves and versus conventional higher strength steel (HSS) plates. For example, among the HSLA-65 plate steels, those produced by Q&T exhibit minimal variability in mechanical properties, especially in thicker plates. Besides variability in mechanical properties depending on plate thickness, the CR and CR-AC plate steels exhibit a relatively higher yield strength to ultimate tensile strength (YS/UTS) ratio than do DQT and Q&T steels. Such differences in processing and properties of HSLA-65 plate steels could potentially affect the selection and control of various secondary fabrication practices, including arc welding. Consequently, fabricators must exercise extreme caution when transferring allowable limits of certified secondary fabrication practices from one type of HSLA-65 plate steel to another, even for the same plate thickness.

Keywords arc welding, heat treatment, HSLA-65 plate steels, mechanical properties, microstructure, plate manufacturing, secondary fabrication practices

1. Introduction

Currently, HSS plates used in structural and machinery applications of naval surface ships are procured to military specifications such as MIL-S-22698 (ordinary and higher strength steels). HSS contains low to medium carbon (up to 0.18 wt.%) and small amounts of alloying elements (Table 1). These steels are commonly processed by conventional hot rolling (in the austenitic range above Ar_3 temperature, i.e., α-γ transformation temperature) and/or normalizing to achieve the desired mechanical properties. The most commonly used grade, MIL-S-22698, Grade DH-36, is specified with a minimum of 352 MPa (51 ksi) yield strength and 490 MPa (71 ksi) ultimate tensile strength (Table 2). Furthermore, MIL-S-22698 requires aluminum-treated steels to be normalized when the thickness exceeds 25.4 mm (1 in.) and columbium- or vanadium-treated steels to be normalized when the thickness exceeds 12.7 mm (½ in.), thereby indicating possible adverse effects of microalloying additions on processing, microstructure development, and mechanical properties.

Although MIL-S-22698 allows CR or thermomechanical controlled processing (TMCP) of Grade DH-36 as a substitute for normalizing, the American Bureau of Shipping (ABS) does not allow automatic substitution of CR grade for TMCP grade. ABS differentiates between CR and TMCP as follows (Ref 1): CR involves final rolling in the temperature range traditionally used for normalizing (above both A_3 and γ-recrystallization temperature), which results in recrystallized austenite. TMCP involves a high proportion of rolling reduction near or below A_3 temperature. This results in nonrecrystallized austenite and may be followed by AC. TMCP steels that are controlled rolled and subsequently accelerated cooled require special approval of the Bureau to be accepted as a substitute for normalized grade. This requirement is perhaps based on the premise that one must exercise extreme caution when transferring the application of certified secondary fabrication practices from one type of plate steel to another, even when plate thickness remains the same, due to (a) inherent differences introduced in the range of mechanical properties through differences in processing and microstructure and (b) a need to control secondary processing methods to retain or recreate microstructures with performance characteristics similar to those of the base metal.

1.1 HSLA Steels

In recent years, several domestic and foreign steel producers have used a variety of advanced manufacturing techniques to achieve significant improvements in strength, toughness, formability, and weldability of new grades of high-strength plate steel over that of HSS grades. These advanced manufacturing technologies include the use of:

- Lower carbon content, lower alloy content, and control of residual elements
- Microalloy (columbium or niobium, titanium, and vanadium) additions

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Table 1 MIL-S-22698 chemical composition (in wt.%) requirements of higher strength steel (HSS) plate

C	Mn	P	S	Si	Ni	Cr	Mo	Cu	Cb	V
0.18 max	0.90-1.60	0.04 max	0.04 max	0.10-0.50	0.40 max	0.25 max	0.08 max	0.35 max	0.05 max	0.10 max

Table 2 MIL-S-22698 mechanical property requirements of DH-36 steel plate

	Tensile properties(a)			CVN impact properties at −20 °C (−4 °F)	
Plate thickness	YS, MPa (ksi)	UTS, MPa (ksi)	EL(b), %	Longitudinal, J (ft-lb)	Transverse, J (ft-lb)
≤50 mm (2 in.)	352 (51) min	490-621 (71-90)	19 min	34 (25)	23 (17)

(a) For plates wider than 610 mm (24 in.), the tensile test specimen is taken in the transverse direction. (b) Minimum elongation is 19% in 203 mm (8 in.) gauge length specimen or 22% in 50 mm (2 in.) gauge length specimen.

- Controlled rolling (with final finish rolling near Ar_3 temperature)
- Accelerated cooling after controlled rolling
- Direct quenching and tempering after controlled rolling
- Inclusion shape control

Steels produced using the methods above are variously referred to as high-strength low-alloy (HSLA) steels, TMCP steels, CR steels, AC steels, or DQT steels. These steels exhibit a fine-grained microstructure that provides a superior combination of high-strength and excellent toughness. Depending on the structural applications, the increases in the strength level of these steels (over HSS) allow either an increase in design stresses or an appropriate reduction in section thickness. These steels have been commercially available with yield strengths up to 552 MPa (80 ksi) for structural plate applications (e.g., ASTM A656; Ref 2) and line pipe (e.g., API 5L; Ref 3). Recently, the U.S. shipbuilding industry has targeted the development and certification of HSLA-65 plate steel to replace (where feasible) DH-36 HSS plate steel (MIL-S-22698) that is widely used in ship construction. The motivations for the deployment of HSLA-65 plate steel are twofold: (a) initial cost savings through reduced fabrication costs, and (b) life-cycle cost savings through weight reduction. Use of HSLA-65 plate steel is expected to allow designers to use a thinner gauge plate steel that is stronger than HSS plate without having to use the next higher strength grade plate steel (i.e., HY-80 or HSLA-80) that is already certified for naval construction. The HY-80 or HSLA-80 plate steels are also much costlier to procure and fabricate.

