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INVESTIGATION OF STEELS FOR IMPROVED WELDABILITY IN SHIP CONSTR--ETC(U)
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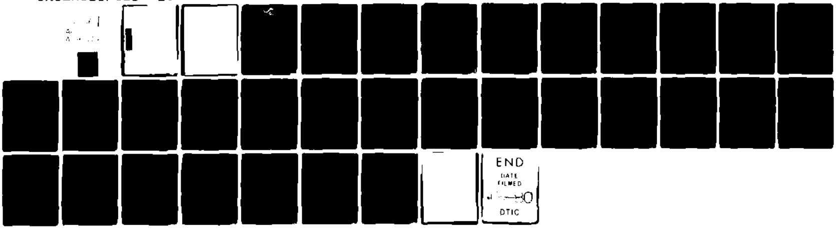

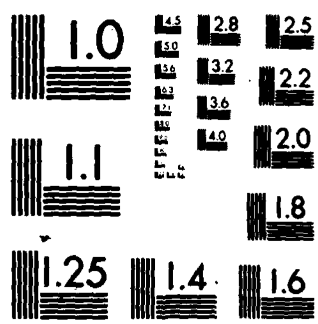

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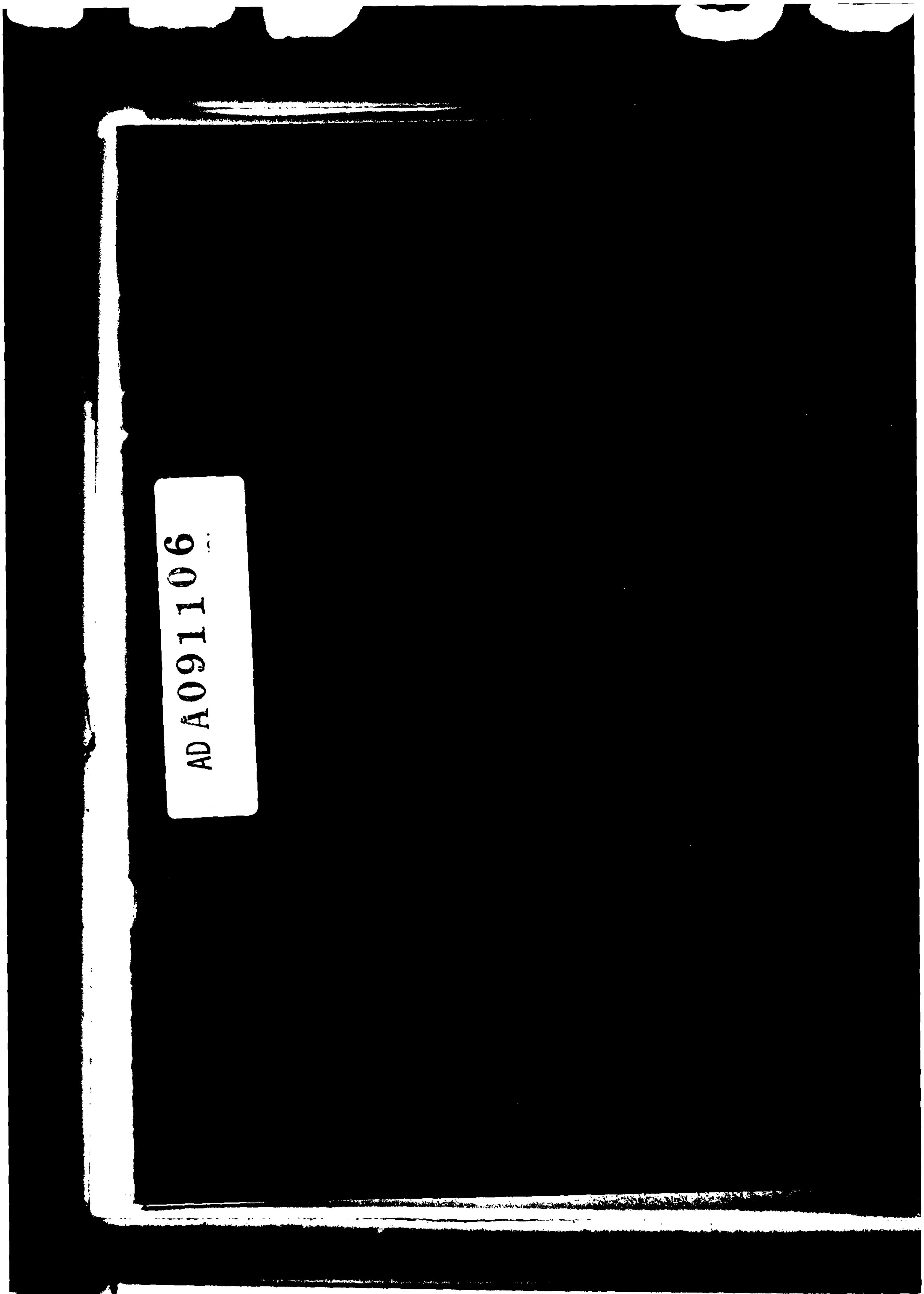

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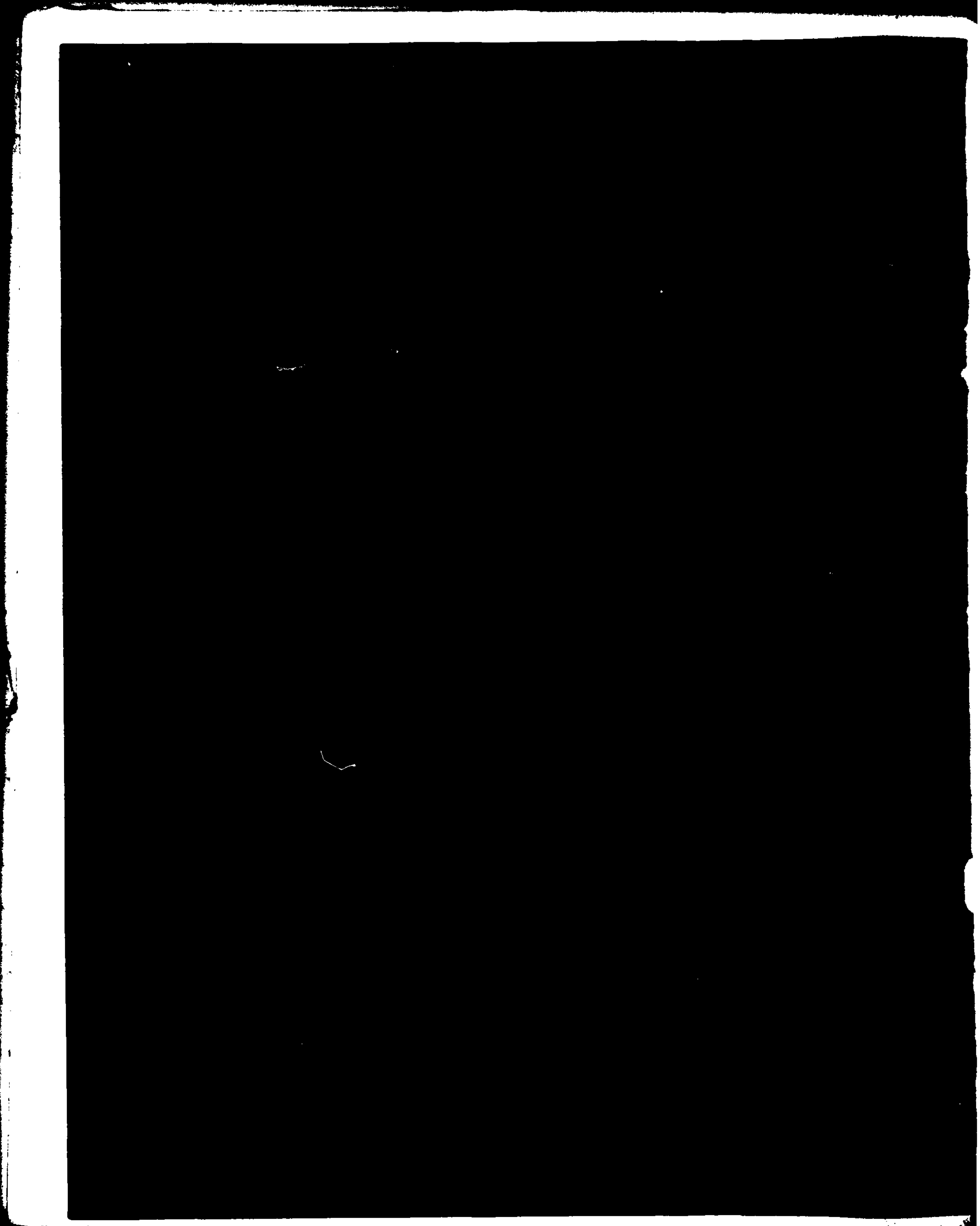