To allow the widespread use of HSLA-65 plate steels in structural applications, including ship structures, the following aspects must be addressed:

- Availability of production quantities of HSLA-65 plate steels (up to 32 mm, or 1¼ in., thick)
- Detailed testing and evaluation to identify appropriate fabrication practices
- Approval and certification based on acceptable test results

ASTM International has issued a new material specification A945 (Ref 4) for "High-Strength Low-Alloy Structural Steel Plate with Low Carbon and Restricted Sulfur for Improved Weldability, Formability, and Toughness," with a 448 MPa (65 ksi) minimum yield strength and 538 MPa (78 ksi) minimum tensile strength corresponding to HSLA-65 grade. Several domestic and foreign steel producers have demonstrated the capability to provide HSLA steel as either line pipes or plates processed using CR, AC, DQT, or conventional Q&T manufacturing technologies, with yield strengths of 448 MPa (65 ksi) and higher, in sections up to 32 mm (1¼ in.) in thickness.

Initial applications of HSLA-65 plate steels for hull structures will primarily use arc welding and flame straightening procedures qualified for HSS per MIL-STD-1689. For the longer term, the HSLA-65 steel plate may also be used in other fabricated components, such as machinery, in which additional fabrication practices such as hot and cold forming, normalizing, and post-weld stress relieving may be used. These practices may be implemented by vendors or shipyards in the fabrication of machinery bases and machinery components or during ship overhaul and repair using additional fabrication standards such as MIL-STD-278.

From a metallurgical perspective, there are concerns that the superior combination of high strength and excellent toughness of the HSLA-65 plate steels may be substantially degraded if subjected to the same allowable procedures for HSS plate steels. For example, the reaustenitization during hot forming, normalizing, or abusive flame straightening may eliminate the fine-grained microstructure of the HSLA-65 steel plate, and thus adversely alter the mechanical properties. The cold forming procedures may further increase the relatively high yield strength/ultimate tensile strength (YS/UTS) ratio of certain HSLA-65 plate steels, and thereby decrease the structural stability of these steels during certain types of service loading (Ref 5).

Further, the HSLA-65 plate steels containing columbium, vanadium, and titanium microalloy additions may experience embrittlement during stress relieving (Ref 6), especially in the coarse-grained region of the weld heat-affected zone (HAZ). The HAZ of weldments in low-carbon HSLA-65 plate steels may experience excessive softening when welded using excessively high heat-input (over 3 kJ/mm, or 80 kJ/in.) conditions (Ref 7). These considerations suggest that appropriate limits on fabrication practices must be established to prevent possible degradation of mechanical properties, based on an evaluation of the effects of such fabrication practices on the microstructure and properties of HSLA-65 plate steels.

To address these issues, this technical article provides an overview of the metallurgy of HSLA-65 plate steels and how it impacts the selection and control of secondary fabrication practices.

2. Metallurgy of HSLA-65 Plate Steels

ASTM A945 specification (Ref 4) covers HSLA-50 plate steels up to 50 mm (2 in.) thick and HSLA-65 plate steels up to 32 mm (1¼ in.) thick. Both of these grades are characterized

Table 3 ASTM A945 chemical composition (in wt.%) requirements of HSLA-65 steel plate (single values are maxima)

C	Mn	P	S	Si	Ni	Cr	Mo	Cu	V	Cb	Al
0.10	1.10-1.65	0.025	0.010	0.10-0.50	0.40	0.20	0.08	0.35	0.10	0.05	0.08

by a low carbon content (up to 0.10 wt.% max) and restricted sulfur content (0.010 wt.% max), and are made to a killed-steel fine-grain practice. These characteristics provide the HSLA plate steels with significant improvements in weldability, formability, and toughness over conventional hot-rolled or normalized plate steels. The A945 specification requires that the HSLA plate steels be provided in one of the following three conditions: as-finished, normalized, or Q&T. The as-finished types include as-rolled; thermo-mechanical controlled rolled; and accelerated cooled.

2.1 Chemical Composition and Processing

Table 3 provides the chemical composition requirements of HSLA-65 steel plate per A945 specification. These requirements are intrinsically related to the effects of processing conditions on microstructure development and mechanical properties, including formability and weldability. For example, to obtain improved weldability, the carbon content of HSLA-65 plate steels is limited to 0.10 wt.% (Ref 8, 9). The manganese content ranges from 1.10 to 1.65 wt.% to provide adequate strength and toughness. Silicon is added in controlled amounts to provide deoxidation and solid solution strengthening. Nickel, chromium, molybdenum, and copper may be present as residual elements or as deliberate additions to provide additional strengthening, but maximum limits are imposed to prevent excessive increases in strength and hardenability or substantial reductions in weldability.

A low sulfur content (0.010 wt.% max) reduces the occurrence of soft manganese sulfide inclusions and thereby improves formability (Ref 8) and upper shelf notch toughness. In wrought plate steels, excessive sulfur tends to form elongated Type II MnS stringers, which reduce through-thickness ductility (i.e., along the Z-direction) and thereby reduce the fracture resistance of the plate steels in complex loading situations as well.