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Ship Structure Committee

An Interagency Advisory Committee
Dedicated to Improving the Structure of Ships

Address Correspondence to:

Secretary, Ship Structure Committee
U.S. Coast Guard Headquarters,(G-M/TP 13)
Washington, D.C. 20593

JULY 1980

SR-1256

Much of the modernization taking place in the world shipbuilding industry in the last decade has centered around the use of new, more efficient welding techniques. The potential increase in productivity with new high-deposition rate welding processes is considerable. However, in order to take full advantage of the benefits of the new welding practices, additional metallurgical control appears necessary for minimizing heat-affected zone and weld-metal property degradation.

The Ship Structure Committee is now sponsoring a project directed toward determining the weld procedure and metallurgical control necessary to develop adequate toughness in the weldment, using high-deposition rate welding procedures. This report describes the first phase of that work.

Henry H. Bell
Rear Admiral, U.S. Coast Guard
Chairman, Ship Structure Committee

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16. Abstract

This report covers the first phase of a three-year study to select the optimum materials and welding parameters to improve resistance to degradation of the heat-affected-zone properties in weldments made with high-deposition-rate processes. Two production steels and twenty 500-pound laboratory heats of steels of varying chemical compositions reflecting low carbon and sulfur content, silicon-aluminum deoxidation practice, globular sulfides, fine titanium nitrides, and treatments with rare earth metals, boron and calcium have been recommended for further examination.

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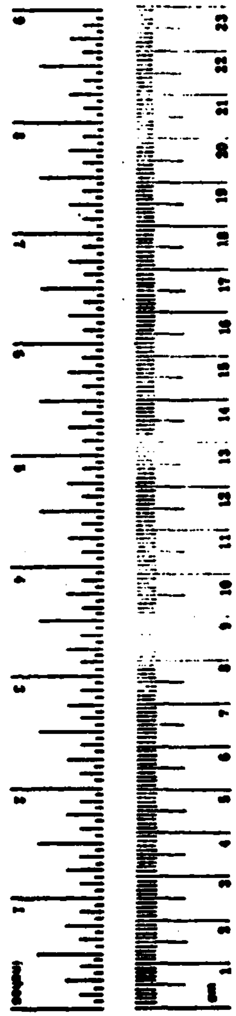
METRIC CONVERSION FACTORS

Approximate Conversions to Metric Measures

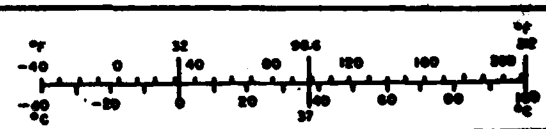
Symbol	When You Know	Multiply by	To Find	Symbol
		LENGTH		
in	inches	*2.5	centimeters	cm
ft	feet	30	centimeters	cm
yd	yards	0.9	meters	m
mi	miles	1.6	kilometers	km
		AREA		
in^2	square inches	6.5	square centimeters	cm^2
ft^2	square feet	0.09	square meters	m^2
yd^2	square yards	0.8	square meters	m^2
mi^2	square miles	2.6	square kilometers	km^2
	acres	0.4	hectares	ha
		MASS (weight)		
oz	ounces	28	grams	g
lb	pounds	0.45	kilograms	kg
	short tons (2000 lb)	0.9	tonnes	t
		VOLUME		
tsp	teaspoons	5	milliliters	ml
Tbsp	tablespoons	15	milliliters	ml
fl oz	fluid ounces	30	milliliters	ml
c	cups	0.24	liters	l
pt	pints	0.47	liters	l
qt	quarts	0.95	liters	l
gal	gallons	3.8	liters	l
ft^3	cubic feet	0.03	cubic meters	m^3
yd^3	cubic yards	0.76	cubic meters	m^3
		TEMPERATURE (exact)		
°F	Fahrenheit temperature	5/9 (after subtracting 32)	Celsius temperature	°C

*1 in ≡ 2.54 (exactly). For other exact conversions and more detailed tables, see NBS Misc. Publ. 286, Units of Weights and Measures, Price $2.25, SD Catalog No. C13.10:286.



Approximate Conversions from Metric Measures

Symbol	When You Know	Multiply by	To Find	Symbol
		LENGTH		
mm	millimeters	0.04	inches	in
cm	centimeters	0.4	inches	in
m	meters	3.3	feet	ft
m	meters	1.1	yards	yd
km	kilometers	0.6	miles	mi
		AREA		
cm^2	square centimeters	0.16	square inches	in^2
m^2	square meters	1.2	square yards	yd^2
km^2	square kilometers	0.4	square miles	mi^2
ha	hectares (10,000 m^2)	2.5	acres	
		MASS (weight)		
g	grams	0.035	ounces	oz
kg	kilograms	2.2	pounds	lb
t	tonnes (1000 kg)	1.1	short tons	
		VOLUME		
ml	milliliters	0.03	fluid ounces	fl oz
l	liters	2.1	pints	pt
l	liters	1.06	quarts	qt
l	liters	0.26	gallons	gal
m^3	cubic meters	35	cubic feet	ft^3
m^3	cubic meters	1.3	cubic yards	yd^3
		TEMPERATURE (exact)		
°C	Celsius temperature	9/5 (then add 32)	Fahrenheit temperature	°F



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Introduction

Background

An appreciable portion of the cost of ship construction can be attributed to welding. For example, it has been estimated that 30 to 50 percent of the total man-hours spent in hull construction is associated with welding.[1)*] Therefore, to reduce costs and to help the United States achieve a competitive position in world shipbuilding, welding techniques are required that will speed up the welding process and still maintain high-quality weldments with a satisfactory degree of strength and notch toughness.