As shown in Table 3, the HSLA-65 plate steels may contain columbium, vanadium, and aluminum additions. Although not specified, titanium may also be added. Columbium, vanadium, and titanium microalloy additions form solid solution in austenite, increase the austenite recrystallization temperature (Ref 10), and thereby limit austenite recrystallization and grain growth during controlled rolling above the Ar_3 temperature. Columbium is more effective in limiting austenite recrystallization compared with vanadium or titanium (Ref 8, 10). The microalloy additions also form columbium and titanium carbonitrides and vanadium nitrides or carbides (Ref 8, 10), which pin austenite grain boundaries during controlled rolling and also provide precipitation strengthening. The carbide-forming tendency decreases from columbium, titanium, vanadium. molybdenum, chromium, to manganese. However, the occurrence of TiC invariably results in poor toughness. In controlled amounts (0.01-0.02 wt.%), titanium fixes solute nitrogen as TiN. During fusion welding, the TiN precipitates limit austenite grain growth in the weld HAZ and thereby limit excessive increases in hardenability or substantial decreases in HAZ strength and toughness.

Aluminum serves as both a deoxidant and a grain refiner. It is also used to fix solute nitrogen as AlN to minimize strain aging in plate material and to accelerate austenite to ferrite transformation, thereby improving the recovery of toughness in the weld HAZ (Ref 11). Excessive aluminum addition reduces base metal toughness.

Controlled rolling practice involves rough rolling above the austenite recrystallization temperature followed by finish rolling near the Ar_3 temperature. Rough rolling performed above the γ-recrystallization temperature is typically used to achieve approximately 60% reduction in thickness, while finish rolling near the Ar_3 temperature is used to achieve the remaining 40% reduction in thickness (Ref 8). Finish rolling provides plastically deformed austenite (γ) grains that are characterized by more ferrite (α) nucleation sites (Ref 8). Subsequent transformation of the deformed austenite to fine-grained equiaxed (polygonal) ferrite significantly increases the strength and toughness of these controlled rolled steels compared with conventional hot rolling or normalizing.

Figure 1 (Ref 12) schematically illustrates the effects of CR conditions on the (ultimate) tensile strength, fracture appearance transition temperature (FATT) and ferrite grain size (d_α). Excessive deformation in the intercritical range below Ar_3 can result in work hardening of the newly formed ferrite grains and can cause deterioration in the low-temperature toughness (i.e., raise FATT).

AC after CR reduces ferrite grain size, increases both strength and toughness at reduced carbon equivalent number, and thereby improves weldability (Ref 12). After controlled rolling above the Ar_3 temperature, accelerated cooling is required to preclude the formation of upper bainite from the ferrite nucleation sites in the deformed austenite. Figure 2 schematically illustrates the effects of AC conditions on tensile strength and FATT. Excessive rapid cooling of these steels can cause undesirable microstructures to form and thereby deteriorate low-temperature toughness (i.e., raise FATT).

DQT involves hot rolling (above the austenite recrystallization temperature) followed by quenching of the recrystallized austenite prior to the start of the austenite to ferrite transformation (Ref 13, 14). Depending on carbon content, alloy content, and cooling rate, either martensite, a mixture of martensite and bainite, or bainite may form. Subsequent tempering at 450-700 °C (~850-1250 °F) produces a network of carbides in a polygonal ferrite matrix. HSLA steels produced by DQT processing exhibit a higher hardenability compared with those produced by conventional Q&T processing. The increase in hardenability is attributed to higher hot rolling temperature, which takes alloy additions into solution, increases austenite grain size, and minimizes the rate of austenite to ferrite transformation (Ref 14). The tempering temperature primarily affects the impact toughness properties (Ref 14).

Table 4 summarizes the effects of various alloy additions on both processing and properties within the composition range specified for HSLA-65 plate steels. Table 4 also includes the effects of titanium, although titanium addition is not specified in A945 specification. The Ar_3 temperature of low-carbon mi-

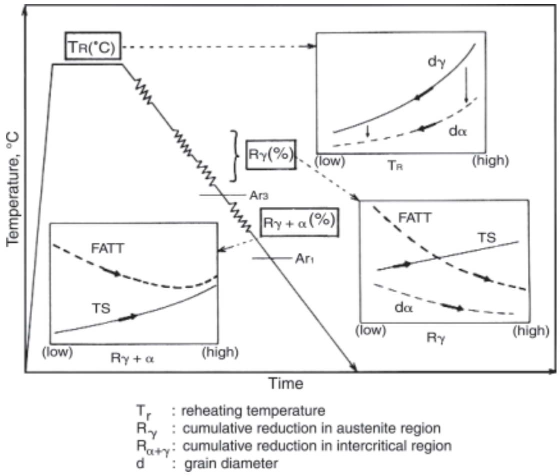

Fig. 1 Schematic diagram illustrating the effects of controlled rolling conditions on (ultimate) tensile strength, FATT, and ferrite grain size: T_R, reheating temperature; R_γ, cumulative reduction at temperatures above Ar_3; $R_{\gamma+\alpha}$, cumulative reduction in the intercritical region (i.e., between Ar_3 and Ar_1); d, grain diameter (Ref 12)