Faster welding translates to the use of high-heat-input welding processes such as multiple-wire submerged-arc (SA), electrogas (EG), and electroslag (ES) welding. Although stick electrodes are still used to a great extent in shipbuilding, the trend is toward higher-heat-input processes, particularly in the larger yards. Multiple-wire SA welding is being employed in shipyard panel lines; ES welding is being used for vertical butt welds for side-shell construction and for butt-welded longitudinal stiffeners.[2,3,4)] The Japanese have used the high-heat-input processes to a greater extent than others and have also employed many more welding engineers in shipbuilding;[2,4)] these factors have certainly enhanced Japan's development of improved welding practices for ships.

The use of high-heat-input welding, however, can cause notch-toughness degradation in the heat-affected zone (HAZ) of weldments.[1)] This is a matter of concern, particularly in critical areas of the ship; such degradation limits the extent to which high-heat-input welding can be used. Thus, the American Bureau of Shipping (ABS) Rules[5)] restrict the use of high-heat-input welding in highly stressed side-shell members such as the bilge strake and sheer strake.6,7) The test primarily used to assess notch toughness in the HAZ is the Charpy V-notch (CVN) test, as specified by ABS.

HAZ-toughness degradation is usually encountered to a greater extent in higher strength shipbuilding steels such as EH36[7)] (51 ksi minimum yield point), and yet where permitted the high-strength grades are being increasingly used in place of the ordinary-strength hull steels (34 ksi minimum yield point).

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* See References.

Several U. S. Government-sponsored projects have been undertaken to extend the use of high-heat-input welding processes in shipbuilding. An exploratory program was carried out by Bethlehem Steel Corporation[7)] in cooperation with the U. S. Maritime Administration (MARAD) on the evaluation of toughness of EG and ES weldments of ship-plate grades ABS, B, CS, and EH36 and ASTM A203 Grade A 2-1/2 percent Ni alloy steel. In that program, useful notch-toughness data were obtained by using various kinds of toughness tests on some of the standard ship steels.

A program is being sponsored by MARAD and monitored by the National Bureau of Standards (NBS)[8)] to determine whether ship-plate steels with improved notch toughness for low-temperature service (LNG tankers) can retain safisfactory toughness in the HAZ when relatively high-heat-input welding practices (up to about 175 kJ/inch) are used. The preliminary results of this study indicate that the best HAZ toughness was obtained for three low-sulfur Cb-treated steels.

An important aspect of HAZ notch-toughness evaluation is the relevance of the evaluation procedure to actual service behavior. Current ABS Rules[5)] require assessment of weldment toughness by testing CVN specimens with the center of the notch located in the weld metal, on the fusion line, at 1, 3, and 5 mm from the fusion line, and in the base metal. A program aimed at determining the relevance of low toughness in the HAZ to the structural performance of ship-steel weldments has been contracted by the U. S. Coast Guard;[9)] this program is being carried out by U. S. Steel and sponsored by the Ship Structure Committee. (SSC).

Good HAZ toughness is an important consideration not only in ships, but also in many other structures. Achieving satisfactory HAZ toughness with higher welding heat inputs is desirable in all applications because it reduces fabrication costs.

Considerable effort has, therefore, been directed toward the development of steels that would exhibit what is judged to be satisfactory HAZ toughness when welded at high heat inputs. The technical literature contains numerous examples of investigations conducted with this purpose in mind. In these studies, standard compositions (ship plate and otherwise) as well as experimental compositions have been used. The proposals and suggestions put forth in the technical literature serve as the basis for the present SSC-funded investigation.

Objective

This investigation is directed toward the development of economical ship-plate steels having improved weld-HAZ notch toughness when welded at high heat inputs. The ultimate objective is to determine which steels achieve the best HAZ toughness and not to achieve a specific toughness level. The HAZ toughness of weldments made with the commonly used lower heat inputs (such as 75 kJ/inch) should also be satisfactory. It is also the objective of this study to identify the metallurgical factors contributing to improved HAZ behavior. The ship-plate steels being considered are those which would be satisfactory for use at ordinary temperatures, and not at low temperatures. Ordinary-temperature applications are those which involve service temperatures ranging from 32°F down to -40°F, the lowest test temperature mentioned in the ABS Rules, Section 43, for Grades E and EH.[5)]

This report on Phase I of this investigation covers a literature survey aimed at identifying the steels and compositional features best suited for high-heat-input welding and the subsequent development of a testing program to be carried out on laboratory heats having compositions such as those described in the literature survey. These aspects of the study are described herein.

Literature Survey

General Considerations

The notch toughness in the HAZ of a weldment is a complex function of many factors, especially the steel composition and the welding heat input. These factors influence the microstructure of the HAZ, which in turn has a major influence on notch toughness. The HAZ is composed of both coarse-grain and fine-grain regions, as well as subcritically heated regions. The HAZ is quite narrow (except when very high heat inputs are used, such as in ES welding), and the relative influence of the different microstructures in the HAZ on overall weldment behavior is difficult to assess. Because of the narrowness of the HAZ, it is difficult to evaluate precisely a particular microstructure in it; this is because the test specimen is most likely to encompass various microstructures that may also include weld metal and/or base metal. Furthermore, there is no agreement as to which laboratory fracture toughness test is best capable of providing an appropriate evaluation of HAZ behavior or of its relevance to overall weldment behavior.

Therefore, one cannot always be certain that a trend established by one testing method will be repeated when another method is used. The tests most commonly reported in the literature for HAZ evaluation are the CVN test (both for evaluation of actual HAZ's and of simulated HAZ's) and the crack-opening-displacement (COD) test. The latter test is very much favored by The Welding Institute.[10,11)]

With regard to the influence of microstructure, grain size is recognized as having a major effect on notch-toughness behavior, with the finer grain region exhibiting the better toughness. Thus, in the HAZ of a weld, it is the coarse-grain region that is generally considered to have the poorest toughness. For ship-plate applications, the weld metal can be selected to provide a desired level of toughness as can the parent plate. The major concern in a ship-plate weldment, therefore, is the coarse-grain region of the HAZ.

In addition to being affected by grain size, HAZ toughness is influenced by transformation products and hardness. Hardness of the HAZ alone is not a good criterion; for example, high-temperature bainite, which has a relatively soft microstructure, has poor notch toughness. In the ship-plate-type steels generally used for ordinary applications, the HAZ microstructure is usually ferrite-pearlite, possibly with some bainite, depending upon cooling rate. A ferrite-pearlite HAZ microstructure can have good notch toughness, as is later discussed. The presence of martensite or bainite in the microstructure generally impairs toughness unless the carbon content is at a suitably low level, but considerable alloy content is required to achieve a low-carbon martensitic or bainitic microstructure.[12)] Furthermore, such steels are often higher strength quenched-and-tempered alloy steels, and under high-heat-input conditions, strength is difficult to maintain in the HAZ of weldments of higher strength steels. Therefore, the literature has not promoted such steels for high-heat-input welding; rather, the emphasis in the literature is on a ferrite-pearlite microstructure for high-heat-input welding, and it seems appropriate to confine this study to a system providing such a microstructure.