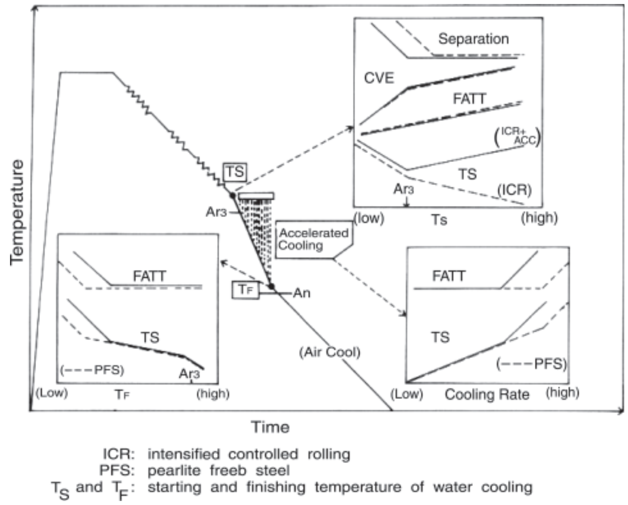

Fig. 2 Schematic diagram illustrating the effects of accelerated cooling conditions on (ultimate) tensile strength and FATT: T_S and T_F, starting and finishing temperature of water cooling; ICR, intensified controlled rolling; ACC, accelerated cooling; PFS, pearlite-free steel (Ref 12)

croalloyed steel plates is statistically related to their chemical composition as follows:

$$Ar_3\ (°C) = 910 - 310(C) - 80(Mn) - 80(Mo) - 55(Ni) - 20(Cu) - 15(Cr) + 0.35(t - 8)$$

where elemental content is in wt.% and t is the plate thickness in mm (Ref 15). This statistical relationship is valid for t = 8-30 mm. Lowering the Ar_3 temperature enables the steelmakers to further refine both matrix grains and various microstructural constituents, thereby allowing them to improve both strength and toughness simultaneously. Indiscriminate lowering of Ar_3 temperature might substantially reduce the operational envelopes for both plate rolling and various secondary fabrication practices, besides loss of control over microstructure development.

2.2 Microstructure

The room temperature microstructure of HSLA steel plates depends on manufacturing technique and section thickness. The CR plate steels contain bands of refined, equiaxed ferrite grains and lamellar pearlite (Ref 8, 16). Bethstar plate steels, formerly produced by Bethlehem Steel Corporation to meet ASTM A656 Type 7 requirements, are typical examples of this type of HSLA steel. The ferrite grains typically range in size between ASTM 10 and 12. Both the ferrite and pearlite often appear as bands. The pearlite content increases with increasing carbon content and section thickness (Ref 16).

In thin plates, much of the ferrite is characterized by an equiaxed morphology. The equiaxed morphology of the ferrite is related to the severe reduction in thickness during the final finish rolling in the intercritical region near Ar_3 temperature. Severe reductions generally reduce the ferrite recrystallization temperature and thereby allow the formation of equiaxed ferrite, even when the final finishing temperature is low (i.e., the elongated ferrite formed during finish rolling undergoes recrystallization to form equiaxed ferrite).

In comparison, thicker plates generally contain limited amounts of equiaxed (polygonal) ferrite and show larger amounts of elongated ferrite with substructure. Subsequent to finish rolling below the γ-recrystallization temperature or below the Ar_3 temperature, these steel plates exhibit deformed austenite grains containing deformation bands and deformed ferrite grains containing a subgrain structure. On cooling, the deformed austenite transforms to equiaxed ferrite, whereas the deformed ferrite experiences minimal recovery and shows a subgrain structure with deformation bands. Such microstructures are sometimes referred to as "warm-worked ferrite" (Ref 12, 16). In the latter case, the ferrite recrystallization temperature is likely higher than the final finish rolling temperature.

The combined presence of a banded structure and elongated warm worked ferrite network in high-strength line pipe steels has occasionally caused splitting-type fracture during Charpy V-notch (CVN) impact testing (Ref 17). Splits were also observed on fracture surfaces of full-scale tests of line pipe made from these steels. The splitting-type fracture was observed in CR steels that contained columbium and were finish rolled at temperatures well below the Ar_3 temperature. These steels also showed a lowering of the upper shelf energy during CVN impact testing. The splits were observed in a direction normal to the primary fracture plane of the CVN test specimen and parallel to the plane of the plate surface. Light microscopy examination of the cross section of the split did not show any evidence of plastic deformation. Speich and Dabkowski attributed splitting to a crystallographic texture effect that arises when rolling is performed at temperatures significantly below the Ar_3 temperature (Ref 18).

The CR-AC HSLA steels often contain refined polygonal ferrite grains (finer than the normalized grade) with some deformation bands or substructures and uniformly distributed refined bainite instead of lamellar pearlite. In general, the bainite content increases with increasing columbium content (Ref 19, 20). These steels are also referred to as acicular ferrite steels or pearlite-free (PF) steels (Ref 6, 19).

The as-rolled (Ref 20) and normalized grades exhibit pearlite colonies in a relatively fine-grained, equiaxed ferritic matrix. These microstructures provide excellent toughness. In

Table 4 Effects of alloy additions on processing and properties of HSLA-65 steel

Alloy element	Effect(s) of alloy addition
Carbon, C	Suppresses the γ-α transformation (Ar_3) temperature; provides strength and toughness; excessive amounts (>0.10 wt.%) increase hardenability and reduce weldability
Manganese, Mn Molybdenum, Mo Nickel, Ni Copper, Cu Chromium, Cr	Suppress the γ-α transformation (Ar_3) temperature; provide strength and toughness; excessive amounts increase hardenability and reduce weldability
Sulfur, S	A low sulfur content (0.010 wt.% max.) reduces the occurrence of manganese sulfide inclusions but improves formability and upper shelf notch toughness
Phosphorus, P	Causes embrittlement during stress relief or slow cooling from rolling temperatures or heat treatment
Columbium, Cb	In controlled amounts (0.02-0.05 wt.%) increases austenite recrystallization temperature; provides strengthening by forming thermally stable, Cb(C,N) and Cb,Ti(C,N) precipitates; during fusion welding, the precipitates limit austenite grain growth in the weld HAZ, and thereby limit hardenability or improve weldability; excessive amounts (>0.05 wt.%) can potentially impair HAZ toughness in high heat-input weldments
Vanadium, V	Increases austenite recrystallization temperature; provides room temperature strengthening by forming VN, V(C,N), and (V,Ti)N precipitates in ferrite
Aluminum, Al	Deoxidant and grain refiner; in controlled amounts, fixes solute nitrogen as AlN and thereby improves recovery of HAZ toughness and weldability; in excessive amounts (>0.08 wt.%) reduces base metal toughness
Titanium, Ti	In controlled amounts (0.01-0.02 wt.%), acts as a grain refiner, increases austenite recrystallization temperature, fixes solute nitrogen as TiN, and provides strengthening by forming thermally stable, complex Ti(C,N) precipitates; during fusion welding, TiN precipitates limit austenite grain growth in the weld HAZ, thereby limiting hardenability and improving HAZ strength and toughness; precipitation of TiC invariably reduces HAZ toughness, especially at low temperatures