HAZ toughness is spoken of in two ways: (1) in terms of the absolute level of toughness in the HAZ and (2) in terms of the amount of degradation in the HAZ relative to the toughness of the parent plate. The better the toughness of the parent plate, the more likely it is that degradation will occur in the HAZ. However, the HAZ may still have enough toughness for the intended application. The question naturally

arises as to what the assessment of HAZ toughness should realistically be based upon. In this study, emphasis is placed upon the absolute level of toughness as a more realistic indicator of HAZ behavior.

In surveying the literature, use was made of the Lockheed Dialogue Data Bases with major emphasis on Metals Abstracts, Engineering Index, and National Technical Information Service. These data bases cover all the major technical literature. The Welding Institute member reports and bulletins were also reviewed.

Information was searched out not only on HAZ toughness in actual weldments, but also on the influence of heating and cooling in simulated welding studies (Gleeble) and in grain-coarsening studies. Considerable attention was given to establishing the compositional features that were reported to result in good notch toughness in the HAZ or in simulated HAZ's when employing high-heat inputs. This is the major thrust of Phase I, the subject of this report. Phase II would then involve the making and evaluation of laboratory heats incorporating the selected compositional features in a laboratory testing program.

Grain-coarsening behavior is discussed first.

Grain-Coarsening Studies

As previously stated, reduced notch toughness in the HAZ is believed to be at least partly associated with the coarse grains that develop in the HAZ, particularly when welding with high-heat inputs. Therefore, the grain-coarsening characteristics of a steel would have a bearing on HAZ toughness.

The theoretical aspects of the control of austenitic grain size by small insoluble particles is explained by Gladman and his associates.[13,14,15] Small particles that are not dissolved serve to pin the grains and restrict grain growth. The critical particle radius, r^*, for restricting grain growth is a function of the product of the matrix grain size, R_o, and the volume fraction of precipitate, f. Grain growth occurs when particle coalescence causes the particle size to exceed r^*. Critical particle size, r^*, decreases with increasing temperature, whereas the actual particle size increases with increasing temperature. Also, r^* decreases as the matrix grain size decreases. It is important to note that as a result of particle coalescence or growth, grain coarsening

occurs at temperatures below those required for complete solution of the precipitate, although the two temperatures may be close.[16] Gladman[15] further points out that when the grain-refining elements that form an alloy precipitate (such as Al and N which form AlN) are present in an amount exceeding their solubility product, the maximum fraction of fine particles is formed when these elements are present in their stoichiometric ratio (approx. 2 for AlN). This is a point to keep in mind when designing compositions to resist HAZ grain growth.

The precipitate phases that have been found to be most effective in pinning grain boundaries to prevent grain coarsening are aluminum nitride and nitrides and carbides of columbium, vanadium, and titanium.[17] For aluminum nitride, columbium carbonitride, and vanadium nitride, grain coarsening can be impeded up to a temperature of about 1830 to 1920°F.[14,16] For titanium nitride (TiN), however, resistance to grain coarsening up to about 2190 to 2370°F can be obtained.[17,18,19] To obtain this higher resistance to grain coarsening with TiN, however, the steel must not be reheated twice through the transformation range.[17,18] Reheating twice causes progressive refinement of the austenite grain size and this, in turn, reduces the r* (Gladman's equation) of effective particles. When the existing particles become larger than r*, their ability to pin the boundaries of the refined grains at the higher heating temperatures is diminished and the grain-coarsening temperature (GCT) drops to about 2000°F. This feature is said to a limiting factor in the production of ingot-cast steels with a fine as-rolled grain size, because at least portions of the ingot usually go through two reheatings during processing to plate (ingot soaking and slab reheating).[19] In the production of continuous-cast slabs, however, only one reheating is involved.

Even with one reheating, the implication is that a subsequent normalizing heat treatment would further refine the austenite grain size and serve to lower the GCT. Or even if a normalizing heat treatment were not employed, the reheating associated with the heat of welding (which, although for only a short time, involves a very high temperature) would constitute a second reheat that might produce coarse grains at temperatures well below 2190 to 2370°F. Although the above work on TiN was aimed at developing a fine grain in as-rolled product (on the assumption that finer ferrite grains would develop from fine austenite grains), the work has a bearing on the grain size developed in a HAZ from welding.

George and Irani[17] recommend that the Ti and N contents should be maintained at relatively low levels to achieve a high GCT since inclusion-size particles of TiN have a negligible effect in controlling grain size. For the most efficient use of Ti as a grain refiner, George and Irani recommend that the Ti and N contents should be maintained at levels where the solubility product for precipitation in the liquid is not exceeded. If the solubility product is exceeded, the TiN particles that form in the liquid state will be relatively large and ineffective for restricting grain growth. The low solubility product of TiN (about 10^{-6} at 2370°F)[20] thus translates to very low Ti and N contents.

It is further cautioned that it is the fine TiN particles that are responsible for a high GCT, provided there is no TiC present.[17] The TiC goes into solution more easily and enhances the growth of TiN particles. Thus, it is claimed that the nitrogen content should be in excess of the stoichiometric amount necessary to completely combine with Ti. Since the stoichiometric ratio of Ti to N for TiN is about 3.5, the Ti to N ratio should be somewhat less than 3.5, but close enough to develop the greatest number of fine precipitates.

Matsuda and Okumura[21] show that about 0.005 percent TiN can be dissolved by heat treating at 2280°F for 10 hours (a very long time) or at 2460°F (a very high temperature) for 1 hour. This TiN can then be reprecipitated upon reheating. The TiN that reprecipitated was found to be coarser (0.01 μm or 100Å) after heating to 2100°F at a rate of 360°F per minute than that precipitated (0.005 μm, 50Å) after heating at a lower rate of 2.9°F per minute. The smaller TiN particles were expected to better inhibit grain coarsening upon subsequent rapid heating (to 2460°F in 1 second) to simulate heating from welding, but instead, the coarser particles (from prior heating at 360°F per min) provided the finer austenite grain size. The austenite grain growth in the samples with the finer TiN precipitates was attributed to dissolution of some of the very fine particles and to Ostwald ripening or particle growth of the remaining coarser particles with increased holding time at an elevated temperature.

O'Donnell[22] and George, et al.,[23] found that additions of both titanium and columbium to cast structural carbon steels produced a lower GCT than that obtained with the titanium addition alone (2175° vs 2290°F).[22] Vanadium, however, was found to either increase the GCT in the presence of titanium or not to change it at all.

Wyszkowski[24] reported that rapid heating of 0.40 percent carbon aluminum-killed steels stimulated rapid grain growth (particularly when supersaturated with AlN) because of the decrease in size of the first austenitic grains. Vanadium and titanium additions were particularly effective in the prevention of rapid grain growth.

In summary, the above studies of grain-coarsening indicate that fine stable precipitates result in effective pinning of grain boundaries and that fine TiN particles provide the highest GCT. However, the GCT obtained with TiN is very much affected by subsequent heating practices, and multiple reheats should generally be avoided (so that Ostwald ripening or a reduction of the critical particle size do not serve to lower the GCT). To achieve a high GCT, the Ti and N contents should be maintained at low levels, and the ratio of Ti to N should be less than the stoichiometric ratio of 3.5. In Ti-treated steel, Cb appears to lower the GCT, whereas V does not. Thus, to obtain a strength increase in Ti-treated steel, it would appear that a vanadium addition rather than a columbium addition would be more beneficial.

HAZ Notch Toughness

Carbon Steels - As pointed out in the introduction, high-heat-input welding processes are being employed for welding low-strength ship-plate grades. Very little information, however, is available on the notch toughness of the HAZ. In a project report by Bethlehem Steel[7], 1-inch-thick as-rolled semikilled ABS Class B steel was shown to exhibit minimum CVN impact values in the HAZ of 8 to 10 ft-lb at 32°F when EG-welded (638 kJ/in heat input) and ES-welded (380 kJ/in). The crack-starter drop-weight nil-ductility-transition (NDT) temperatures of the HAZ were 20 and 30°F, respectively. For 1-1/4-inch-thick normalized Si-Al-killed ABS CS steel, the minimum CVN energy absorption in the HAZ when similarly welded with about the same heat inputs was 33 to 42 ft-lb at -4°F, and the NDT temperatures were -10°F (EG) and -40°F (ES). These values for the HAZ of the CS steel are considered very good, although significant degradation occurred in some instances in comparison with the excellent base-metal toughness.

In recent U. S. Steel tests, the HAZ of an ES weld in laboratory-melted as-rolled 1-inch-thick ASTM A36 steel (0.20 C, 0.95 Mn) made by a Si-Al-killed deoxidation practice exhibited an average CVN value of 49 ft-lb at 0°F.[25] Thus, even under high-heat-input welding conditions, ordinary-strength structural steels can develop quite good notch toughness in the HAZ

despite the grain coarsening which occurs under high-heat-input conditions.

Columbium-Treated Steel - Dolby, et al., state that Cb-treated steels (generally 0.02 to 0.05% Cb) show an impairment of HAZ toughness with high-heat-input welding of 125 kJ/inch and greater.[26,27,28] This impairment is attributed to the Cb suppressing the formation of proeutectoid ferrite and promoting the formation of upper bainite.[26] With ES welds on 0.16 C, 1.3 Mn steels, the CVN 20-ft-lb temperature was about 32°F in the HAZ of Si-Al-killed 0.02 Cb steel versus about -22°F in the HAZ of Si-Al-killed steel without Cb.[28] Crack-opening-displacement (COD) tests on these same steels showed an 0.1-mm COD at -75°F for the Si-Al-killed Cb steel versus <-165°F for the Si-Al-killed steel without Cb.

Other investigators, such as Hannerz,[29,30] also found increased embrittlement in the HAZ of Cb-bearing steels as the heat input increased (or the cooling rate of the HAZ decreased). Also, embrittlement from Cb occurred in 0.03 C steel as well as in 0.19 C steel (when heat inputs above about 80 kJ/inch were used). Similar embrittlement resulted from high heat inputs in low-carbon Cb-containing steels reported by Kaee et al.[31,32] At lower heat inputs, Cb does not seem to impair the HAZ toughness, and it is roughly estimated that the critical heat input may be about 75 kJ/inch.[33,34,35,36]

Benter[37] verifies that the HAZ toughness (as judged by CVN tests) of a Cb-treated steel similar to ABS EH is impaired when welding with EG or ES processes. However, in crack-starter explosion-bulge tests conducted on the 2-inch-thick parent plate and on ES-welded plate, a fracture-transition-elastic (FTE) temperature of over 30°F was observed for the parent plate and about 25°F for the ES-welded plate, with no cracking in the vicinity of the HAZ where the CVN 20-ft-lb temperature averaged about 60°F. Such data imply that overall weldment behavior is not as bad as may be indicated by small-scale tests of the worst portion of the HAZ. This is a debatable stance, for data developed in other contract work[9] (not yet reported) indicate that the zone with the poorest notch toughness may indeed determine overall behavior.

It was previously mentioned that in a MARAD-sponsored program[8] relatively good HAZ toughness was observed for three low-sulfur Cb-treated steels welded with heat inputs up to about 175 kJ/inch. One of these steels was Ca-treated and another was rare-earth-metal-(REM) treated. The improved notch-toughness behavior of these steels may appear exceptional

in view of the general literature viewpoint that Cb impairs HAZ toughness at higher heat inputs; however, the favorable response of these Cb-treated steels may be associated with their low-sulfur content and/or other special addition agents.

Vanadium-Treated Steel - Notch toughness in the HAZ progressively deteriorates as the vanadium content increases above 0.10 percent and as the heat input increases.[30,38] However, the notch toughness in the HAZ does not deteriorate at vanadium contents below 0.10 percent, even with high-heat inputs.[34,38,39] There may be a slight improvement in the HAZ toughness with a vanadium addition of about 0.05 percent. Thus, a small vanadium addition should be useful in achieving additional strength in the base metal.

Titanium-Treated Steel - A U. S. patent issued to Kanazawa, et al.,[40] indicates that with heating cycles corresponding to heat inputs of more than 127 kJ/inch (usually around 250 kJ/inch), a marked improvement in the toughness of a simulated HAZ can be obtained by treating the steel with a small amount of titanium (around 0.015 to 0.04%). The improved toughness results from fine TiN precipitates (smaller than 0.05 μm in size) that inhibit austenitic grain growth in the HAZ. To obtain this improvement in toughness, the steel ingot must be cooled at a rate ≥9°F per minute down to 2010°F and must not be reheated more than one time in subsequent steps above 2010°F. These restrictions are similar to those mentioned previously in the section on grain-coarsening studies. In that section however, it was noted that the steel was not to be reheated twice through the transformation range. Kanazawa also points out that the Ti to N ratio should be ≤3.5, which is the stoichiometric ratio. Kanazawa states that the most desirable Al range is 0.0005 to 0.015 percent; but Al contents in this range are not generally considered adequate to fully kill the steel and might make Ti recovery erratic. However, in many of the examples cited the steels had Al contents over 0.02 percent.

Boron (0.001 to 0.006%) is another addition agent used in several of the Kanazawa steels. Boron also forms a nitride (which could interfere with the TiN reaction), but the tendency to form TiN is stronger. The invented steels contain about 0.12 C and 1.30 Mn. Sulfur content is not mentioned in the patent,[40] but other references to this development indicate that the sulfur content is low.[41-43] For most of the invented steels, the plate product is quenched-and-tempered (another reheating through the critical), but a few examples of normalized or as-rolled plate are given.

The improved HAZ notch toughness is attributed to the precipitation of fine TiN (<0.05 μm) in the HAZ, which serves to refine the HAZ microstructure.[40-43] This fine TiN is claimed to do two things — inhibit austenitic grain growth and stimulate ferrite transformation. These actions produced a refined HAZ structure with a smaller unit facet diameter for brittle fracture.