comparison with other HSLA grades, the as-rolled and normalized grades provide a lower strength at comparable carbon equivalent (Ref 21). A reduction in strength at comparable carbon equivalent perhaps occurs due to a relatively larger ferrite grain size, while a concurrent improvement in toughness may be related to the simultaneous presence of isolated pearlite colonies.

The DQT and Q&T steels exhibit a relatively fine-grained bainite, mixed structure containing bainite and martensite, or tempered martensite depending on carbon content, specific alloy additions, and quenching rate (Ref 13). Fine-sized intragranular and intergranular carbide particles are normally present. The carbide particles along grain boundaries often show a blocky morphology. The tempering temperature primarily affects the size distribution of the intragranular particles and the morphology of the grain boundary particles and does not normally affect the matrix grain size.

2.3 Mechanical Properties

Table 5 shows the mechanical property requirements of HSLA-65 steel plate up to 32 mm (1¼ in.) thickness, per A945 specification.

The heavier gauge HSLA steels produced using CR or CR-AC techniques may sometimes show unacceptable variability in mechanical properties from plate to plate and occasionally within the same plate. Normally, these steels, irrespective of plate thickness, exhibit a higher YS/UTS ratio approaching 0.93, thereby indicating a minimal strain hardening capability (Ref 5, 15). Compared with the CR and CR-AC HSLA steels, the conventional Q&T grades of steels provide more uniform mechanical properties, especially in higher thicknesses. Microalloyed HSLA steels produced using the DQT and Q&T techniques exhibit a comparatively low YS/UTS ratio (Ref 13, 21) typically ranging from 0.80 to 0.90. In certain applications such as those involved in seismic sensitive regions, the DQT and Q&T steels may be preferred over CR or CR-AC steels.

The various types of HSLA-65 steels often show a very high CVN impact toughness, 200 J (150 ft-lb) at −40 °C (−40 °F), in their as-received condition, much in excess of A945 specification requirements. It is unclear why the A945 specification requires only a minimum of 41 J (30 ft-lb) at −40 °C (−40 °F) for CVN impact toughness when the various plate manufacturing techniques can consistently provide 4-6 times higher values. Perhaps, the A945 specification requirements are based on the CVN impact toughness requirements for comparable steels (see Tables 2 and 5). Despite this, plate procurement specifications must take additional effort to require high CVN impact toughness values when placing purchase orders for the supply of various HSLA-65 plates in their as-received condition. Otherwise, when plates that barely meet the A945 specification requirements for CVN impact toughness are procured for subsequent fabrication of components, the various secondary fabrication methods may not have the luxury of a large toughness "cushion" to work with. This might translate into severe fabrication difficulties involving higher repair and rejection rates, and possibly lead to unacceptable increases in component manufacturing costs.

3. Secondary Fabrication Practices

3.1 Arc Welding

The low carbon content, low alloy content, and low hardenability of the HSLA-65 plate steels generally improve the resistance of these steels to HAC in the weld HAZ even when arc welded using low energy input (i.e., fast cooling rates). Besides energy input, the susceptibility of these steels to HAC in the weld HAZ is dependent on the alloy content or carbon equivalent number (CEN) of the base metal and the hydrogen content of the welding process and consumables. HAC is known to occur especially in high-strength structural steels that have the potential to form high-carbon (over 0.12 wt.%) twinned martensite when the following conditions are present *simultaneously:* (a) a source of dissolved hydrogen; (b) a susceptible (martensitic) microstructure; (c) high residual tensile stress distribution; (d) a temperature range that does not allow significant solid-state diffusion of atomic hydrogen from the

Table 5 ASTM A945 mechanical property requirements for HSLA-65 steel plate

Tensile properties(a)			CVN impact properties at −40 °C (−40 °F)	
YS, MPa (ksi)	UTS, MPa (ksi)	EL(b), %	Longitudinal, J (ft-lb)	Transverse, J (ft-lb)
448 (65) min	538-690 (78-100)	18 min	41 (30)	27 (20)

(a) For plates wider than 610 mm (24 in.), the tensile test specimen is taken in the transverse direction. (b) Minimum elongation is 18% in 203 mm (8 in.) gauge length specimen or 22% in 50 mm (2 in.) gauge length specimen. For plates wider than 610 mm (24 in.), the elongation requirement is reduced by two percentage points.

steel; and (e) a time delay after welding that allows atomic hydrogen to accumulate at internal "flaws" in the steel.

Commonly, either one of two different approaches, such as a critical hardness approach or a critical hydrogen approach, is used in selecting an arc welding procedure to reduce the HAC susceptibility of structural steels (Ref 22, 23).