The notch toughness of the Kanazawa steels was evaluated by CVN tests at 32°F on material simulating the HAZ (Gleeble-type evaluations). This test temperature is fairly high, and it is unfortunate that information on behavior at lower temperatures was not reported.

Gondo, et al.,[44] claim that with suitable heat treating and processing steps, fine TiN precipitates can be produced in wrought product made from ingot-cast steel, which, in the cast condition, contains coarse TiN. They state that adequate TiN can be taken into solution (at least 0.004%) at reheating temperatures of 2280 to 2550°F for subsequent precipitation as fine TiN. Gondo usually uses 2460°F as a soaking temperature, which is somewhat high for slab reheating. Kanazawa, et al.,[42] indicate that about 0.004 Ti is taken into solution with a short-time thermal cycle having a peak temperature of 2370°F. Matsuda and Okumura,[21] however, indicate that several hours would be needed. Gondo prescribes specific processing and heat-treating steps to subsequently reprecipitate fine TiN.

REM-Boron-Treated Steel - Funakoshi, et al.,[45] and Sanbongi, et al.,[46] claim that steels containing proper amounts of REM and boron show excellent notch toughness at the weld bond in ES-welded joints. Processing restrictions are not cited. The recommended amounts of REM and B are about 0.02 to 0.03 percent and 0.002 to 0.0035 percent, respectively. The S content in the steels studied was low, about 0.005 percent. The combination of REM and B was effective in developing improved HAZ toughness, whereas either element alone was much less effective. It was noted that REM raises the proeutectoid ferrite-transformation temperature and B slows the nucleation of proeutectoid ferrite. The combined addition raises the proeutectoid ferrite-transformation temperature range more than does the addition of REM alone. The formation of fine ferrite grains at the weld bond in REM-B steel is attributed to BN accelerating the nucleation of fine ferrite grains inside the prior austenite grains, with REM (in the form of ultrafine REM oxysulfides) contributing by providing nucleation sites for BN.

Calcium-Treated Steel With Nitrides - Y. Kasamatsu, et al.,[47,48] describe another method of achieving finer grain size and good HAZ toughness in high-heat-input welds when using the usual processing steps for ingot-cast steel. In this method, Ca (or Mg) is employed along with Ti (or Zr) to develop fine precipitates. It is stated that very fine inclusions containing Ca (or Mg) form, and these also act as seeds for precipitating TiN. Both kinds of inclusions act to prevent grain growth, and the formation of a fine ferrite-pearlite structure (and bainite) is favored. The preferred composition range for Ti is 0.008 to 0.020 percent and for N is 0.002 to 0.008 percent. A range for Ca is also cited, but little if any Ca dissolves in the base metal; instead, its presence is observed in inclusions. A further addition of cerium (the major constituent in REM) is said to provide additional benefits in toughness in the HAZ at a distance of 2 to 4 mm from the fusion line; this is reported to result from the formation of fine, spherical particles of Ce (or REM) sulfides. Other alloy-addition elements such as Cb, V, B, Ni, Cr, and Mo are also noted, mainly because they improve strength.

Steel With Low Silicon Plus Boron - Y. Kawaguchi, et al.,[49] claim that a low-silicon content (<0.10%) plus a boron addition (0.0015 to 0.0027%) in a steel containing 0.005 to 0.025 percent N improves toughness in the weld bond made with high-heat-input welding. The low-Si content promotes the formation in the HAZ of polygonal ferrite within the grains and suppresses the formation of proeutectoid ferrite in the prior austenite grain boundaries. The low-Si content also reduced hardenability and eliminated the formation of martensite islands in the V-B steel studied. However, low-Si contents may contribute to weld metal porosity, particularly when using basic welding fluxes. Boron is said to improve weld-bond toughness through the formation of fine ferrite at the prior austenite grain boundaries when the B/N ratio is controlled between 0.2 and 0.6.

Carbon Content - It is generally recognized that a relatively low-carbon content produces the best notch toughness in both the base metal and the HAZ. A higher carbon content tends to produce higher hardness and martensite in the HAZ, which detracts from notch toughness. This is true for ordinary carbon steel as well as for HSLA steels,[12,50,51] and it applies to high as well as low-heat inputs.

In the steels with promising HAZ toughness described above (for example, the Ti-treated steels), the carbon content is frequently around 0.12 percent. This carbon level is a reasonable compromise for achieving a good combination of toughness and strength.

Deoxidation Practice - Steel deoxidation with silicon and aluminum produces the best toughness in the base plate (as in ABS Grades D and E), although it is questionable whether the Al benefits the toughness of the coarse-grained HAZ.[26,27,28] The aluminum, however, by combining with free nitrogen, helps to minimize strain-aging embrittlement that can occur in the subcritical HAZ.[50,52] Together with its favorable effect on the base metal, aluminum is believed to have an overriding positive benefit on weldment behavior.

Nearly all the promising compositions previously discussed were Si-Al-killed steels. Kanazawa, et al.,[40] recommended an Al range of 0.0005 to 0.015 percent for Ti-treated steel, but then proceeded to use higher amounts in many of the compositions described as invented. Al, which more fully deoxidizes the steel, provides for a higher recovery of highly reactive elements such as Ti, Ca, and REM which can then carry out their allotted functions.

Sulfur Content and Sulfide Shape - Reducing the sulfur content of a steel or adding elements, such as REM or Ca that will form globular sulfides rather than stringers, raises the CVN shelf energy, particularly in the transverse direction where the CVN shelf energy may be low due to straightaway rolling. The favorable influences of increased cleanliness and globular inclusions carry over into the HAZ, and generally higher CVN values may be found in the HAZ after such treatment.[8,27,51] The major effect will, undoubtedly, be to raise the shelf energy (which is reflected along the whole CVN curve, but to a lesser extent as the bottom of the transition curve is approached) rather than to have much influence on, say, the CVN 15-ft-lb temperature, which is near the bottom of the transition curve.

Jesseman and Schmid[36] point out that a REM addition to ABS EH32 steel (not containing Cb or V) did not improve notch toughness in the coarse-grained region of the HAZ (except at the highest testing temperatures) when welding with 50 and 75-kJ/inch heat input. This steel, as well as the reference steel without REM, contained about 0.008 percent S. Comparison with a steel with higher sulfur content was not available.

Summary of HAZ Toughness Survey - The compositional features having a favorable effect on HAZ toughness (actual or simulated heat-affected zones) when welding with high-heat inputs appear to be as follows:

1. A low-carbon content.

2. A Si-Al deoxidation practice.

3. A low-sulfur content.

4. Globular, rather than stringered, sulfides (by inclusion-shape control). This, however, is apparently less important if the sulfur content is already low.

5. Fine titanium nitrides in the microstructure to inhibit grain growth. Such nitrides when developed are sensitive to subsequent heating steps and will not as effectively restrict grain growth if the particles subsequently coarsen or if the critical particle size decreases.

6. Treatment with REM and boron to accelerate the nucleation of fine ferrite grains within the prior austenite grains.

7. Treatment with calcium in combination with Ti to develop fine precipitates for restricting grain growth. A Ce addition provides further benefits.

8. Treatment of a low-silicon steel with boron to promote the formation of fine polygonal ferrite.

Items 5 through 8 are aimed at developing a fine microstructure in the HAZ by means of fine precipitates. Fine precipitates can be effective in developing a finer microstructure in two ways. They can (1) serve to raise the grain-coarsening temperature in order to reduce the austenitic grain size in the HAZ and to reduce the width of the coarse-grained region that develops and (2) act as nucleation sites for the development of a fine ferrite-pearlite structure.

With regard to Cb or V additions, neither seems to help the HAZ toughness under high-heat inputs, but the literature generally indicates that V would be less harmful and, hence, might be the better element to add for increased strength. Possibly, around 0.05 percent V might even improve the notch toughness slightly.

Processing

The processing of plate product (for example, how it is hot rolled and whether or not it is heat treated) has a substantial effect on the base-plate notch toughness, but may not have much effect on HAZ toughness. Regardless of base-plate microstructure and toughness, the high heat of welding may develop the characteristic coarse-grained region in the HAZ. Thus, whether the plate is hot rolled at a high temperature or heat treated, the HAZ toughness may be similar.

However, as the Japanese literature in particular points out, a fine precipitate can have a favorable effect on austenitic grain size and on the fineness of the transformation structure in the HAZ; an accompanying sensitivity of such particles to processing steps is sometimes cited.[40)] In such instances, therefore, processing, and its influence on particle size, does make a difference.

In most of the studies on HAZ toughness, the parent plate was in the heat-treated condition—either normalized or quenched-and-tempered. This certainly seems advisable in order to assure that the parent plate is as tough or tougher than the HAZ. Heat treatment almost becomes necessary if temperatures as low as -40°F are to be considered. Even at higher temperatures such as 0 to 32°F, good notch toughness in the base metal may have a beneficial carry-over effect on the HAZ toughness. This could be important in weldment testing (where various microstructures are necessarily tested), but would not be a factor in simulated-HAZ Gleeble tests.

Development of Testing Program

Testing Procedures

No universally accepted laboratory fracture-toughness test is available for evaluating HAZ behavior. Because of the different microstructures in the HAZ of an actual weld, problems arise as to notch location and the influence on toughness of the various microstructures that a propagating crack encounters.

Gleeble specimens can be heat-treated to simulate any portion of a HAZ and can then be tested as CVN specimens for evaluation of notch toughness. Although individual Gleeble CVN specimens do not contain the various microstructures that are characteristic of an actual weld, they can serve to rank

the relative toughness performance of specific locations in a weld HAZ, thus indicating which compositions are likely to exhibit high HAZ toughness. Therefore, Gleeble CVN specimens have been selected to identify those compositions that show promise of developing good HAZ toughness. (Gleeble specimens were frequently used in the studies referred to earlier.)

Two simulated HAZ conditions have been selected for evaluating all steels. Both conditions involve heating to a high peak temperature (about 2500°F) and heating and cooling to simulate heat inputs of about 180 and 800 kJ/inch. The 180-kJ/inch heat input simulates a high-heat input for a two-pass SA weld of 1-inch plate, and the 800-kJ/inch heat input simulates ES welding. The Gleeble tests will serve as screening tests to identify the most promising compositions. CVN transition temperature behavior will be determined for simulated HAZ conditions and for the base metals as well.

Weldments will then be made for the selected promising compositions and for two reference steels (all as 1-inch-thick plate). Three heat inputs will be used, approximately 75, 180, and 800 kJ/inch. The 75 kJ/inch heat input constitutes present-day typical practice. The welds will be longitudinal so that transverse tests may be conducted. The transverse notch-toughness tests will be CVN traverses (for establishment of transition behavior) at five locations (weld metal, fusion line, and 1, 3, and 5 mm from the fusion line) in accordance with ABS[5)] and USCG requirements. Consideration will be given, however, to testing at 1, 4, and 7 mm from the fusion line on the ES welds because the HAZ extends further into the base metal. In addition, transverse crack-starter drop-weight NDT tests[53)] will be performed in two locations (base metal and HAZ) as originally proposed.

Standard tension tests will also be performed, as well as other tests and examinations to help establish the reasons for the behavior obtained.

Heat Treatment

All the plate product (except as will be noted) will be normalized because normalizing generally provides the good base-metal notch toughness that may well be required for critical locations in a ship. For example, normalizing the steels selected should develop adequate base-metal toughness to meet a 20 or 25 ft-lb toughness requirement at -40°F, the lowest temperature of interest in this study. Controlled rolling could produce the desired toughness in many instances,

and this would be a possible alternative in production. However, because this procedure is difficult to simulate consistently in the laboratory; normalizing is judged to be the best procedure for this study.

Materials

Plate samples will be obtained from two production steels for inclusion in this study as reference materials. One steel is a calcium-argon-blown Cb-treated ABS V-051 steel on which considerable background work has been done in the MARAD/NBS study.[8] This steel met 20 ft-lbs (CVN) at -60°F in the HAZ with heat inputs as high as 150 kJ/inch for 1-inch plate (2 passes). The other steel is an ABS Grade CS steel, which has the same composition as ABS DS and which also meets the composition requirements for ABS Grade D.

It is planned to make and evaluate at least twenty 500-lb laboratory heats. The heats will be vacuum-melted to avoid a high frequency of inclusions. The aim compositions are based upon the literature survey and are listed in Tables I through III.

Table IA shows two laboratory steels that are intended to duplicate the reference production heats. With regard to the remaining steels, the base composition selected was 0.12 C, 1.35 Mn, and 0.006 S (Steel 3, Table IB). This C-Mn combination frequently appeared in the literature. The relatively low-carbon and high-manganese contents would favor better notch toughness both in the base metal and in the HAZ. The low-sulfur content has also been shown to favor good toughness. Also, all the steels except Steel 9 (Table IIA) are Si-Al-killed.

Small amounts of the so-called residuals, Cu, Ni, Cr and Mo, have been added to all the steels except those containing Cb or V. These additions were made to help increase the strength and counteract the effect of the relatively low-carbon content. Also, the nitrogen contents of all the steels was slightly higher than is typical of OH or BOP steels, and this too would tend to increase strength. These conditions of composition may automatically occur in electric-furnace steels in which the residuals are picked up from the scrap additions. The use of residuals and a somewhat higher nitrogen content is frequently employed to achieve the 50-ksi minimum-yield-point requirement in ASTM Specification A537 (in the absence of Cb or V). The aim yield point in this study is about 50 ksi.