The critical hardness approach relates the application of preheat and interpass temperature controls to the need for reducing the hardness of the coarse-grained region of the HAZ below a critical level. However, this critical hardness level has been shown to decrease with decreasing carbon content and decreasing CEN (Ref 24). A low carbon content and a low CEN should in general reduce the hardenability of a steel and can be expected to produce base metal microstructures with a low hardness. However, because the HSLA steels that are low in sulfur and inclusion content provide only a minimal number of sites for ferrite nucleation in the HAZ, a microstructure with a higher hardness or hardenability may occur within the weld HAZ (Ref 24).

The critical hydrogen approach relates the application of preheat, interpass, and, when necessary, postsoak temperature controls to reduce the diffusible hydrogen content of the weld metal (or weldment) below a critical level. This critical level is dependent on steel chemical composition, thickness, weld joint design, weld heat input, and weld restraint.

In either approach, besides weld energy input, the necessary preheat, interpass, and post-soak temperature controls are determined by the following parameters: (a) hydrogen content of the welding process and consumables; (b) diffusible hydrogen content of the weld metal; (c) carbon equivalent number (CEN) of the base metal; (d) section thickness and weld joint design; and (e) weld restraint.

Different types of statistically determined, empirical carbon-equivalent formulas that correlate steel type, chemical composition, and welding conditions (including restraint) with either the maximum hardness of the coarse-grained region of the HAZ or the width of the softened HAZ region are reported in literature. These carbon-equivalent formulas relate the chemical composition of a wide variety of structural steels and weld metal or HAZ cooling rate with the maximum hardness in the HAZ to estimate the HAC susceptibility of steel.

It must be recognized that many of these formulas were developed for conventional structural steels and are not particularly suitable for evaluating the HAC susceptibility of HSLA steels (Ref 6). As mentioned before, the availability of minimal sites for ferrite nucleation within the HAZ locally increases the hardenability of the low inclusion content HSLA steels, thus many of the commonly available carbon equivalent formulas relating structural steel chemical composition with HAZ cooling rate and maximum hardness in the HAZ often do not provide meaningful results. Consequently, an evaluation of the suitability of a specific carbon equivalent formula for a given type of HSLA steel is often required. For example, in evaluating the applicability of five different carbon equivalent formulas to 12 to 25 mm (½ to 1 in.) thick Q&T HSLA steels, Zaczek and Cweik (Ref 25) showed that the Lorenz and Dueren carbon-equivalent formula can be used to estimate the maximum HAZ hardness when the cooling time between 800 and 500 °C (1472 and 932 °F) is about 6 s.

Furthermore, even when an appropriate carbon equivalent formula has been identified, should the plate manufacturing procedures indicate substantial plate to plate variations in carbon equivalent number and yield strength, then frequent and periodic verification of the applicability of the carbon equivalent formula will be required. On the basis of evaluation of the suitability of a specific carbon-equivalent formula for a given type of HSLA steel, one may be able to determine the welding conditions necessary to lower the hardness of the HAZ below a critical limit.

In contrast to the HAC susceptibility of low heat-input welding situations of low-carbon HSLA steels, the high heat-input welding situations may present significant loss of strength and toughness in the HAZ. Furthermore, excessive increases in weld energy input also widen the width of the softened HAZ (Ref 6, 7, 25-27). For example, Lundin et al. (Ref 26, 27) have determined that, in the case of AC-50 steel (a low-carbon steel containing manganese additions that has been controlled rolled and control cooled to produce a banded structure containing ferrite and pearlite), the width of the soft zone was related to the cooling between 800 and 500 °C (1472 and 932 °F), and the degree of softening was related to the plate chemical composition. While Lundin et al. (Ref 26, 27) related the cooling time between 800 and 500 °C (1472 and 932 °F) to welding conditions (heat input) and plate thickness through the Uwer and Degenkolbe equation, they related the degree of softening to the plate chemical composition through Yurioka's equation for CEN.

In HSLA steels, HAZ softening occurs when high heat-input conditions are used, as limited austenite grain growth produces only a marginal increase in hardenability. The limited hardenability of these steels is primarily related to the minimal alloy content. In addition, losses in strength and toughness occur when the strengthening precipitates dissolve. Dissolution of carbonitrides releases nitrogen into the matrix that can be detrimental to the low-temperature impact toughness. The loss of toughness in the HAZ has been attributed to the effects of coarse grain size, formation of higher volume fractions of upper bainite and martensite-austenite (M-A) islands, and the precipitation of columbium and vanadium carbonitrides in the coarse-grained regions of the HAZ, which experience temperatures near Ac_3 (Ref 12, 28).

While base metal composition control and use of welding conditions that minimize the formation of upper bainite are necessary to reduce the loss in HAZ toughness, in particular, a reduction in carbon, columbium, and soluble nitrogen content is recommended to minimize the effects of these elements on HAZ toughness. In extremely low-carbon, columbium-containing HSLA steels, HAZ grain boundary embrittlement can occur due to intragranular precipitation of Cb(C,N) (Ref 6). The addition of controlled amounts of aluminum (0.06-0.08 wt.%) and titanium (0.01-0.02 wt.%) to combine with soluble nitrogen to form AlN (Ref 19) and TiN (Ref 28) particles has been shown to improve toughness recovery in the HAZ. The strain associated with the precipitation of AlN is believed to accelerate austenite to ferrite transformation (Ref 13, 21) and thereby reduce hardenability. Sawhill et al. (Ref 29) have reported that titanium addition in combination with columbium addition improves HAZ toughness in hot-rolled HSLA line pipe steels containing 0.16 to 0.33 wt.% molybdenum. Although these line pipe steels typically exhibit an acicular ferrite microstructure in the unaffected base metal, at low titanium to nitrogen ratios (about 2), the titanium addition formed thermally stable precipitate particles that restricted the HAZ austenite grain growth during weld thermal cycle and also promoted transformation of austenite to refined bainite packets in the coarse-grained region of the weld HAZ.

From a structural stability standpoint, the greater width and the extensive softening that occur in the weld HAZ during high heat-input welding will not likely pose serious problems in conventional overmatched weldments (wherein the weld metal tensile strength is marginally higher than that of the base metal) because of the plastic constraint offered by the high-strength weld metal and base metal on either side of the softened HAZ (Ref 30, 31). Design engineers often select welding electrodes to provide weld deposits that typically exceed the ultimate tensile strength of the steels being welded. This practice is called overmatching and is used primarily to “insure” or protect the weld deposit from the presence of fabrication-related weld flaws. These flaws when present and subjected to occasional excessive service loads can potentially lead to catastrophic consequences. However, overmatching of heavier gauge high-strength steels may often require the use of preheat, interpass, and post-soak temperature control during welding, which could be quite expensive. In such instances, undermatching (wherein the weld metal tensile strength is marginally lower than that of the base metal) of base metal strength with a lower strength weld metal is a viable option when the following two conditions are simultaneously met: (a) the undermatched weld metal offers very high toughness, particularly exceptional low-temperature impact toughness; (b) undermatching allows the use of cost-effective fabrication practices.

Although overmatching requires the use of relatively high alloy, higher (tensile) strength welding consumables, permissible limits for welding energy input, and the extent of HAZ softening for a particular combination of plate thickness and welding consumable type need to be identified (Ref 32, 33).

In contrast to an overmatched weldment, the greater width and the excessive softening in the HAZ of an undermatched weldment in HSLA-65 steel may show limited fracture resistance and an increased sensitivity to structural instability (Ref 31-33). Despite several fabrication-related advantages in producing undermatched weldments, an excessively undermatched weldment in HSLA-65 steel produced at high weld heat input that exhibits coarse microstructural constituents in the weld metal and a softened HAZ characterized by a lower toughness than the base metal can be “less-forgiving” than an overmatched weldment. Use of a high-toughness, undermatched weld metal may promote a favorable structural response under a variety of extreme loading conditions. Consequently, evaluation of the effects of appropriate combinations of heat input, electrode (or filler metal) type, level of undermatching of strength, weld metal toughness, and service loading to minimize structural instability is essential. This should include the simultaneous effects of welding conditions on the width, strength, and toughness of the weld HAZ.

Further, any variability in base metal yield strength that occurs due to limitations in plate manufacturing control (Ref 13) can potentially complicate meaningful analyses of the results. In this context, one may appreciate the utility of advanced engineering analysis techniques that are based on finite-element and finite-difference modeling. These techniques allow one to perform cost effective and reliable analyses of the effects of several possible variations in base metal and weld metal yield strengths and respective YS/UTS ratios on both the strength and toughness of the weld HAZ and to assess the differences in the mechanical behavior of the weldment systems.

3.2 Cold and Hot Forming

A fabrication standard typically used in naval construction, MIL-STD-1689A, allows cold forming of HSS steels at temperatures below 635 °C (1175 °F) but does not allow cold forming between 204 °C (400 °F) and 371 °C (700 °F), presumably due to a possibility for temper embrittlement. MIL-STD-1689A allows hot forming of HSS steels and weldments at temperatures between 843 °C (1550 °F) and 913 °C (1675 °F) and up to 1204 °C (2200 °F) for HSS steels less than 12.7 mm (½ in.) thick.

The low sulfur content of HSLA steels provides them with excellent cold formability. In fact, TARTAN steels produced (formerly by the Lukens Steel Company, Coatesville, PA) using controlled rolling practices have been successfully demonstrated in producing 180° bends at 1.8*T* bend radius (where *T* is the plate thickness) and 90° bends at 0.8*T* bend radius, with no evidence of surface cracking (Ref 34). Cold working of CR steels increases the yield strength and further increases the YS/UTS ratio. Compared with conventional Q&T steels that exhibit a YS/UTS ratio in the range of 0.78 to 0.85, the CR and CR-AC HSLA steels tend to exhibit a higher YS/UTS ratio up to 0.93 (Ref 5).

Increases in the YS/UTS ratio reduce the strain hardening capacity of a material during service loading and may cause strain localization around internal or surface defects. Such strain localization can potentially impair the structural integrity of a component and cause unstable collapse. This consideration potentially restricts the amount of allowable cold working for HSLA steels, and requires an evaluation of the allowable degree of cold work for a permissible YS/UTS ratio for specific applications.

The CR and CR-AC HSLA steels should only be cold worked at temperatures below Ac_1 (i.e., eutectoid or γ to $\alpha+\gamma$ transformation temperature) and preferably below 650 °C (1200 °F) (Ref 35). It is preferable to have working temperatures about 40 °C (or 100 °F) below the Ac_1 temperature or tempering temperature, whichever is lower. Working at temperatures higher than the Ac_1 temperature or tempering temperature will likely alter (destroy) the base metal microstructure that was carefully obtained through the CR, AC, DQT, or Q&T treatments.