TABLE IA - Reference Lab Heats

Steel Type	No.	C	Mn	S	Si	Al	Cb	N	Ca
		Composition, percent							
ABS V-051	1	0.13	1.38	0.006	0.16	0.03	0.027	0.004	Yes
ABS CS	2	0.14	1.25	0.020	0.22	"	-	"	-

TABLE IB - Base Steel + Effects of S, Ca, Cb, V, REM

Type	No.	C	Mn	S	Si	Al	Cu	Ni	Cr	Mo	V	Cb	N	Ca	REM
Base	3	0.12	1.35	0.006	0.30	0.03	0.20	0.15	0.15	0.04	-	-	0.008	-	-
Base	4	"	"	0.020	"	"	"	"	"	"	-	-	"	-	-
Base + Ca	5	"	"	0.006	"	"	"	"	"	"	-	-	"	Yes	-
Base + Ca + Cb	6	"	"	"	"	"	-	-	-	-		0.025	"	"	-
Base + Ca + V	7	"	"	"	"	"	-	-	-	-	0.03	-	"	"	-
Base + REM + Cb	8	"	"	"	"	"	-	-	-	-	-	0.025	"	-	0.030

TABLE IIA - Ti-N Combinations

Type	No.	C	Mn	S	Si	Al	Cu	Ni	Cr	Mo	N	Ti	B
TiN	9	0.12	1.35	0.006	0.30	0.01	0.20	0.15	0.15	0.04	0.005	0.013	-
TiN-Al	10	"	"	"	"	0.03	"	"	"	"	"	"	-
TiN-Al	11	"	"	"	"	"	"	"	"	"	0.010	0.025	-
TiN-Al+B	12	"	"	"	"	"	"	"	"	"	"	"	0.003

TABLE IIB - REM-B Combinations

Type	No.	C	Mn	S	Si	Al	Cu	Ni	Cr	Mo	N	Ti	B	REM
REM-B	13	0.12	1.35	0.006	0.30	0.03	0.20	0.15	0.15	0.04	0.005	-	0.003	0.30
REM-B-N	14	"	"	"	"	"	"	"	"	"	0.010	-	"	"
REM-B-Ti	15	"	"	"	"	"	"	"	"	"	0.005	0.015	"	"

TABLE IIIA - Ca-Treated TiN Steels

Type	No.	C	Mn	S	Si	Al	Cu	Ni	Cr	Mo	N	Ti	B	Ca	REM
TiN-Ca	16	0.12	1.35	0.006	0.30	0.030	0.20	0.15	0.15	0.04	0.008	0.020	-	Yes	-
TiN-Ca	17	"	"	"	"	"	"	"	"	"	0.006	0.015	-	"	-
TiN-Ca+ REM	18	"	"	"	"	"	"	"	"	"	"	"	-	"	0.030
TiN-Ca+ B	19	"	"	"	"	"	"	"	"	"	"	"	0.003	"	-

TABLE IIIB - Low-Silicon Plus Boron Steel

Type	No.	C	Mn	S	Si	Al	V	N	B
Low Si-B	20	0.12	1.35	0.006	0.04	0.030	0.06	0.007	0.003

Table IB shows the effects of various elements on the base composition. Steel 4 is included to show the influence of higher sulfur in this particular base composition. Steel 5 is calcium treated to determine whether the rounding of the sulfides brought about by the calcium (inclusion-shape control) will provide further improvement to the already low-sulfur Steel 3. Steel 6 is similar to Steel 5, but contains Cb instead of residuals for strengthening. The Cb will probably give more strengthening than the residuals. It will be determined whether Cb in this Ca-treated heat provides the good notch toughness (at a fairly high-heat input) exhibited by the ABS V-051 production heat (similar in composition to Steel 1). Steel 7 will be used to determine the influence of V in Ca-treated steel. Steel 8 is similar to Steel 6, but contains REM instead of Ca so that the relative effects of the shape-controlling addition agents, REM and Ca can be determined.

Table IIA shows the aim compositions for four steels with Ti-N additions. The purpose of adding TiN is to develop fine nitrides to minimize austenitic grain growth, as discussed in the Japanese literature and in the studies on grain coarsening. The aim Ti to N ratio for the four steels is 2.5, somewhat less than the stoichiometric ratio of 3.5, as advocated.[17] Steel 9 has an Al content of 0.01 percent, which is lower than that of the other steels, but within the recommended range of 0.0005 to 0.015 percent. It may be difficult to achieve this Al content. The other three steels have an Al content typical for Al-killed steels and, as previously noted, an Al content of about 0.03 percent is frequently obtained in the invented steels proposed by the Japanese. Steels 11 and 12 have higher Ti and N levels than Steels 9 and 10. Steel 12 also contains boron. Boron, along with titanium, is often noted in the Japanese-invented steels although its presence is not explained. Titanium and boron are frequently observed together when a Grainal addition is made to a heat as a means of adding boron. The Grainal addition contains Ti to help protect the boron (Ti combines with N). These four steels will be processed differently from all the other steels, as will be described later.

Table IIB shows aim compositions based upon the REM-B-treated steels that have been described. Both BN and REM are intended to promote the formation of fine ferrite grains and thus develop good HAZ toughness. Steel 14 has a higher nitrogen content than Steel 13. The Japanese patent[46] on this type of steel also covers other added elements such as titanium. Titanium is added to Steel 15 to assess any interaction that may occur.

Four Ca-treated Ti-N steels are shown in Table IIIA. As described in the literature survey,[47,48] both Ca and Ti are intended to develop fine precipitates which, in turn, favor the development of a fine ferrite-pearlite microstructure in the HAZ. Steels 16 and 17 have slightly different Ti and N levels. REM was added to Steel 18 because it reportedly improves toughness 2 to 4 mm from the bond line. Steel 19 is Steel 17 plus boron. Boron promotes ferrite formation[45,46] as does calcium,[47,48] and it seems worthwhile to determine whether there is any synergistic effects from this combination.

Steel 20 (Table IIIB) is the recommended composition (low Si plus B) that was described by Kawaguchi, et al.,[49] in the literature survey.

Processing of Laboratory Steels

In the Japanese literature, major emphasis is placed upon the development of fine nitrides for achieving a fine austenitic grain size or a fine transformation structure, and the sensitivity of such nitrides to processing steps is cited.[40] To develop the fine nitrides and improved toughness, the steel ingot must be cooled at a rate ≥9°F per minute down to 2010°F and must not be reheated more than once above 2010°F in subsequent processing (such would be the case for continuous-cast product). This practice will be adhered to in this evaluation for the Ti-N Steels 9 through 12 in Table IIA. The 500-lb ingots cool at about 70°F per minute, well over 9°F per minute. After the ingots are cooled to room temperature, they will be reheated to 2350°F and rolled directly to 1-inch plate (continuous-cast slabs would also be fairly rapidly cooled to room temperature for conditioning and then reheated for rolling to plate).

Gleeble CVN tests will then be conducted on plate in both the as-rolled and the normalized conditions to determine whether the normalizing treatment has any influence on subsequent simulated HAZ behavior. If little or no effect of normalizing is observed in these tests, normalized product will be subsequently tested.

The remaining 16 steels are not supposed to be as sensitive to processing parameters and can presumably be made by the usual ingot-making procedures. Although cooling rates of full-sized ingots will not be duplicated on the laboratory heats, the mill process steps for reduction to plate will be simulated. All the remaining steel ingots will be air-cooled

to room temperature, heated to 2350°F, and rolled to approximately 5 inches thick and air-cooled (to simulate slabbing). The 5-inch product will then be reheated to 2350°F (as pointed out previously, at least portions of an ingot usually go through two reheatings during processing to plate) and rolled to 1-inch-thick plate and air-cooled (to simulate plate processing). All the product will then be normalized because normalizing generally provides the good base-metal notch toughness that may well be required for critical locations in a ship.

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