Hot forming of CR or CR-AC type HSLA steels in the austenitic range will eliminate the "as-received" microstructure and properties of the base plate. Consequently, application of such treatments will not be advisable. In exceptional situations, it may be possible to use hot forming procedures that simulate the thermal history, and type and extent of plastic deformation achieved during controlled rolling operations; such procedures will likely include accelerated cooling after hot forming and, when necessary, perhaps require a subsequent low-temperature tempering treatment to restore the microstructure and properties of the base metal. A potential concern includes variability in mechanical properties that might occur across the thickness of the plate after hot forming. Furthermore, as columbium, titanium, and vanadium in solid solution in austenite raise the recrystallization temperature of HSLA steels, excessive scaling may occur due to the increased forming temperature, and this may interfere with some hot forming operations and also increase energy costs.

To address the above concerns, "formability diagrams" may be used (or developed when appropriate) for each of the HSLA steel types and appropriate section thicknesses to provide guidelines for the selection of forming processes and parameters that would provide excellent microstructural stability and high processing efficiency during forming.

3.3 Normalizing and Stress Relieving

MIL-STD-1689A allows normalizing and stress relieving of welded (or formed) components to improve dimensional stability, reduce residual stresses, improve resistance to stress corrosion cracking, and improve the mechanical properties (ductility and toughness) of weldments (Ref 35, 36). As mentioned before, when such a (stress relief) heat treatment is necessary for components fabricated using HSLA steels, the temperature should be limited to 650 °C (1200 °F).

Exposure to heat treatment temperatures higher than the Ac_1 temperature or the tempering temperature will alter the base metal microstructure that was carefully obtained through the CR, AC, DQT, or Q&T treatments. As mentioned previously, use of a higher heat treatment temperature will require the subsequent application of an accelerated cooling (direct quenching) treatment and perhaps a low-temperature tempering treatment to regain desirable combination of microstructures and properties across the weld HAZ and base metal. In welded structures, the application of a high-temperature postweld heat treatment must also be consistent with the metallurgical response of the weld metal.

Whenever the application of a higher heat treatment temperature is impractical, MIL-STD-1689 permits extending the heat treatment time at a permissible lower temperature, down to 510 °C (950 °F), with a holding time of 1 h/5 mm (i.e., 5 h/in.) of thickness. However, this practice has been shown to promote temper embrittlement in conventional medium-carbon high-strength steels, particularly when columbium and vanadium are present (Ref 21, 37). Additionally, the use of extended heat treatment times (over 100 h) or multiple heat treatments with cumulative heat treatment times in excess of 100 h, has been shown to reduce both the yield and ultimate tensile strengths to below specified minimums (Ref 38). An advanced understanding of heat treatment temperature-time effects on the strength and properties of low carbon microalloyed HSLA plate steels produced by CR, CR-AC, DQT, or Q&T techniques is currently unavailable and is required to facilitate the certification of these steels in the as-fabricated or stress-relieved condition.

3.4 Flame Straightening

Flame straightening of low-carbon and low-alloy, high-strength steel structures is often used in naval and commercial shipbuilding operations (Ref 39, 40). The area to be straightened is heated with a flame using either spot, vee (i.e., triangular), or linear patterns, either directly or indirectly to a "dull red" condition about 600-650 °C (1100 to1200 °F) followed by a water spray quench. In the U.S. shipyards, the spot heating pattern is widely used. A variety of spot patterns are used depending on the size of the panel and the amount of distortion to be removed. Linear heating patterns are widely used in European and Japanese shipyards and are considered suitable for achieving required straightening in unrestrained structures (Ref 40). Repetitive heating and quenching of selected areas is performed to correct distortion. An adequate amount of plastic deformation occurs in the distorted member or section prior to distortion removal. When feasible, mechanical straightening (pressing or jacking) is used either independently or concurrently with thermal straightening to achieve an adequate amount of plastic deformation. Flame straightening is known to raise the ductile to brittle transition temperature in several structural steels (Ref 40). Mechanical straightening at high temperatures may reduce strength but does not significantly affect toughness.

The plastic strain that occurs during flame straightening results from linear or volume expansion and contraction that accompany heating and quenching cycles and is thus related to the coefficient of linear or volume expansion. In the temperature ranges traditionally used for flame straightening, despite the presence of significant microstructural differences between HSS and HSLA-65 steels, one may not see appreciable differences in the coefficients of linear and volume expansion at the temperature ranges used for flame straightening. Consequently, the shipyards may be able to apply the flame straightening procedures currently qualified for HSS steels to HSLA-65 steels. This would still require validation testing, and the levels and distributions of residual stresses that may be present in the as-received HSLA-65 plate steels (i.e., prior to welding or fabrication) may show significant variations with regard to HSS on the one hand, and vis-à-vis the plate manufacturing method for the type of HSLA steel plate on the other. Such factors could alter residual stress development after welding or fabrication and potentially complicate the application of flame straightening procedures.

Currently, the flame straightening practices followed in most shipyards are proprietary to the shipyards. These practices are largely empirical or less scientific in nature. Flame straightening technology may be an appropriate candidate for development of expert systems or knowledge-based systems.

4. Summary

The five types of HSLA-65 plate steels show significant differences in microstructure and range of mechanical properties depending on the individual plate manufacturing technique. In particular, the CR and CR-AC type HSLA steels exhibit a higher YS/TS ratio compared with DQT- and Q&T-type HSLA steels. Selection and control of secondary processing methods for these plate steels must attempt to retain or recreate microstructures with performance characteristics similar to those of the base metal. In view of inherent differences in processing, microstructure, and properties of the five types of HSLA-65 plate steels, fabricators must exercise extreme caution when transferring allowable limits of certified secondary fabrication practices from one type of HSLA-65 plate steel to another, even for the same plate thickness.

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